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of the

American Foundrymen's Society

Proceedings of the

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Atlantic City, N. J.

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OFFICERS

of the

AMERICAN FOUNDRYMEN'S SOCIETY

(INCORPORATED)

(Fiscal Year Ending June 30, 1953)

- *President, I. R. WAGNER, Electric Steel Castings Co., Indianapolis, Ind.
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^{*} Member, Executive Committee

SUMMARY OF PROCEEDINGS OF THE INTERNATIONAL FOUNDRY CONGRESS

THE INTERNATIONAL FOUNDRY CONGRESS AND SHOW (56th Annual Meeting of A.F.S.) was held in Atlantic City, N. J. May 1-7, 1952. The Biennial Foundry and Allied Industries Show was staged by the Society in conjunction with the Congress. Exhibits were displayed in Convention Hall with over 260 companies showing their products.

Credit for the excellent reception and hospitality received by those attending is given to members from the Chesapeake, Metropolitan and the Philadelphia Chapters of A.F.S. under the General Chairmanship of C. V. Nass, Beardsley & Piper Div., Pettibone Mulliken Corp., Chicago. This Committee excelled in its splendid work and hospitality as host to the thousands of members and guests attending.

Approximately 14,000 foundrymen, metallurgists, and representatives from allied industries throughout the United States and from 33 countries outside of continental United States attended the International Foundry Congress. During the seven days of the Congress, over 60 sessions were held for presentation of over 125 timely technical papers and committee reports. Two Gray Iron Shop Course Sessions and four Sand Shop Course Sessions were held. Two of the latter sessions were sponsored jointly by the Sand and the Brass and Bronze Divisions. A Round-Table Luncheon was sponsored by each Division. The Charles Edgar Hoyt Lecture was delivered by John S. Bugas, Ford Motor Co., Dearborn, Mich. on the subject "Industry's Responsibility to Youth." There were ten Exchange Papers presented at the Congress.

A summary of the sessions held follows:

OFFICIAL OPENING CEREMONY Thursday, May 1, 11:45 A.M.

Presiding-C. V. NASS, Chairman, International Congress Committees.

BRASS AND BRONZE ROUND TABLE LUNCHEON Thursday, May 1, 12:00 Noon

Presiding-B. A. Miller, Baldwin-Lima-Hamilton Corp., Phil-

Co-Chairman-H. L. Smith, Federated Metals Div., American Smelting & Refining Co., Pittsburgh.

Subject_"Practical Problems_Take Your Pick."

ALUMINUM AND MAGNESIUM SESSION Monday, May 1, 2:00 P.M.

Presiding-A. Cristello, American Light Alloys, Inc., Little Falls, N. I.

Co-Chairman-H. E. Elliott, The Dow Chemical Co., Bay

Recorder-A. Cristello.

"Aluminum Casting Defects-Their Identification, Causes and Cures," D. L. LaVelle, American Smelting & Refining Co., Barber, N. J.

"Heat Treatment of Magnesium Alloy Castings," E. M. Gingerich, Central Metallurgical Div., Aluminum Company of America, Cleveland.

"Alloying Zirconium to Magnesium," W. P. Saunders and F. P. Strieter, The Dow Chemical Co., Midland, Mich.

ALUMINUM AND MAGNESIUM SESSION

Thursday, May 1, 4:00 P.M.

Presiding-Hiram Brown, Solar Aircraft Co., Des Moines,

Co-Chairman-W. E. Sicha, Aluminum Company of America, Cleveland.

Recorder-Hiram Brown.

"Investment Casting of Aluminum," H. Rosenthal and S. Lipson, Pitman-Dunn Laboratories, Frankford Arsenal, Philadel-

"Some Practical Applications of Permeable Metal Casting Plaster," R. F. Dalton, United States Gypsum Co., Chicago.

SAND SHOP COURSE—NON-FERROUS Thursday, May 1, 4:00 P.M.

Presiding-W. B. Scott, National Bearing Div., American Brake Shoe Co., Meadville, Pa.

Co-Chairman-J. W. Clarke, General Electric Co., Erie, Pa. Subject-"Synthetic vs. Naturally Bonded Sands."

Discussion Leaders-F. S. Brewster, Harry W. Dietert Co., Detroit; C. A. Sanders, American Colloid Co., Chicago; H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.; H. H. Fairfield, Wm. Kennedy & Sons Ltd., Owen Sound, Ont.

ALUMINUM AND MAGNESIUM SESSION

Friday, May 2, 10:00 A.M.

Presiding-W. J. Klayer, Aluminum Industries, Inc., Cincin-

Co-Chairman-R. C. Boehm, Wellman Bronze & Aluminum Co., Cleveland.

Recorder—W. J. Klayer.
"Mechanical Properties of Sand Cast Magnesium-Zinc-Zirconium Alloys," J. F. Hildebrand and F. P. Strieter, The Dow Chemical Co., Midland, Michigan.

"High Strength Non-Heat-Treated Aluminum Casting Alloys," Walter Bonsack, Christiansen Corp., Chicago, presented by Chas.

"Zinc in Aluminum Casting Alloys," D. L. Colwell, Apex Smelting Co., Chicago.

Brass and Bronze Session

Friday, May 2, 10:00 A.M. Presiding-G. P. Halliwell, H. Kramer & Co., Chicago.

Co-Chairman-H. J. Roast, Foundry Consultant, London, Ont., Canada.

Recorder-R. A. Colton, American Smelting & Refining Co., Barber, N. J.

"Effects of Gating Practice on Leak Tightness of 85-5-5-5 and 81-3-7-9 Alloy Castings," J. G. Kura, Battelle Memorial Institute, Columbus, and L. W. Eastwood, Kaiser Aluminum & Chemical Corp., Spokane, Wash.

"Influence of Aluminum on Properties of Cast Gun-Metal and Removal of Aluminum by Slag," Official Swedish Exchange Paper, A. V. Larsson, AB Svenska Metallverken, Vasteras, Sweden.

PATTERN SESSION

Friday, May 2, 10:00 A.M.

Presiding-E. T. Kindt, Kindt-Collins Co., Cleveland.

Co-Chairman-Vaughan C. Reid, City Pattern Foundry & Machine Co., Detroit.

Recorder-F. J. Oklessen, Motor Patterns Co., Cleveland. "Core Boxes from Core Plugs," J. E. Mathias, Accurate Match Plate Co., Chicago.

"Pattern Making to Compensate for Shrink Allowance," M. K. Young, United States Gypsum Co., Chicago.

SAND SESSION

Friday, May 2, 10:00 A.M.

Presiding-C. C. Sigerfoos, Michigan State College, East Lansing

Co-Chairman-V. M. Rowell, Archer-Daniels-Midland Co., Cleveland.

Recorder-F. P. Goettman, Standard Sand Co., Grand Haven, Mich.

"Olivine Application in the Foundry," G. S. Schaller, University of Washington, Seattle.

"The Mechanism of Bonding in Foundry Molding Sands," L. M. Diran and H. F. Taylor, Massachusetts Institute of Technology, Cambridge.

"Practical Aspects of Olivine as a Molding Material," Official Norwegian Exchange Paper-John Sissener and Bjorn Langum, A/S Myrens Verksted, Bentsbrought, Oslo, Norway.

ALUMINUM AND MAGNESIUM ROUND TABLE LUNCHEON Friday, May 2, 12:00 Noon

Presiding-W. E. Sicha, Aluminum Company of America, Cleveland.

Discussion Leaders: (Aluminum) T. D. Stay, Reynolds Metals Co., Cleveland; (Magnesium) C. C. Hitchcock & Sons, Inc., Minneapolis.

Subject-"Sand Casting Practice for Aluminum and Magnesium Alloys."

PATTERN ROUND TABLE LUNCHEON Friday, May 2, 12:00 Noon

Presiding-E. T. Kindt, Kindt-Collins Co., Cleveland.

Co-Chairman-A. F. Pfeiffer, Allis-Chalmers Mfg. Co., Mil-

"Pattern Apprentice Training Program," F. W. Burgdorfer, Missouri Pattern Works, St. Louis.

"New Methods in Metal Patternmaking," H. J. Jacobson, Industrial Pattern Works, Chicago.

BRASS AND BRONZE SESSION Friday, May 2, 2:00 P.M.

Presiding-C. A. Robeck, Gibson & Kirk Co., Baltimore. Co-Chairman-B. F. Shepherd, Ingersoll-Rand Co., Phillipsburg, N. I.

Recorder-F. L. Riddell, H. Kramer & Co., Chicago.

"Effects of Mold Materials on Leak Tightness and Mechanical Properties of 85-5-5-5 and 81-3-7-9 Alloy Castings," J. G. Kura, Battelle Memorial Institute, Columbus, and L. W. Eastwood,

Kaiser Aluminum & Chemical Corp., Spokane, Wash..
"Fracture Characteristics and Melt Quality of 85-5-5-5 Red Brass and Other Copper-Base Alloys," Research Progress Report -C. Upthegrove, University of Michigan, Ann Arbor, and F. B. Rote, Albion Malleable Iron Co., Albion, Mich.

"Refining and Elimination of Gases from Metals and Alloys," H. Lepp, F. Monneret and P. Rame, France.

MALLEABLE AND REFRACTORIES SESSION Friday, May 2, 2:00 P.M.

Presiding-W. G. Ferrell, Auto Specialties Manufacturing Co., St. Joseph, Michigan.

Co-Chairman-Eric Welander, John Deere Malleable Works, East Moline, Ill.

Recorder-Wm. Zeunik, National Malleable & Steel Castings Co., Indianapolis, and R. A. Witschey, A. P. Green Fire Brick Co., Chicago.

"Refractory Practice as Applied to Malleable Duplexing," C. W. Meyer, Central Foundry Div., Saginaw Malleable Iron Plant, GMC, Saginaw, Mich.

"Air Furnace Refractory Practice in Cupola-Air Furnace Duplexing," W. C. Corbeau, National Malleable & Steel Castings Co., Cleveland.

"Refractories Applied to Batch Type Air Furnaces," C. O. Schopp and P. F. Ulmer, Link-Belt Co., Indianapolis.

SAND SHOP COURSE—NON-FERROUS Friday, May 2, 4:00 P.M.

Presiding-R. H. Jacoby, The Key Company, East St. Louis,

Co-Chairman-B. N. Ames, New York Naval Shipyard, Brook-

Recorder-G. M. Etherington. American Brake Shoe Co., Mahwah, N. J. Subject—"Symposium on Shell Molding."

Discussion Leaders-Vaughan C. Reid, City Pattern Foundry & Machine Co., Detroit; R. W. Crannell, Lehigh Foundries, Inc., Easton, Pa.; N. A. Kahn, New York Naval Shipyard, Brooklyn; E. N. Harrison, Econocast, Inc., Chattanooga, Tenn.; H. K. Salzberg, The Borden Co., Bainbridge, N. Y.

ANNUAL BUSINESS MEETING

Saturday, May 3, 9:30 A.M. Presiding-Walter L. Seelbach, A.F.S. President.

President Seelbach called the meeting to order as the Annual Business meeting of the American Foundrymen's Society. President Seelbach then introduced and called on various representatives of foundry associations and societies from abroad. He then presented the President's Annual Address. See page xiii.

Following this presentation President Seelbach called on Secretary-Treasurer Wm. W. Maloney who announced the 1952 A.F.S. Apprentice Contest winners as follows.

Wood Patternmaking Division

1st-David Lee Schroeder, Universal Foundry Co., Oshkosh, Wis.

2nd-Fred R. Fiorentini, Modern Pattern Co., Cleveland 3rd—Donald A. Siebert, Royal Pattern Works, Cleveland Metal Patternmaking Division

1st-Luther R. Hardin, Ford Motor Co., Dearborn, Mich. 2nd-Donald E. Wagley, City Pattern & Foundry Co., South Bend. Ind.

3rd-Burdette C. Henricks, Pontiac Motor Div. of GMC, Pontiac, Mich.

Gray Iron Molding Division

1st-Robert P. Schmidt, Brown & Sharpe Mfg. Co., Providence, R. I.

2nd-Jerome J. Rodman, Whitin Machine Works, Whitins-

3rd-Ronald J. Kurber, International Harvester Co., Chicago Steel Molding Division

1st-Henry C. Grube, Waukesha Foundry Co., Waukesha, Wis. 2nd-Adolph C. Zinn, Waukesha Foundry Co., Waukesha, Wis. 3rd-Roland Lalumiere, Canadian Car & Foundry Co., Ltd., Montreal, P. Q., Canada.

Non-Ferrous Molding Division

1st-Fernand Lemieux, Dominion Engineering Works, Ltd., Lachine, P. Q., Canada.

2nd-Jack Monaco, Dominion Engineering Works, Ltd., Lachine, P. Q., Canada.

3rd-Gaston Vigneault, Miller's Brass Foundry, Trois Rivieres, P. O., Canada.

The Society arranged to have the five first-prize winners present at the Convention and to receive their awards in person. The first-prize winners were called to the platform and President Seelbach presented each with the first prize, a check for \$100.00 and an engraved Certificate of Award prefaced by a few words of commendation and encouragement.

President Seelbach then called on Secretary-Treasurer Maloney who reported on the nominations of Officers and Directors for the coming year and stated that no additional nominees had been received in accordance with the procedure prescribed in Art. X of the Society By-Laws. He therefore cast the unanimous ballot of the membership of A.F.S. for the election of the fol-

President (to serve one year):

I. R. Wagner, Electric Steel Casting Co., Indianapolis, Ind.

Vice-President (to serve one year):

Carter L. Collins, Albion Malleable Iron Co., Albion, Mich.

Directors (to serve three years):

Martin A. Fladoes, Sivyer Steel Casting Co., Milwaukee. Walter J. Klayer, Aluminum Industries, Inc., Cincinnati.

J. O. Klein, Texas Foundries, Inc., Lufkin, Texas.

A. D. Matheson, French & Hecht Division, Kelsey-Hayes Wheel Co., Davenport, Iowa.

H. G. Robertson, American Steel Foundries, Alliance, Ohio.

Director (to serve one year):

Walter L. Seelbach, Superior Foundry, Inc., Cleveland.

President Seelbach next introduced John S. Bugas, Ford Motor Co., who delivered the Charles Edgar Hoyt Lecture.

CHARLES EDGAR HOYT ANNUAL LECTURE Saturday, May 3, 11:00 A.M.

"Industry's Responsibility to Youth," John S. Bugas, Vice-President, Industrial Relations, Ford Motor Co., Dearborn, Michigan.

PATTERN SESSION Saturday, May 3, 2:00 P.M.

Presiding-V. J . Sedlon, Master Pattern Co., Cleveland. Co-Chairman-H. C. Swanson, Arrow Pattern and Foundry, Chicago.

Recorder.—F. J. Oklessen, Motor Patterns Co., Cleveland.
"Practical Construction of Wood Patterns," Walter Siebert, Elyria Foundry Div., Industrial Brown Hoist Co., Elyria, Ohio. "Patterns for Malleable and Steel Foundries," J. M. Kreiner, National Malleable & Steel Castings Co., Cleveland.

SAND SESSION Saturday, May 3, 2:00 P.M.

Presiding-H. J. Williams, New Jersey Silica Sand Co., Millville, N. J.

Co-Chairman-Stanton Walker, National Industrial Sand Assn., Washington, D. C.

Recorder-D. C. Williams, Ohio State University, Columbus. "Mineral Perlite and Its Uses in the Foundry," E. D. Boyle and H. R. Wolfer, Puget Sound Naval Shipyard, Bremerton,

"Influence of Specimen Tube Loading and Riddle Size on Molding Sand Test Results," D. I. Huizenga, Albion Malleable Iron Co., Albion, Mich., and K. E. Spray, U. S. Bureau of Ships, Washington, D. C.

"A Study of the A.F.S. Fineness Test," Official Exchange Paper, Institute of Australian Foundrymen (Victorian Division)-H. A. Stephens, Commonwealth Scientific and Industrial Research Organization, Melbourne, Australia,

ALUMINUM AND MAGNESIUM SESSION Saturday, May 3, 3:00 P.M.

Presiding-R. F. Thomson, Research Laboratories Div., General Motors Corp., Detroit.

Co-Chairman—T. D. Stay, Reynolds Metals Co., Cleveland.

A.F.S. Aluminum and Magnesium Research Progress Report-Motion Picture, "Effect of Gating Design on Casting Quality." K. Grube and J. H. Jackson, Battelle Memorial Institute, Colum-

BRASS AND BRONZE SESSION Saturday, May 3, 3:00 P.M.

Presiding-W. A. Mader, Oberdorfer Foundries, Inc., Syracuse, N. Y.

Co-Chairman-B. W. Schafer, Kuhlman Electric Co., Bay

Recorder-C. P. Kotowicz, Ampco Metal, Inc., Milwaukee.

"Effect of Lead and Nickel on Grain Size of Certain Cast Copper-Base Alloys," R. A. Colton, F. L. Turk and D. L. Lavelle, American Smelting & Refining Co., Barber, N. J.

"Use of Aluminum Bronze in High Pressure Castings," T. C. Bunch and G. E. Dalbey, Mare Island Naval Shipyard, Vallejo,

"Melting in High Speed Reverbatory Furnace," R. G. Carlson and W. B. Scott, National Bearing Div., American Brake Shoe Co., Meadville, Pa.

MALLEABLE SESSION Saturday, May 3, 4:00 P.M.

Presiding-W. K. Bock, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman-F. B. Rote, Albion Malleable Iron Co., Albion, Mich.

Recorders-W. K. Bock and F. B. Rote.

"A Test for Hot Tearing Tendency," A.F.S. Research Progress Report-A. E. Lange and R. W. Heine, University of Wisconsin,

"Effects of Melting Furnace Atmosphere on Fluidity, Hot Tearing Tendency, and Other Properties of Malleable Iron," A.F.S. Research Progress Report. A. E. Lange and R. W. Heine, University of Wisconsin, Madison.

SAFETY, HYGIENE AND AIR POLLUTION SESSION Saturday, May 3, 4:00 P.M.

Presiding-A. L. Hunt, National Bearing Div., American Brake Shoe Co., St. Louis.

Co-Chairman-C. K. Faunt, Christensen & Olsen Foundry Co., Chicago.

Recorder-H. F. Scobie, American Foundrymen's Society, Chi-

"Ventilation at Non-Ferrous Melting and Pouring Opera tions," H. J. Weber, American Brake Shoe Co., Chicago.

"How to Maintain Foundry Ventilation and Dust Collecting Systems," K. M. Smith, Caterpillar Tractor Co., Peoria, Ill.

SAND SESSION

Saturday, May 3, 4:00 P.M.

Presiding-O. J. Myers, Archer-Daniels-Midland Co., Minneapolis.

Recorder-F. S. Brewster, Harry W. Dietert Co., Detroit. "Influence of Oven Atmosphere on Drying of Molds and Baking of Cores," Georges Ulmer and Maurice DeCrop, Centre Technique des Industries de la Fonderie, Paris, France.

Flowability of Foundry Sands," L. M. Diran, A. J. Shaler and H. F. Taylor, Massachusetts Institute of Technology, Cambridge.

INTERNATIONAL BANQUET Saturday, May 3, 7:30 P.M.

Presiding-Walter L. Seelbach, President, American Foundrymen's Society.

The Annual A.F.S. Dinner of the Society's 56th Annual Meeting and International Congress was called to order by President Walter L. Seelbach, presiding.

President Seelbach requested the audience to stand in silence for a moment in memory of the late Dr. Guido Vanzetti, Milan, Italy, president of the International Congress of Foundry Technical Associations during its formative years.

President Seelbach then called on A.F.S. Past President Ralph Teetor, Chairman of the A.F.S. Board of Awards who presided over presentation of A.F.S. Gold Medal Awards and Honorary Life Memberships to eight foundrymen whose contributions to the art of metal casting and to the Society have been outstanding:

THE WM. H. McFadden Gold Medal

Awarded to: Henton Morrogh, British Cast Iron Research Association, Alvechurch, Birmingham, England, "For outstanding work and development in the field of spheroidal cast iron."

THE JOS. S. SEAMAN GOLD MEDAL

Awarded to: Albert Portevin, Institut de France, Paris, France "For technical contributions over many years to the arts and sciences of the foundry industry."

THE JOHN A. PENTON GOLD MEDAL

Awarded to: Frank G. Steinebach, Penton Publishing Co., Cleveland, "For outstanding contributions to the foundry industry and the Society."

THE PETER L. SIMPSON GOLD MEDAL

Awarded to: Albert P. Gagnebin and Keith D. Millis, The International Nickel Co., New York "For outstanding work and development in the field of spheroidal cast iron."

HONORARY LIFE MEMBERSHIPS IN A.F.S.

Awarded to: William G. Mixer, Consultant of Flint, Michigan "For a lifetime of service and application of engineering to the

Alexander I. Krynitsky, National Bureau of Standards, Washington, D. C. "For a lifetime of fundamental research in the field of metal casting."

Walter L. Seelbach, Superior Foundry, Inc., Cleveland "In recognition of his services to the American Foundrymen's Society as its President, 1951-52."

INTERNATIONAL AWARD OF HONOR

International Vice-President L. N. Shannon presented the International Award of Honor—a bronze replica of Cellini's Perseus provided by Mario Olivo of Milan, Italy, to Dr. Eugenio Mortara in the name of the late Carlo Vanzetti of Milan, past president of the International Committee and one of its original organizers. Dr. Vanzetti's son, Guido, who was to have accepted the award for his late father, died in March, 1952. Dr. Mortara

in turn presented the award to A.F.S. President Seelbach, entrusting its care to A.F.S. as custodian for the Vanzetti family.

GRAY IRON SESSION

Sunday, May 4, 2:00 P.M.

Presiding-J. S. Vanick, International Nickel Co., New York. Recorder-Chas. Mooney, Olney Foundry Div., Link-Belt Co., Philadelphia.

"A New Process for the Industrial Production of Nodular Iron With Low Treatment Costs," Official Exchange Paper, Italian Metallurgical Association-Carlo Longaretti and Mario Noris, Institute for Siderurgy, Finsider, Genoa, Italy.

"Production of Heavy Spheroidal Graphite Cast Iron Gears," Corrado Galletto, Centro d'Informazioni del Nickel, Milano,

"Gases and Naturally Occurring Blowholes in Foundry Practice"-Official Exchange Paper, Association Technique de Fonderie de France-A. M. Portevin, Institut de France, Paris.

CANADIAN DINNER

Sunday, May 4, 7:00 P.M.

Chairman-J. J. McFadyen, Galt, Ont.

GRAY IRON SESSION Monday, May 5, 10:00 A.M.

Presiding-J. D. Sheley, The Black Clawson Co., Hamilton, Ohio.

Co-Chairman-R. A. Clark, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

Recorder-W. T. Bourke, American Brake Shoe Co., Mahwah, N. J. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Steel Melt in a Cupola," H. W. Lownie, Jr. "How Iron and Iron an

and C. T. Greenidge, Battelle Memorial Institute, Columbus, and D. E. Krause, Gray Iron Research Institute, Columbus. "Melting Iron in a Basic-Lined, Water-Cooled Cupola," W. W.

Levi, Lynchburg Foundry Co., Radford, Va.

"Metallurgical Blast Cupola," Official Belgian Exchange Paper -Robert Doat, Compagnie General des Conduites d'Eau, Liege, Belgium, and M. A. DeBock, Consulting Engineer, Brussels,

STEEL SESSION

Monday, May 5, 10:00 A.M.

Presiding-C. B. Jenni, General Steel Castings Corp., Eddy-

Co-Chairman-Dale Hall, Oklahoma Steel Casting Co., Tulsa,

Recorder-L. H. Hahn, Sivyer Steel Castings Co., Chicago. "The Substitution of Boron for Alloys in Cast Steel," R. A. Dyke, Jr., and C. K. Donoho, American Cast Iron Pipe Co.,

Birmingham, Ala. "Application of Chills to Increasing the Feeding Range of Risers," E. T. Myskowski, H. F. Bishop and W. S. Pellini, Naval

Research Laboratory, Washington, D. C.

"Development of Steel Foundries in India," Official Exchange Paper, Institute of Indian Foundrymen-N. G. Chakrabarti, Calcutta.

SAFETY, HYGIENE AND AIR POLLUTION SESSION Monday, May 5, 10:00 A.M.

Presiding-F. W. Shipley, Caterpillar Tractor Co., Peoria, Ill. Co-Chairman-J. R. Allan, International Harvester Co., Chi-

Recorder-H. F. Scobie, American Foundrymen's Society, Chi-

cago.
"The A.F.S. Safety, Hygiene and Air Pollution Program," J. R. Allan, Chairman, A.F.S. Safety, Hygiene and Air Pollution Committee-International Harvester Co., Chicago.

"The Foundryman Looks at Air Pollution," N. H. Keyser and H. P. Munger, Battelle Memorial Institute, Columbus

"Health Problems of the Metal Castings Industry," D. A. Irwin, M.D., Aluminum Company of America, Pittsburgh.

> TIMESTUDY AND METHODS SESSION Monday, May 5, 10:00 A.M.

Presiding-L. W. Lehman, John Deere Van Brunt Co., Hori-

Co-Chairman-H. R. Williams, Williams Management Engineering, Milwaukee.

Recorder-H. R. Williams.

"Use of Motion Pictures for Foundry Motion and Timestudy," W. K. Richardson, Purdue University, Lafayette, Ind.

"Application of Motion Pictures for Motion and Timestudy in the Foundry," M. T. Sell, Sterling Foundry Co., Wellington,

"Applied Waste Control Principles," J. R. Walley, Helmick and Associates, Canton, Ohio.

MALLEABLE SESSION

Monday, May 5, 10:00 A.M.

Presiding-W. M. Albrecht, Chain Belt Co., Milwaukee. Co-Chairman-P. F. Ulmer, Link-Belt Co., Indianapolis.

Recorder-L. G. Osborne, Lakeside Malleable Castings Co.,

Racine, Wis.

"Effect of Phosphorus Content on Graphitization Rate and Mechanical Properties of Black Heart Malleable Iron," J. E. Rehder, Department of Mines and Technical Surveys, Ottawa, Ont., Canada.

"Malleable Iron Annealing Time Reduced," W. G. Wilson and N. F. Tisdale, Jr., Molybdenum Corp. of America, Pittsburgh.

SAND SESSION

Monday, May 5, 10:00 A.M.

Presiding-J. B. Caine, Consultant, Wyoming, Ohio. Co-Chairman-H. W. Dietert, Harry W. Dietert Co., Detroit.

Recorder—R. G. Thorpe, Cornell University, Ithaca, N. Y. "Pneumatic Reclamation for Foundry Sands," C. E. Wennin-

ger, National Engineering Co., Chicago.

"Reclaiming Used Molding Sands by Air Scrubbing," H. H. Fairfield, J. McConachie, and H. F. Graham, Wm. Kennedy & Sons, Ltd., Owen Sound, Ont., Canada.

"Sand Control with Particular Reference to the Prevention of Scabbing." Official Exchange Paper, Institute of British Foundrymen—W. B. Parkes, British Cast Iron Research Assn., Birmingham, England.

> MALLEABLE ROUND TABLE LUNCHEON Monday, May 5, 12:00 Noon

Presiding-J. H. Lansing, Malleable Founders' Society, Cleveland.

Co-Chairman-W. A. Kennedy, The Grinnell Co., Inc., Providence, R. I.

Subject-"Dielectric Core Baking."

Discussion Leaders-K. H. Hamblin, The Grinnell Co., Inc., Providence, R. I.; P. C. DeBruyne, Moline Malleable Iron Co., St. Charles, Illinois.

Subject-"Fluidity Tests."

Discussion Leader-P. C. Rosenthal, University of Wisconsin, Madison.

> CHAPTER EDUCATION ACTIVITIES SESSION Monday, May 5, 2:00 P.M.

Presiding-F. G. Sefing, International Nickel Co., Inc., New

Co-Chairman-E. W. Horlebein, Gibson & Kirk Co., Balti-

Recorder-W. J. Hebard, Continental Foundry & Machine Co., '

East Chicago, Ind.
Speakers—Northwestern Pennsylvania Chapter—E. M. Strick, Erie Malleable Iron Co., Erie, Pa.; Southern California Chapter Robert Gregg, Reliance Regulator Div., American Meter Co., Alhambra, Calif.; Philadelphia Chapter-A. N. Kraft, Wilkening Mfg. Co., Philadelphia, and H. E. Mandel, Pennsylvania Foundry Supply and Sand Co., Philadelphia.

> PLANT AND PLANT EQUIPMENT SESSION Monday, May 5, 2:00 P.M.

Presiding-James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman-R. J. Wolf, Stone & Webster Engineering Corp.,

Recorders-James Thomson and R. J. Wolf.

"Development and Application of Molding Machines for the Production of Light Castings," A. S. Hedberg, Wells Mfg. Co.,

"Development and Application of Machines for the Making of Molds by Slinging," Martin Putz, Mattison Machine Works, Rockford, Ill.

SAFETY, HYGIENE AND AIR POLLUTION SESSION Monday, May 5, 2:00 P.M.

Presiding—J. M. Kane, American Air Filter Co., Louisville. Recorder—H. F. Scobie, American Foundrymen's Society. "The Single Objective Approach to Foundry Safety," Dan Farrell, United States Steel Co., Pittsburgh.
"Air Pollution and the Cupola," J. C. Radcliffe, Ford Motor

Co., Dearborn, Mich.

GRAY IRON SESSION Monday, May 5, 2:00 P.M.

Presiding—J. T. MacKenzie, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman-T. J. Wood, American Brake Shoe Co., Mahwah, N. J.

Recorder—H. W. Lownie, Jr., Battelle Memorial Institute, Columbus, Ohio.

"Production Experiences with a Basic Cupola," S. F. Carter, American Cast Iron Pipe Co., Birmingham, Ala.

"Factors Affecting Fluidity of Cast Iron," L. F. Porter and P. C. Rosenthal, University of Wisconsin, Madison.

"Influence of Phosphorus on Hot Tear Resistance of Plain and Alloy Gray Iron," J. C. Hamaker, Jr., General Iron Works Co., Denver, and W. P. Wood, University of Michigan, Ann Arbor.

STEEL SESSION

Monday, May 5, 2:00 P.M.

Presiding-R. H. Jacoby, The Key Company, East St. Louis, Ill.

Co-Chairman—R. H. Frank, Bonney-Floyd Co., Columbus, Ohio.

Recorder-J. R. Goldsmith, Crane Co., Chicago.

"Metallurgy and Mechanics of Hot Tearing," H. F. Bishop, C. G. Ackerlind and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

"Hot Tear Formation in Steel Castings," U. K. Bhattacharya, C. M. Adams and H. F. Taylor, Massachusetts Institute of Tech-

nology, Cambridge.

"Investigation of Hot Tears in Steel Castings," Research Progress Report (in three parts): Part I, Gordon Johnson, Armour Research Foundation, Chicago; Part II, G. A. Lillieqvist, American Steel Foundries, East Chicago, Ind.; and Part III, Clyde Wyman, Committee Chairman, Burnside Steel Foundry Co., Chicago.

GRAY IRON SHOP COURSE Monday, May 5, 4:00 P.M.

Presiding—K. H. Priestley, Vassar Electroloy Products, Inc., Vassar, Mich.

Co-Chairman—E. J. Burke, Hanna Furnace Corporation, Buffalo.

Subject_"Scrap Control by Sampling."

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Discussion Leader—R. W. Gardner, Dearborn Iron Foundry, Ford Motor Co., Dearborn, Mich.

SAND SHOP COURSE Monday, May 5, 4:00 P.M.

Presiding—H. W. Meyer, General Steel Castings Corp., Granite City. III.

Co-Chairman—F. S. Brewster, Harry W. Dietert Co., Detroit. Subject—"Use of High Density Molding Materials in the Foundry."

Discussion Leader-W. M. Peterson, M. A. Bell Co., St. Louis.

CHAPTER EDUCATIONAL ACTIVITIES SESSION Monday, May 5,4:00 P.M.

Presiding—F. G. Sefing, International Nickel Co., New York. Co-Chairman—E. W. Horlebein, Gibson & Kirk Co., Baltimore.

Recorder-W. J. Hebard, Continental Foundry & Machine Co.,

East Chicago, Ind.

Speakers—Northwestern Pennsylvania Chapter—E. M. Strick, Erie Malleable Iron Co., Erie, Pa.; Southern California Chapter—Robert Gregg, Reliance Regulator Div., American Meter Co., Alhambra, Calif.; Philadelphia Chapter—A. N. Kraft, Wilkering Mfg. Co., Philadelphia, and H. E. Mandel, Pennsylvania Foundry Supply and Sand Co., Philadelphia.

FOUNDRY COST SESSION Monday, May 5, 4:00 P.M.

Presiding—R. L. Lee, Grede Foundries, Inc., Milwaukee. Co-Chairman—G. E. Tisdaie, Zenith Foundry Co., Milwaukee. Recorder—G. E. Tisdale.

Subject-Question and Answer Panel.

Discussion Leaders—C. S. Roberts, Dodge Steel Co., Philadelphia; C. E. Westover, Westo.er Engineers, Milwaukee; W. A. Gluntz, Gluntz Brass & Aluminum Foundry Co., Cleveland.

PLANT AND PLANT EQUIPMENT SESSION Monday, May 5, 4:00 P.M.

Presiding—James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.

Co-Chairman—H. W. Johnson, Wells Mfg. Co., Skokie, Ill. Recorders—James Thomson and H. W. Johnson.

"Jolt Rollover Molding Machines for Medium and Heavy Castings," K. M. Smith, Caterpillar Tractor Co., Peoria, Ill. "Special and Automatic Molding Machines," K. M. Smith,

Caterpillar Tractor Co., Peoria, Ill.

Motion Picture, "Mechanization in Molding, II," H. C. Weimer, Beardsley & Piper Div., Pettibone Mulliken Corp., Chicago.

GRAY IRON SESSION Monday, May 5, 4:00 P.M.

Presiding—W. H. White, Jackson Iron & Steel Works, Jackson,

Co-Chairman—Carl Harmon, Hanna Furnace Corp., Buffalo. Recorder—C. 'C. Sigerfoos, Michigan State College, East Lansing, Mich.

"Commercial Experience with Higher Silicon Nodular Irons," R. Schneidewind, University of Michigan, Ann Arbor, and H. H. Wilder, Vanadium Corporation of America, Detroit.

"Effects of Cerium on Graphite Formation in Alloy Cast Iron," F. A. Rowe, University of Washington, Seattle, and H. A. Johnson, General Electric Co., Richland, Wash.

A.F.S. ALUMNI DINNER Monday, May 5,7:15 P.M.

Presiding...W. L. Woody, National Malleable & Steel Castings Co., Cleveland.

Speaker—Wm. J. Grede, President, National Association of Manufacturers—President, Grede Foundries, Inc., Milwaukee. Subject—"Our Free Enterprise."

FOUNDRY COST SESSION Tuesday, May 6, 10:00 A.M.

Presiding—R. L. Lee, Grede Foundries, Inc., Milwaukee. Co-Chairman—M. E. Annich, American Brake Shoe Co., Mahwah, N. J.

Recorder_M. E. Annich.

"Pricing Castings Using Standard Costs," J. A. Westover, Westover Engineers, Milwaukee.

EDUCATIONAL SESSION Tuesday, May 6, 10:00 A.M.

Presiding—G. J. Barker, University of Wisconsin, Madison. Co-Chairman—W. J. Hebard, Continental Foundry & Machine Co., East Chicago, Ind.

Recorder_W. J. Hebard.

"Management's Responsibility in the Training Program,"
J. D. Judge, Hamilton Foundry & Machine Co., Hamilton, Ohio.
"Apprentice Training—It Is Needed," M. C. Sandes, Mare
Island Naval Shipyard, Vallejo, Calif.

GRAY IRON SESSION Tuesday, May 6, 10:00 A.M.

Presiding—F. T. McGuire, Deere & Co., Moline, Ill. Co-Chairman—W. A. Hambley, Chas. A. Krause Milling Co., Birmingham, Mich.

Recorder—C. T. Marek, Purdue University, Lafayette, Ind. "Effect of Sulphur on the Fluidity of Gray Cast Iron," L. F. Porter and P. C. Rosenthal, University of Wisconsin, Madison. "Internal Porosity in Gray Iron Castings," J. C. Hamaker, Jr., General Iron Works Co., Denver; W. P. Wood, University of Michigan, Ann Arbor, and F. B. Rote, Albion Malleable Iron Co., Albion, Mich.

"Effect of Hydrogen on Graphitization," B. F. Brown, North Carolina State College, Raleigh, and M. F. Hawkes, Carnegie Institute of Technology, Pittsburgh.

HEAT TRANSFER SESSION

Tuesday, May 6, 10:00 A.M.

Presiding-H. A. Schwartz, National Malleable and Steel Castings Co., Cleveland.

Co-Chairman-E. C. Troy, Foundry Engineer, Palmyra, N. J. Recorder-C. E. Sims, Battelle Memorial Institute, Columbus. "Heat Flow in Moist Sand," Research Progress Report-V. Paschkis, Columbia University, New York.

"Solidification of Nodular Iron in Sand Molds," R. P. Dunphy and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

STEEL SESSION—STATISTICAL QUALITY CONTROL Tuesday, May 6, 10:00 A.M.

Presiding-H. H. Johnson, National Malleable & Steel Castings Co., Sharon, Pa.

Co-Chairman-R. W. Gardner, Dearborn Iron Foundry, Ford Motor Co., Dearborn, Mich.

Recorder-H. H. Johnson.

"The Whys and Wherefores of Statistical Quality Control," W. R. Weaver (President, American Society for Quality Control), Republic Steel Corp., Cleveland.

"Foundry Management Looks at Statistical Quality Control," E. L. Fay, Deere & Co., Moline, Ill.

GRAY IRON ROUND TABLE LUNCHEON

Tuesday, May 6, 12:00 Noon

Presiding-R. J. Allen, Worthington Pump & Machinery Corp., Harrison, N. J.

Co-Chairman-C. O. Burgess, Gray Iron Founders' Society,

Subject-"Reclamation of Gray Iron Castings."

"British Report," A. B. Everest and F. A. Ball, Mond Nickel Co., Ltd., London.

EQUIPMENT AND SUPPLIES LUNCHEON

Tuesday, May 6, 12:00 Noon

Presiding-C. V. Nass, President, Foundry Equipment Manufacturers Assn.; L. H. Heyl, President, Foundry Facings Manufacturers Assn.

Speaker-Francis E. Fisher, Chief, Foundry Equipment and Supplies Sec., Metal Working Equipment Div., NPA, Washington, D. C.

REFRACTORIES SESSION

Tuesday, May 6, 2:00 P.M.

Presiding-R. H. Stone, Vesuvius Crucible Co., Swissvale,

Co-Chairman-W. R. Jaeschke, The Whiting Corp., Harvey, 111.

Recorders-R. H. Stone and W. R. Jaeschke.

"Fundamentals of Foundry Refractories," J. S. McDowell and J. D. Custer, Harbison-Walker Refractories Co., Pittsburgh. "Effect of Slag on Furnace Linings," L. B. Wyckoff, Lewiston,

STEEL SESSION—STATISTICAL QUALITY CONTROL

Tuesday, May 6, 2:00 P.M.

Presiding-A. A. Evans, International Harvester Co., Indian-

Co-Chairman-Geo. Ver Beke, John Deere Malleable Works, East Moline, Ill.

Recorder-H. H. Johnson, National Malleable & Steel Castings Co., Sharon, Pa.

"Quality Control in a Malleable Iron Foundry," E. F. Price and O. K. Hunsaker, Dayton Malleable Iron Co., Ironton, Ohio. "Quality Control in a Jobbing Steel Foundry," H. H. Fairfield, Wm. Kennedy & Sons, Ltd., Owen Sound, Ont., Canada.

TIMESTUDY AND METHODS SESSION

Tuesday, May 6, 2:00 P.M.

Presiding-J. E. Hyland, John Deere Spreader Works, East Moline, Ill.

Co-Chairman-C. J. Pruett, McWane Cast Iron Pipe Co., Birmingham, Ala.

Recorder_C. J. Pruett.

"Standards for Rough Chipping and Removing Welds," Dean

Van Order, Burnside Steel Foundry Co., Chicago.
"A Melt Department Incentive Plan," E. G. Tetzlaff, Pelton Steel Casting Co., Milwaukee.

HEAT TRANSFER SESSION

Tuesday, May 6, 2:00 P.M.

Presiding-H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman-E. C. Troy, Foundry Engineer, Palmyra, N. J.

Recorder_J. B. Caine, Consultant, Cincinnati.

"A Simplified Analysis of Riser Treatments," C. M. Adams, Jr., and H. F. Taylor, Massachusetts Institute of Technology, Cam-

"Space, Time and Temperature Relations During Casting of Metals," Official Dutch Exchange Paper-J. S. Abcouwer, Werkspoor N. V., Utrecht, Holland.

GRAY IRON SESSION

Tuesday, May 6,4:00 P.M.

Presiding-J. E. Rehder, Department of Mines & Technical Surveys, Ottawa, Ont., Canada.

Co-Chairman-V. A. Crosby, Climax Molybdenum Co., De-

Recorder-C. F. Walton, Case Institute of Technology, Cleveland.

"Effect of Percentage of Nodular Graphite on Certain Mechanical Properties of Magnesium-Treated Cast Iron," R. W. Lindsay, Pennsylvania State College, State College, Pa., and Alvin Shames, North American Aviation, Inc., Columbus (formerly with Battelle Memorial Institute, Columbus).

"Time of Formation of Spherulites in Hypo- and Hyper-Eutectic Irons," J. Keverian, C. M. Adams and H. F. Taylor, Massachusetts Institute of Technology, Cambridge.

"A Study of the Formation of Nodular Graphite," Fredrik Hurum, Norway Institute of Technology, Trondheim.

GRAY IRON SHOP COURSE

Tuesday, May 6, 4:00 P.M.

Presiding-W. W. Levi, Lynchburg Foundry Co., Radford, Va. Co-Chairman-E. J. Burke, Hanna Furnace Corp., Buffalo. Subject-"Mechanics of Tuyeres."

Discussion Leaders-H. H. Wilder, Vanadium Corp. of America, Detroit; Carl Harmon, Hanna Furnace Corp., Buffalo.

SAND SHOP COURSE

Tuesday, May 6, 4:00 P.M.

Presiding-B. H. Booth, Carpenter Bros., Inc., Milwaukee. Co-Chairman-R. H. Olmsted, Whitehead Bros. Co., New York.

Subject-"Household Hints and Tips."

Panel Members-T. E. Barlow, Eastern Clay Products Div., International Chemicals and Minerals Co., Jackson, Ohio; T. W. Curry, Lynchburg Foundry Co., Lynchburg, Va.; R. L. Mc-Ilwaine, National Engineering Co., Chicago; O. J. Myers, Archer-Daniels-Midland Co., Minneapolis; C. V. Nass, Beardsley & Piper Div., Pettibone-Mulliken Corp., Chicago; J. A. Rassenfoss, American Steel Foundries, East Chicago, Ind.; J. S. Schumacher, Hill & Griffith Co., Cincinnati.

REFRACTORIES AND GRAY IRON SESSION

Tuesday, May 6,4:00 P.M.

Presiding-R. A. Witschey, A. P. Green Fire Brick Co., Chi-

Co-Chairman_S. F. Carter, American Cast Iron Pipe Co., Birmingham, Ala.

Recorders—R. A. Witschey and S. F. Carter.
"Basic Refractories for Cupola Service," M. W. Demler, Harbison-Walker Refractories Co., Pittsburgh.
"Refractories for the Basic Cupola," H. M. Kraner, Bethlehem

Steel Co., Bethlehem, Pa.

INTERNATIONAL EDUCATION DINNER

Tuesday, May 6,7:00 P.M.

Presiding-I. R. Wagner, Vice-President A.F.S.

Co-Chairman-C. L. Carter, Albion Malleable Iron Co., Albion, Mich.

Speakers-Noel P. Newman, Past President Institute of British Foundrymen, Newman Hender & Co., Gloucester, England; Tom Makemson, Secretary Institute of British Foundrymen, Deansgate, Manchester, England.

GRAY IRON SESSION Wednesday, May 7, 10:00 A.M.

Presiding-A. P. Gagnebin, International Nickel Co., New York.

Co-Chairman-R. Schneidewind, University of Michigan, Ann Arbor.

Recorder-G. A. Timmons, Climax Molybdenum Co., Detroit. "Physical Properties of Spheroidal Graphite Cast Iron," Technical Report No. 1 for Watertown Arsenal-C. C. Reynolds and H. F. Taylor, Massachusetts Institute of Technology, Cambridge.

"A Solidification Dilatometer and Its Application to Gray Iron," R. P. Dunphy and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

"Influence of Some Residual Elements, and Their Neutralization, in Magnesium-Treated Nodular Iron," H. Morrogh, British Cast Iron Research Association, Birmingham, England.

SAND SESSION

Wednesday, May 7, 10:00 A.M.

Presiding-H. W. Dietert, Harry W. Dietert Co., Detroit. Co-Chairman-E. C. Zirzow, Werner G. Smith, Inc., Cleveland.

Recorder-R. H. Jacoby, The Key Co., East St. Louis, Ill. "Veining Tendencies of Cores," Progress Report of Sand Committee 8-J-V. M. Rowell, Chairman, Archer-Daniels-Midland Co., Cleveland.

"Steel Sands at Elevated Temperatures," Progress Report of Research Project-R. G. Thorpe, A. E. Riccardo, P. L. Widener and P. E. Kyle, Cornell University, Ithaca, N. Y.

"Effect of Binders and Additives on Ductility in Molding Sands at Elevated Temperatures," R. F. Morey, C. G. Ackerlind and W. S. Pellini, Naval Research Laboratory, Washington, D. C.

STEEL SESSION

Wednesday, May 7, 10:00 A.M.

Presiding-C. H. Lorig, Battelle Memorial Institute, Colum-

Co-Chairman-Charles Locke, Atlas Foundry & Machine Co., Tacoma, Wash.

Recorder-E. C. Troy, Consultant, Palmyra, N. J.

"Steel Desulphurization with Injected Carbide," S. F. Carter,

American Cast Iron Pipe Co., Birmingham, Ala.
"Cutting with Ultra-High Purity Oxygen," E. H. Roper and J. F. Kiernan, Air Reduction Sales Co., New York.

"The Manufacture of Cast Edge Tools in Rotary Furnace Steel," Official Exchange Paper, Institute of Australian Foundrymen (New South Wales Div.), Sydney-Gordon Keech, Keech Castings Pty., Ltd.; A. T. Batty and W. M. Dummett, Institute of Australian Foundrymen, Sydney, Australia.

STEEL ROUND TABLE LUNCHEON

Wednesday, May 7, 12:00 Noon

Presiding-J. B. Caine, Consultant, Wyoming, Ohio. Co-Chairman-V. E. Zang, Unitcast Corporation, Toledo, Ohio.

Subject-"What Are Your Problems."

GRAY IRON SESSION

Wednesday, May 7, 2:00 P.M.

Presiding-J. S. Vanick, International Nickel Co., New York. Co-Chairman-T. E. Eagan, Cooper-Bessemer Corp., Grove City, Pa.

Recorder_J. H. Schaum, Bureau of Standards, Washington,

"Gating and Risering of Gray Iron," Research Progress Report -W. A. Schmidt and H. F. Taylor, Massachusetts Institute of Technology, Cambridge.

"Gating and Risering Terminology," Committee Report-N. A. Birch, Chairman, National Bearing Div., American Brake Shoe Co., Meadville, Pa.

PRESIDENT'S ANNUAL ADDRESS

To PREPARE an Annual Report on the activities and progress of this Society is not a simple undertaking. If the American Foundrymen's Society were much smaller, as it once was, the President could single out for recognition a few well-known people and then talk of what we hope some day to accomplish. If it were much larger, we could confine ourselves, in limited time, to the broad aspects and impersonal problems of an entire industry.

The very nature of the American Foundrymen's Society makes such an escape impossible. Because this is a Technical Society, we must be factual, not academic. Because our activities are of greatest value to the individual, we cannot hide ourselves in an ivory tower of impersonality. Because A.F.S. is close to the hearts of its members, you have a right to know what

your Society is doing on your behalf.

Because A.F.S. is recognized primarily as a Technical Society, its work is directed primarily toward technical advancement of foundry operations. This work is carried on by eight Technical Divisions-Gray Iron. Steel, Malleable, Brass and Bronze, Light Metals, Sand, Patternmaking and Education-and by a number of General Interest Committees dealing with such subjects as Refractories, Costs, Plant and Plant Equipment, Heat Transfer and Timestudy and Methods.

Technical Committee Activities

Some 600 men serve voluntarily on the Technical Committees of the Society, and it is their willingness to impart their experience and knowledge that has gained A.F.S. its broad recognition. As to the caliber of these men, no corporation in the country is big enough or powerful enough to command the brains and time and effort they give willingly to A.F.S., for the most part without compensation of any kind. How can you discount or minimize their contributions to foundry operations? Yet many otherwise intelligent foundry operators take little or no advantage of this great wealth of information so readily available.

The Technical Divisions and Committees of A.F.S. form the cornerstone of our technical work. Their members prepare books and other publications, develop the annual Convention programs, sponsor and supervise research work, write articles for AMERICAN FOUNDRYMAN and the technical press, consult with the Society on technical-practical problems, present foundry talks before Chapters, Regional Conferences, and Conventions, and they meet many times-largely unpublicized-to advance the Society and industry in

many ways

Anyone interested is urged to serve on these Com-

mittees of the Society, if only he is willing to contribute of his knowledge and experience in order to gain from others. I can think of no way for a young man to better himself and improve his knowledge than to serve on one of these Committees. Progressive foundry managers should encourage their key men to

participate.

A Technical Society, of course, must point the way as well as to record what others have done. Hence, in 1946 the A.F.S. Board approved a plan for A.F.S.sponsored research under which each Technical Division may initiate for approval a research project involving a maximum annual expenditure of \$5000. Once approved, bids are sought for doing the work, definitive contracts are let by the Divisions, and divisional Research Committees supervise the progress.

A.F.S.-Sponsored Research Projects

At present seven Research projects are under way: "Hydraulics of Light Metal Flow into Molds" by the Light Metals Division at Battelle Memorial Insti-

"Centrifugal Casting of Light Metal Alloys," a special project of the Light Metals Division at Canadian Bureau of Mines and Technical Surveys.

"Effect of Melting Conditions on the Behavior of Malleable Iron" by the Malleable Division at University of Wisconsin.

"High Temperature Properties of Molding Sands" by the Sand Division at Cornell University.

'Relation of Cores to Hot Tearing" by the Steel Division at Armour Research Foundation.

"Risering of Gray Iron," by the Gray Iron Division

at the Massachusetts Institute of Technology. "Fundamentals of Heat Flow During Casting Solidification" by the Heat Transfer Committee at Colum-

bia University.

Much important and valuable information has come from these projects; for example, the several A.F.S. films on "Fluid Flow in Transparent Molds." Several of these films have been shown extensively in Europe, and one has been presented to the British Steelfound-

ers Association for European use.

Long a major activity of the Technical Committees is the preparation of foundry publications, now an important A.F.S. function. At present some 50 titles are available, many of them constantly being revised and brought up to date, on various phases of foundry practice. They include: recommendations for good foundry practice, symposia on specific metals, foundry costs, educational guides, and handbooks on Cast Metals, Alloy Irons, Cupola Operations, Sand Test Methods, Sand and Core Practice, Brass and Bronze, and Casting Defects. A most interesting "History of

^{*} President, Superior Foundry, Inc., Cleveland.

the Metal Castings Industry" was also published by the Society a few years ago.

Twelve major publications are now under development in an intensified program to disseminate up-to-date information. This fiscal year alone, since last July, four major publications have been made available which, incidentally, I consider a rather remarkable feat for our National Office Staff in this busy year of an International Congress and Exhibit.

These new four books include; the excellent Foundry Work, a foundry textbook at the high school level, by Edwin Doe, who now is on a special foundry educational mission in South America for our Government. Next came the newest (sixth) edition of the Foundry Sand Handbook, a complete revision and enlargement, The latest two are: "Copper-Base Alloys Foundry Practices" and a most informative Symposium on Principles of Gating.

In addition to the high school foundry text book, the Society is committed to the publication of two more for teaching purposes, one at the college level, and one at the apprentice level.

These textbooks are only one phase of the Society's educational work. Actually, all A.F.S. activities are of educational nature. The National Apprentice Contests, held annually since 1924, are familiar to many. A.F.S. is a charter sponsor of the Foundry Educational Foundation, which has been remarkably successful in promoting better foundry instruction at engineering schools and colleges. The Foundation's work, under the able direction of George K. Dreher, has been cited more than once as an example of splendid cooperation between an industry and higher education.

The Society's own educational activities are mainly at the secondary school level, and necessarily developed through its local Chapters. Many Chapters are carrying on education courses for foundry workers, and are engaged in intensive programs at local school levels, so as to encourage young men to enter foundry works.

The Chapter Regional Conferences held at engineering schools are a part of this educational work, as are also the Society's 13 student chapters. The member dues of all students in Student Chapters are refunded to the schools. Through cooperation between A.F.S. and the Foundry Educational Foundation, we have arranged for Student Delegates to attend the International Foundry Congress from a number of engineering schools.

Chapter Educational Activities Vital

Not all Chapters have seen the opportunity for carrying on local educational programs. Even if on a small scale, and on a personal-interest basis, much can be done to encourage young men to enter the foundries. We need them, and foundry management should take an active interest in this Chapter work.

Besides being book publishers, Å.F.S. is a magazine publisher on a growing scale. American Foundryman was established in 1938, and only since 1945 in its present size. In seven years it has been made an authoritative source of information on castings practices, operations and methods. Its value to the membership has been constantly increased through editorial improvement and growing editorial content.

Perhaps we may be pardoned for calling it "The FOUNDRYMEN'S OWN Magazine."

In praising ourselves, we can well afford a word of praise for *Foundry* magazine, so well known to so many foundrymen. Considering the size and importance of this industry of ours, I firmly believe there is room—and need—for both *Foundry* and American Foundryman... and I have a strong feeling that a more enlightened realization of the value of *both* periodicals is now at hand—if not past due. If we, as American businessmen, believe in the spirit of competition, then let us apply it first at home.

Today the American Foundrymen's Society has close to 10,000 members in 33 countries of the world, with 54 organized local Chapters. With such an operation, it is not possible to direct all activities from one central point—nor do we try. The major responsibility for our success lies in the hands of our Chapters, where 95 per cent of the entire membership resides.

How well are we doing our job? For the most part, very creditably. Frequent technical meetings, Regional Conferences, educational courses, exhibits of public interest, various training programs, and cooperation with local schools—all are encouraged by the Society. Each year A.F.S. brings to Chicago—at an annual expense of \$10,000—two officers from each regular Chapter for a 2-day conference and discussion of Chapter operations. These conferences have materially raised the level of Chapter operations.

Yet we still have many rough spots, and here are a few capable of correction if we but try. First, foundry management does not encourage its men to take an active part in Chapter activities and so better themselves as employees. Second, a few Chapters, usually dominated by small groups, are inclined to operate more as "clubs" than as Chapters of a major technical society. Third, in some Chapters the programs are too technical and in some too commercial to attract maximum attendance.

Interest of Top Management Necessary

The solution to these ills is both obvious and available . . . greater interest on the part of foundry top management, since management largely pays the bill. The Chapters are powerful tools for the good of your plants, but tools, like men, dull from lack of use.

The Board of Directors of your Society are, I assure you, the real directive force they should be. They have assumed their obligations, and practically every Director spends much time in visiting and inspecting Chapters and keeping in touch with Society developments. The Board members are constantly proposing new ideas or improvements, counseling with Chapters and Committees, and doing a splendid job. No President could ask for greater cooperation than that given me this past year, and I want to express my personal thanks for their loyal, active support.

We do need, however, better rules for the guidance of Board decisions, as Past-President E. W. Horlebein pointed out several years ago. Thanks to his insistence, steps have been taken toward setting up definite Board policies, and this work now should be strongly for-

The work has had two important setbacks to date: the program to establish a permanent A.F.S. Headquarters building, and the new Safety, Hygiene, and

Air Pollution project.

In May, 1950 the Building program was announced, with the object of raising \$100,000 by voluntary subscription of the industry. To date, in a remarkable demonstration of confidence and esteem, a fund of over \$142,000 has been raised. After much discussion, it has been decided to build an entirely new, modern building of about 9,060 square feet area, in one of the suburbs of Chicago. When the required land is purchased, specifications will be drawn and contractor bids obtained.

Meanwhile, we have undertaken another very important activity—the Safety, Hygiene, and Air Pollution Program—out of the need of the industry to set its own engineering standards of plant conditions, both internal and external. This project is backed by the National Castings Council, comprising all the major foundry associations in this country, and a fund of \$350,000 is being raised for a minimum 10-year program. The work is already under way, and substantial funds are now coming in through voluntary contributions.

A.F.S. Safety, Hygiene and Air Pollution Program

The announced basis of subscriptions to the Safety, Hygiene, and Air Pollution Fund was calculated to make it possible for all foundries, large or small, to have a stake in the program, yet to be burdensome to none. We believe the basis is fundamentally sound. It is a program for all foundries—and all foundries should support it, for their own interest if nothing more. The simple fact back of the activity is this: Will we do the job or do we want Government or Labor to do it for us—and tell us to get in line, or else? If you believe, as strongly as I do, that it is time to call a halt to government "benevolence" and interference, with its logical train of consequences, control and seizure. Make arrangements to support the Safety, Hygiene, and Air Pollution Program.

I must, however, in all fairness, thank the members of the National Castings Council for the splendid original impetus given this vital activity. Voluntarily, they have urged the support of their memberships and interests. In such an undertaking, with the unanimous support of every representative group in the industry, it is not conceivable that this program will be allowed

to lag.

Speaking of the foundry trade associations, it is my privilege to report, as I see it, that the relations between these groups and A.F.S. is better today than at

any previous time in my experience.

There exists a fine spirit of friendliness and cooperation, and a much better understanding of our respective functions. Thus you find considerable recognition of the trade associations and their accepted im-

portance.

I would be remiss if I did not also recognize here the part being played by the Exhibitors at the great International Congress, and particularly those exhibitors who are members of the Foundry Equipment Manufacturers Association and the Foundry Facings Manufacturers Association. In my estimation this was the greatest Foundry Show ever staged in conjunction with an A.F.S. Convention.

And finally, I should like to express to every member of our National Office Staff my congratulations for a splendid job in organizing this huge International Congress. Few of those attending this Congress had any conception of what a gigantic undertaking it was. For a Staff which usually is extremely busy, to take on a job of this size and carry it through with practically no additional help, I can only say in all sincerity, "Well done." To our Secretary, W. W. Maloney, our Technical Director, S. C. Massari, H. F. Scobie, Editor of AMERICAN FOUNDRYMAN, and T. B. Koeller, Advertising Manager, to all the Staff members, our sincere thanks and congratulations.

A.F.S. Sources of Income

A.F.S. today is operated on business lines and living within its means, with four major sources of income: Membership Dues, 40 per cent of total income; American Foundryman, 30 per cent; Convention and Exhibits, 20 per cent; Publications, 6 per cent; Miscellaneous, 4 per cent. The Board is endeavoring—as yet not accomplished—to build up a Reserve equal to an average year's expenditures, and until this is done, we cannot undertake needed additional activities without special financing. Thus, the raising of funds for the Safety, Hygiene, and Air Pollution program and the Building project.

A Society, like a business, either progresses or loses ground . . . it cannot stand still. No matter what our finances may be, we can always expect to have more activities urged upon us than we can adequately sup-

port or care for.

As an organization like this grows, and its activities continue to expand, there must necessarily be evolved some rules of conduct and procedure. Only by the adoption and guarding of certain standards and principles can A.F.S. maintain its reputation for integrity, and these basic principles are as follows:

First, A.F.S. must rely upon the faith and good will of many men, and solicit their active interest.

Second, its integrity must be constant, unimpeachable and proof against any pressure.

Third, it must condition the field it serves to consider and to embrace every potential means of progress.

Fourth, it must advocate nothing but analyze all, selecting for dissemination only that which meets accepted high standards.

Lastly, it must be prepared to lend its sponsorship to activities which, while outside its own sphere, de-

serve industry-wide acceptance.

These guiding principles govern the actions of your Board of Directors, the Committees and the Chapters. Especially must they guide the conduct of the Chapters, since the Chapters include 95 per cent of the total membership, and since each Chapter represents the technical interests of the foundries in its area.

It has been rightly said that "A.F.S. is a forum, not a platform," and by the same token the A.F.S. Chapters are not merely social clubs but the life blood of an International Technical Society whose work is impartial, unbiased and authoritative.

Many have labored to make this true. There exists a personal obligation on all of us . . . to keep it so.

REPORT OF THE TREASURER

(Fiscal Year Ended June 30, 1952)

ON JULY 18, following completion of the Annual Audit by George V. Rountree & Co., Certified Public Accountants, Chicago, the Treasurer forwarded to members of the Board complete financial data as follows:

(1) Auditor's Report as of June 30, 1952.

(2) Consolidated Statement of Income and Expense for 1949-50, 1950-51, and 1951-52.

(3) Proposed Income and Expense Budgets (included in the above Consolidated Statement) for fiscal year 1952-53.

The financial statements for 1951-52 reported the following: Income, \$734,778.50 against the Budget of \$657,400; Expense \$552,165.44 against the Budget of \$531,050. Thus, while we budgeted \$126,350 Excess Income over Expense for the fiscal year ended June 30, the actual Excess Income totaled \$182,613.06, or 44.5% more Excess Income than was budgeted.

Total Income of \$734,778, and Excess Income of \$182,613, both were greater than for any previous 12-month period in the his-

tory of the Society.

Income and Expense

The Auditor's Operational Statement shows that in an exhibit year only three major A.F.S. activities can be expected to show a net Income . . . in a non-exhibit year, only two.

Per Cent of Income from A.F.S. Activities

	1950-51	1951-52	Two Years 1950-52
Membership Dues	53.3%	29.5%	37.6%
AMERICAN FOUNDRYMAN	35.4	25.9	29.2
Convention & Exhibits	1.6	38.4	25.8
Publications	8.3	5.2	6.3
Miscellaneous	1.4	1.0	1.1
	100.0%	100.0%	100.0%

A comparison of Income and Expense per member in the past two exhibit years may be of interest, as follows:

	1949-50	1951-52
Total M/S on June 30	9,046	10,033
Total Income	\$578,785.	\$734,778.
Income per Member	\$63.98	\$73.24
Total Expense	\$458,312.	\$552,165.
Expense per Member	\$50.66	\$55.03

The above may be compared with Income and Expense per member in non-exhibit years, as follows:

	1948-49	1950-51
Income per Member	\$32.90	\$41.00
Expense per Member	\$39.90	\$46.20

Thus both Income and Expense per member have increased in the past two exhibit years; whereas Expense per member has increased \$4.37, Income per member has increased \$9.26.

The Auditor's statement shows an increase of 5 Sustaining and 40 Company members in 1951-52; 283 Personal Members (\$20), 467 Affiliate, 14 Associate and 10 Student and Apprentice... total 819. The increase of 45 Company and Sustaining members, admittedly with no Chapter effort and little concentrated effort by the National Office, convinces us that much can be done here.

Income By Activities

During the past two years the statement has been made that the Income of the Society is derived from several activities in the following relative importance over a two-year period: Membership Dues, 40 per cent; AMERICAN FOUNDRYMAN advertising, 30 per cent; Convention and Exhibit revenues, 20 per cent; Publications, 6 per cent; Miscellaneous, 4 per cent.

We have made a new compilation of Income sources during the two fiscal years 1950-51 and 1951-52 which bear out the above statement as to the relative Income-producing importance of our activities, with some variations in the percentages previously discussed. For these two years the percentages are as follows:

*Membership Dues	37.6%
*AMERICAN FOUNDRYMAN	29.2
Conventions and Exhibit	25.8
Publications	6.3
Miscellaneous	1.1
	100.0%

Balance Sheet

The financial position of the Society on June 30, 1952 was some \$215,964 better than on June 30, 1951, or an increase of 53 per cent. If we eliminate an increase of \$33,349 total in the fund principals of the Building, Awards and Safety & Hygiene funds, the increase in General Fund Principal becomes \$182,615 on June 30 this year.

A comparative Condensed Balance Sheet for the past four vears is included with this report, which statement shows a

total of cash and investments as follows:

			Cash	Investments	Total
June	30,	1952	\$297,608	\$330,600	\$628,208
**	22	1951	103,913	308,805	412,718
23	**	1950	229,497	118,805	348,302
**	**	1949	85,785	133,805	219,590
**	**	1948	106,394	178,805	285,199

The increase in investments from \$308,805 a year ago to \$330,600 is due to an additional \$35,000 in U. S. Government securities purchased on behalf of the Building Fund from surplus cash therein.

In observing the Comparative Condensed Balance Sheet statement, it should be pointed out that Accounts Receivable totals for 1951-52 are not comparable with the previous two years, because the figures for the past two years include pledges receivable to the Building Fund (\$16,192) and to the Safety & Hygiene Fund \$ (12,747).

It will be noted that Inventories increased some \$5,000 during the past year, due to the publication of several new books. Little change is noted in either the Deferred and Prepaid items, or in Furniture and Fixtures.

On the Liabilities side, it will be noted that Current Liabilities total less than \$14,000, a good condition. The Reserve set up in 1951 for Canadian Bank Exchange losses has been eliminated this year because the Canadian dollar now is quoted at \$103.65.

The Deferred Dues Income item, \$120,446, compares with \$109,735 a year ago and is the result of dues income being on an accrual basis for accounting purposes.

Fund Principals

The four Fund Principals of the Society are analyzed in the Auditor's Report, as well as the schedule of Award Fund Principal and Securities. These statements show that the total of all Fund Principals of the Society was \$622,775 on June 30, 1952, compared with \$408,811 on June 30, 1951.

Operational Statements

In the Auditor's Report, Operational Statements are shown for each of the major activities of the Society, and all activities are included in a Condensed Operational Statement. It will be noted that in these statements, all Salaries and Burden items have been allocated to the various Society activities, giving a true picture of all Income-producing and Expense-incurring activities maintained today.

Conclusion

The Society's finances appear to be in excellent shape at the end of the Exhibit year 1951-52, with \$236,715 in General Fund cash available for further investments or as operating funds between now and the time when Exhibit Space Rental revenues again become obtainable in January, 1954. In other words, the cash balance as of July 1 must first be utilized to overcome the anticipated Excess Expense of the next 18 months, estimated between \$55,000 and \$70,000 for budget purposes.

Expenses of the Society are under constant control of the Finance Committee and Treasurer, and all Staff members responsible for expenditures are kept continually cost conscious, in keeping with the requirements of their activities.

Respectfully submitted,

WM. W. MALONEY, Treasurer

(*No consideration given to membership subscriptions at \$3 per member, required under second class postal regulations.)

Condensed Statement of Income And Expense

(All Expenses Distributed to Major Activities) Fiscal Year July 1, 1951-June 30, 1952

INCOME		
		Income
	Amount	Per cent
Membership Dues	\$185,493.57	25.3
General Publications	13,492.40	1.8
Special Publications	24,184.38	3.3
AMERICAN FOUNDRYMAN	219,836.38	30.0
Convention and Exhibit	283,052.00	38.6
Income from Investments	5,545.01	0.8
Miscellaneous Income	1,344.30	0.2
TOTAL INCOME	\$732,948.04	100.0
TOTAL EXPENSE	\$550,334.98	
Excess Income over Expense	\$182,613.06	
EXPENSE		==
	Total	Expense
	Amount	Per cent
Membership Service	\$ 26,407.29	4.8
Research	25,014.68	4.5
Other Technical Activities	24,026.93	4.4
General Publications	31,738.61	5.8
Special Publications	29,537.94	5.4
AMERICAN FOUNDRYMAN	188,655.28	34.3
Chapter Operations	68,325.92	12.4
Convention and Exhibit	100,331.72	18.2
General Administration	40,945.90	7.4
Retirement Plan Expense	15,350.71	2.8
TOTAL EXPENSE	\$550,334.98	100.0
		===

Comparative Condensed Balance Sheets

As of June, 1949 to June, 1952, inclusive (4 years)

Assets	June 30, 1949	June 30, 1950	June 30, 1951	June 30, 1952
Cash	\$ 85,785.57	\$229,497.13	\$103,913.45	\$297,608.34
Investment Securities	133,805.60	118,805.60	308,805.60	330,600.00
nventories	35,567.65	35,354.99	27,817.59	32,470.00
Accounts Receivable	4,405.03	22,011.00	49,975.09	46,376.91
Deferred & Prepaid Items	19,533.00	31,530.03	31,258.11	34,378.06
Furniture & Fixtures (net)	9,791.42	16,451.18	15,858.16	15,559.74
TOTAL	\$288,908.27	\$453,649.93	\$537,628.00	\$756,993.05
LIABILITIES				
Current Liabilities	17,760.49	24,867.45	19,736.66	13,771.93
Deferred Income	71,217.06	108,081.29	109,735.45	120,446.02
Reserve for Exchange	-	_	1,344.30	-
und Principals	199,930.72	320,701.19	406,811.59	622,775.10
TOTAL	\$288,908.27	\$453,649.93	\$537,628.00	\$756,993.05

REPORT OF THE SECRETARY

(Fiscal Year Ended June 30, 1952)

This report sums up eight major activities of the Society, with some comments on problems affecting several. The report is confined largely to non-financial matters and to general administrative matters. Financial and technical activities are discussed in separate reports of the Treasurer and Technical Director.

1. Membership

During the past year, A.F.S. membership reached a total of 10,033, a net gain of 827 or approximately 8.9 per cent over the total of 9,340 on June 30, 1951. The budget estimate was for 10,000 members, and the results bear out the prediction of a

year ago that the trend is definitely upward.

The gain in membership occurred in all classes: Sustaining up 5, Company up 40, Personal up 283, Affiliate up 467, Associate up 14, Student and Apprentice up 10. This too reversed the trend of the previous year, when 11 Sustaining and 18 Company members were lost. Most encouraging is the increase of 283 in Personal memberships, for this class provides the nucleus for increased Company memberships.

A compilation of membership gains and losses over the past

five years is shown below:

	1947- 48	1948- 49	1949- 50	1950- 51	1951- 52
Members Gained (all Classes)	1,979	1,755	1,787	2,316	2,876
Members Lost (all causes)	1,259	2,095	2,804	2,022	2,049
Net Gain (or Loss)	720	(340)	(1,017)	294	827
Avge. Gain (Loss) per Month	60	(28.3)	(84.7)	24.5	68.9
Total as of June 30	10,403	10,063	9,046	9,340	10,033
Rate of Turnover	12.1%	20.8%	31.0%	21.6%	20.4%

From the above it should be noted that the rate of turnover during 1951-52 was the lowest in 4 years, that the total membership gain was the largest in 5 years, and that the net gain was the best in 5 years. It should also be stated that the total of 2049 members lost in 1951-52 was 60% more than the total lost in 1947-48, a year which showed a net gain of 720. Thus the problem still remains, as has been pointed up sharply over the past four years, to hold our present membership.

The increased membership during the past year may be attributed to several factors: Concentration on membership in discussions between the Directors and the Chapters, greater interest in the Society as a result of the International Congress, and urging of membership "targets" by the National Office.

In connection with membership targets, the Chapters seem to like the idea for it gives them something definite to shoot at in their membership work without the adverse effects of a full-fledged "campaign." While we should continue the target idea, we suggest that the Directors during 1952-53 avoid urging mem-

bership work as a high-pressure campaign.

During the past year, 18 Chapters made their membership targets as of June 30, against only 8 the previous year. One Chapter (Northern California) which made its target on May 31, dropped below on June 30; two Chapters (Birmingham and Eastern New York) which failed to make it on May 31, did so by June 30. We believe that all Chapters who made it on May 31 or June 30 should be awarded the Membership Certification.

Targets' are considered as totals only and are not based as classes of membership. We endeavored once to target classes as well as totals, but this proved ineffective. At least three Chapters might easily have made their targets with a little extra work, and 7 Chapters actually lost members (totaling 81) during the year. These 7 Chapters are: Central Ohio, Mo-Kan, Ontario, Rochester, Saginaw Valley, Timberline and Western New York.

We must continue to seek greater horizontal membership coverage among the foundry plants, since membership is circulation for American Foundryman and a high turnover of circulation is not desirable from the magazine standpoint. The following

recommendations are offered for 1952-53:

(1) We suggest that the Chapter Contacts Committee continue to concentrate on membership work in Chapter visits, not merely to add more members, but on the sound principle that A.F.S. can best serve the industry in proportion to the number of foundrymen exposed to information on better methods.

(2) We believe that greater effort can and should be made toward the conversion of Personal to Company membership and from Company to Sustaining membership. At the recent Chapter Officers Conference, only one or two Chapters stated that their membership work included any conversion activities. We believe this can be done, as shown by the increase of 40 Company and 5 Sustaining members during the past year.

(3) We intend to renew membership targets during the coming year with a goal of 11,000 for June 30, 1953, without any new Chapters, or 11,500 if several of the present new Chapter

possibilities are realized.

A few membership points should be mentioned here for information purposes: (1) The pro-rationing of membership dues in the Chapters continues. (2) Members who enter military service are carried on the rolls without loss of membership and without further payment of dues. (3) Student members who graduate are carried at the Student rate until their dues become renewable, even though they enter employment. (4) Some complaints on processing time, some of which were justified and some due to slow Chapter approvals, have come to us during the past year, and we intend to concentrate on this point.

Since 1948 A.F.S. has maintained three Canadian bank accounts, established when the Canadian dollar was discounted at 10 per cent below the U. S. dollar. Now, however, the Canadian dollar is worth more than the American dollar, and we therefore intend to recommend to the new Board tomorrow that the Canadian bank accounts be discontinued as no longer serving a purpose,

2. Chapters

One new regular Chapter (Corn Belt at Omaha, Neb.) and two new Student Chapters (University of Michigan and Brooklyn Polytechnic Institute) were established during 1951-52. A number of other Chapter possibilities developed or continued

to show interest during the year.

While it continues to be our policy not to create Chapters, but to encourage and cooperate with new Chapter movements, last December the Board agreed to officially invite foundry groups to consider Chapter status where the prospects indicate that such an invitation might be desirable. Four such invitations were issued during 1951-52: to the New England Foundrymen's Association, the Lehigh Valley Foundrymen's Association, the Conestoga Foundrymen's Group, and the Reading Foundrymen's Association. No encouraging response has been received to date from any of the four groups so invited.

We are now in correspondence with two other groups interested in Chapter status, one at Memphis, Tenn., and one at Norfolk, Va. Memphis now is in non-Chapter territory and the movement has not yet developed with sufficient strength to warrant a prediction. The Virginia group, termed the "Tidewater" group, is being encouraged by the Chesapeake Chapter, and correspondence during the past week indicates that they may want to go this fall. In addition a foundry group in Alberta is interested in forming a section of the British Columbia Chapter, and a very active group in East Texas has formed a section of the Texas Chapter.

During the past year, President Seelbach attended four of the seven Regional Conferences held, and Vice-President Wagner three. The Vice-President also visited the three Northwest Chapters with the Secretary, in October. Compilation of Chapter visits (according to our information) by Officers, Directors and Staff shows the following: President 11, Vice-President 15, Secretary 19, Vice-President-elect 1, Technical Director 10, Editor 12, Safety & Hygiene Director 4, and the Directors visited 49 Chapters at least once.

The Chapter Contacts bulletin, "Off-The-Record," was published six times during the year in an effort to keep the Chapter Officers and Directors posted on various matters. In addition, a minimum of three membership bulletins were mailed to the Chapters every month: New Members Report, Monthly Membership Statement, and Statement of Membership Targets.

The 9th Annual Chapter Officers Conference was held at the Sherman Hotel, Chicago, June 12-13, with a total attendance of 107, including 82 Chapter delegates. All 41 Chapters were represented, including the new Corn Belt Chapter, all but 7 with at least two delegates and 5 with 3 delegates. The total included 37 Chairmen, 36 Vice-Chairmen-Program Chairmen, 2 Secretary-Treasurers, 3 Secretaries and 1 Treasurer. Chairman I. R. Wagner presided and 11 National Officers and Directors were present, less than attended the 1951 Conference, probably due to the early date following the Congress.

During the year 7 Regional Conferences were held by the Chapters at East Lansing, Mich.; Vancouver, B. C.; Purdue University; Davenport, Iowa; Houston, Texas; Milwaukee, Wis.; and Birmingham, Ala.; and a "Sub-Regional" Conference at Three Rivers, Que. The Birmingham Conference now is sponsored jointly with the Tennessee Chapter; the Texas Conference in 1952 will be jointly sponsored with the Tri-State Chapter, and the Michigan Conference hereafter will alternate between Ann Arbor and East Lansing.

Last year the Board expressed the feeling that the National Office should participate more actively in the organization and operation of Regional Conferences, in the interests of improved programs and operations. To date, with the exception of program assistance during the Chapter Officers Conference, this has

not proved feasible.

3. Convention

A total of 13,303 attended the International Foundry Congress & Show and the 56th Annual Convention of A.F.S. in Atlantic City, May 17. The total included 4274 members, 3360 guests, 3282 exhibitor representatives, 275 International guests, 1442 "International Day" guests, and 670 registered ladies. Registration fees of \$2.00 for Members and \$5.00 for Non-Members were charged and seem to have been accepted with few complaints, except from those who did not read our bulletins.

The housing of visitors was exceptionally easy this year, with few complaints. There were some complaints of high hotel and meal charges, and many plants sent less men than usual, although more foundry management attended with their ladies, as expected. We had splendid cooperation from the Atlantic City Convention Bureau (Messrs. Al Skean and Wayne Stetson), from the Convention Hall, and from the service organizations.

The Exhibit proved a sellout, with 243 paid exhibitors using 83,760 sq. ft,, the largest Exhibit since the 100,700 sq. ft. in 1946. There were few complaints on excessive charges to exhibitors, there was no labor trouble of any kind, and the entire Exhibit was smoothly handled by Al Hilbron and his floor managers. A total of 68 technical sessions, 30 Committee meetings, and 25 other events were staged during the Congress week.

The International Banquet established several innovations, including an attendance of 1175. The program proved to be too long, although practically everyone stayed to the end. Although warned in advance, we observed no difficulty in the matter of wine service, and we did notice that most of the objectors were enjoying the service. The Banquet proved to be a highlight of

the Congress.

Last year we predicted that we might expect 100 foundrymen from abroad at the International Congress. Actually the total was closer to 275 from 31 different countries. A.F.S. arranged two Study Tours in advance of the Congress and a total of 148 took advantage of these tours. The Mutual Security Agency of the Government provided expenses for 113 delegates in the United States, each delegate being required to pay his own way to the United States, and 35 other delegates participated as a Thos. Cook & Son party. A total of 145 plants were provided on the Study Tours due to the cooperation of our Chapters in Buffalo, Cleveland, Detroit, Chicago, Cincinnati, Philadelphia and New York, and the Pittsburgh Foundrymen's Association.

Promotional effort for attendance included talks before the Chapters by Directors and the Staff, posters for Chapter display, posters for display by Company and Sustaining members, 75,000 folders distributed free to exhibitors, advertisements in "Iron Age" and "The Foundry," slogan cuts for exhibitor ads, special A.F.S. letterheads, and bulletins or articles in American Found-RYMAN and "Off-The-Record." A registered attendance list is

The results of the International Foundry Congress may be summed up from two communications received and one post-Congress event. On June 6 the Committee of Exhibitors met and unanimously expressed the opinion that, while they had been against Atlantic City to a man in 1951, they admitted that they had been wrong and considered the 1952 International an outstanding success . . even though they recommended not returning to Atlantic City more frequently than every eight or ten vears.

The following communication was received from the National

Castings Council recently:

RESOLVED, That President Seelbach, Vice-President Wagner, Secretary Maloney, Technical Director Massari,

Convention and Exhibit Manager Hilbron and the Staff of the A.F.S. to be commended highly for the success of the 1952 International Foundry Congress and Show at Atlantic City, with special emphasis on the general arrangements of the entire week's program, the splendid exhibition of foundry equipment and supplies, the excellent program of technical papers and discussions, and the fine delegation of overseas guests."

The following communication was received from MSA re-

cently as a result of the Tours:

"We who have benefited so greatly desire here to express our gratitude to the Government and people of the United States of America, to the officials of the Mutual Security Agency who have supervised our study tours, to the American Foundrymen's Society who made arrangements for our plant visits, and whose hospitality we enjoyed at several points on our tours and at Atlantic City, and to the man-agements and staffs of the foundries that we visited who gave so freely of their time and information.'

It might also be well to examine some of the lessons learned from this International Congress for future guidance:

(1) A seven-day foundry show was proved to be too long. (2) It doesn't appear wise to hold the exhibits open on a Sunday, for it offends many.

(3) Advance registration has definite advantages and should

be continued and broadened.

(4) We need a better advance registration of exhibitor representatives confined to actual booth representatives, a point on which the Committee of Exhibitors has agreed.

(5) We need better safety regulations as to exhibitor installations. The Committee of Exhibitors has agreed to our proposal that no exhibitor be allowed to open until his exhibit has been inspected and approved by the A.F.S. Safety & Hygiene Director.

(6) A Banquet with entertainment seems to be universally desired, including the presentation of awards but no featured speaker, the awards to be presented briefly as in 1952. The show could have been shorter this year, probably with a maximum of 4 acts at a cost of about \$2,000. We also believe the service of wines was readily and generally accepted.

(7) We believe it was a mistake to eliminate all night sessions, something tried because Atlantic City is not close to any foundry

(8) Some effort probably should be made to control the number of sessions and papers scheduled by the A.F.S. Divisions, so as to give more time for personal contacts in non-exhibit years and to inspect exhibits in exhibit years. Also, we were forced to add additional papers this year because of exchange papers received, but this cuts down discussion time.

(9) We believe that the Chapter Officers and Directors Dinner should definitely be eliminated in exhibit years because of too

many counter-activities.

(10) We believe the Student Delegate Plan inaugurated this year should be continued in exhibit years only, with an attendant essay contest. Someone, of course, would have to provide the prize money, unless A.F.S. undertook it.

(11) The A.F.S. Booth should be located more prominently at

future exhibits, for many did not even see it at all.

(12) We intend to try some 8:30 am after-breakfast sessions in 1953, since the technical sessions do not begin until 10:00 am.

(13) Some complaints were received on the type of talk given at the Annual Lecture, as being outside A.F.S. activities. We believe the Annual Lecture needs better definition of its purpose, so that all will understand, and for the guidance of the Committee, which is a Committee of the Board.

4. American Foundryman

In 1950 when advertising representatives were first utilized, a three-year budget for display advertising was approved by the Board. During 1951-52 and for the second straight year, the income budget was surpassed, this time by \$18,000 gross. Ameri-CAN FOUNDRYMAN advertising now seems well established and while those who set the three-year budget thought it then quite optimistic, the record speaks exceptionally well for efforts of Advertising Manager Koeller and the Advertising Representatives.

Closer contacts have been maintained with equipment and supplies firms, the total of 664 advertising pages in 1951-52 being a new high, the previous high being 488 in 1950-51. Membership is audited by the Audit Bureau of Circulations (ABC) and gives the magazine a definite standing with advertisers and advertising agencies. Display advertising rates were increased 12 per cent effective July 1951, the full effect of the increase not being felt until July 1952.

In addition to distribution to the membership, a certain number of copies are mailed each month to non-member foundries as "circulation support." We believe that A.F.S. should continue

"circulation support" for American Foundryman.

Editorially, the magazine was completely redesigned in March 1952 in order to give it a fully professional appearance as well as content. Many excellent articles are being printed, and during 1952-53 we intend to publish in the magazine the best of the 1952 Convention papers.

Certain policies covering American Foundryman may well be repeated here as a refresher and to assist the new Directors:

Advertising is sold only on its merits, and no pressure of any kind is exercised under penalty of discrimination in other activities.

No attempt is made to claim greater readership than is known and verifiable.

No special rates or rebates are permitted.

Arrangement of editorial content is primarily for reader convenience.

Editorials are prepared by men of industry, not by the Staff alone.

A.F.S. does not pay for any article published.

Articles are not published merely to please advertisers and gain advertising.

5. Technical Activities

The Technical Director's report will fully cover these activities, although a few general comments are called for.

We now have a full-time Staff man on the Safety, Hygiene, and Air Pollution Program. K. M. Morse joined the Staff on August 1, 1951, as the Program Director, but left unexpectedly on March 1 to go with the U. S. Steel Corporation. As of May 1, W. N. Davis is now Program Director, formerly Senior Engineer for the National Safety Council.

Contributions to the Safety, Hygiene, and Air Pollution Program have not as yet come in strongly. The Staff was prepared in December 1951 to initiate vigorous solicitation, but the National Castings Council requested and was given the first opportunity to urge support by their respective members. The result to date is \$42,000 contributed, which is a good response, but by no means sufficient to finance the Program with assurance.

When the original impetus of the NCC effort died down, the A.F.S. Staff was too occupied with the International Congress to resume it. We now are placing Safety & Hygiene solicitation first on the list of important jobs in 1952-53. First on the list to be solicited will be those who recommended the Program in December 1949 and the members of the Program Steering Committee. It would also be most helpful if the A.F.S. Directors would also indicate the support of their companies.

6. Publications

Technical publications will be covered in the report on the Technical Director.

Last year we pointed out that the manuscripts for 10 or 12 important publications had been delayed some time and stated, "Should a considerable number of these publications be completed during the 2 years, 1951-53, a real problem would be created as to the ability of the present staff to get them all published within a reasonable period of time." This has happened, and the manuscripts for the following publications were fully completed in 1951-52 or are nearly complete now: Foundry Work—pub. Nov. 1951; Sand Handbook—pub. Jan.; Symposium on Gating—pub. June; Copper-Base Alloys—pub. June; Patternmaking Manual—manuscript complete; Symposium on Quality Control—manuscript complete; Glossary of Foundry Terms—manuscript complete; I0-Yr Index to Transactions 1941-51—in process.

In addition, and in lieu of author delays, other publications have been in process and these also are coming due in 1952-53, as now planned: Dust Control Manual, Symposia on Safety, Hygiene, and Air Pollution, Symposium on Molding Methods. The College Foundry Textbook, it now appears, at long last is due this fall after a lapse of time due to changing authors.

The accomplishment of producing 7 new publications in 1951-52, in spite of International Congress pressure, we believe de-

serves mention.

7. Building Project

Great effort has been spent during the past year to find a proper site for the Permanent Headquarters Building, culminating in selection of a property in DesPlaines, Ill., last January. Application for rezoning the property for our purposes was submitted on January 11 and after four of the five necessary steps that must be taken on a petition for rezoning, the DesPlaines City Council turned down the petition on May 5 by a five to four vote. Since then we have examined a number of sites in suburbs near Chicago, the best available of which appears to be quite a distance from the city. We are seeking a one year's extension on our lease in anticipation of building delays.

The DesPlaines project was revived in June when the owner of the property we desired threatened a mandamus suit against the city to force rezoning for commercial purposes. Although we questioned that the interests of the citizens residing in the vicinity were best served by permitting a court judgment to decide what sort of building may be erected on the site in question, the DesPlaines City Council again took adverse and final action on our petition on July 21. We now will take prompt steps to seek a proper site elsewhere. It is unfortunate but unavoidable that the indecision of the DesPlaines City Council has held this matter in abeyance for over six months, and no further site will be considered involving any question of rezoning, even though the corollary to this decision may be the selection of a site less desirably located.

8. General Administration

Four changes in the Executive Staff occurred during the year. We now have a competent new bookkeeper, E. R. May. W. N. Davis has replaced Kenneth M. Morse as Safety, Hygiene, and Air Pollution Director. C. R. McNeill was moved from the American Foundryman staff to the Technical Dept., and H. F. Hardy was hired to replace him. On July 1, R. N. Sheets, Asst. Editor, resigned to take a position in New York.

Six issues of the "Bulletin Board" for National Officers and

Six issues of the "Bulletin Board" for National Officers and Directors were put out during the past fiscal year in an effort

to keep the Directors up to date on our activities.

9. Recognition

We desire to extend our sincere appreciation for the counsel and guidance of the Officers and Directors during the past year, and for the cooperation of our Chapters, Divisions and Committees in achieving an outstanding year. The membership was brought back up to over 10,000, the previous high. The International Congress was an event that brought new prestige to the Society. The Exhibit was a sellout and the most successful financially to date. Advertising in American Foundryman is at a new high, both as to gross and net, and is now well accepted. Our list of foundry publications is the best available anywhere and sales were very good during the past year. As shown in the Treasurer's Report, gross Income reached a new high, as well as a new high for Excess Income over Expense. The leadership of President Seelbach undoubtedly has advanced A.F.S. greatly in recognition by the entire castings industry.

Respectfully submitted, Wm. W. MALONEY Secretary

Breakdown of A. F. S. Membership

Chapter	Sustaining	Company	Personal	Personal (Affiliate)		Student & Apprentice	Honorary	Tota
Birmingham	6	29	75	154	7	3	2	276
British Columbia	0	12	17	20	4	1	0	54
Canton District	1	25	22	73	2	0	0	123
Central Illinois	2	10	23	119	1	5	1	161
Central Indiana	4	33	56	139	2	9	0	243
entral Michigan	1	14	41	94	6	1	0	157
central New York	i	29	52	94	2	3	0	181
central Ohio	3	22	44	79	4	0	2	154
chesapeake	3	24	55	50	38	1	8	179
hicago	22	94	193	442	15	14	6	766
Cincinnati	13	41	85	114	10	7	3	273
	0	5	1	35	0	ó	0	
orn Belt						-		41
Detroit	11	46	126	231	15	5	12	446
astern Canada	6	42	61	174	18	16	0	317
astern New York	1	13	24	61	6	3	0	108
fetropolitan	9	63	137	142	30	13	6	400
fexico City	0	7	33	19	0	2	0	61
lichiana	2	25	64	128	1	0	0	220
lo-Kan	1	6	32	8	2	1	0	50
ortheastern Ohio	21	74	113	333	9	12	9	571
orthern Calif,	4	19	88	93	10	2	0	216
o, Illinois & Southern Wisconsin,	3	16	22	59	1	1	1	103
orthwestern Pennsylvania	2	21	44	109	5	i	0	182
ontario	4	67	113	125	4	0	2	315
regon	2	8	26	75	3	0	0	114
	14	49	111	131	10	2	5	322
hiladelphia								
uad City	5	36	56	148	4	3	2	254
ochester	1	9	41	32	1	0	0	84
aginaw Valley	5	16	25	176	1	14	2	239
Louis District	1	30	110	113	7	2	0	263
outhern California	3	46	133	88	14	3	0	287
ennessee	0	19	40	71	3	4	0	137
exas	5	30	64	80	4	0	2	185
imberline	1	5	24	19	3	0	0	52
oledo	1	19	22	70	1	0	1	114
ri-State	0	10	45	24	2	0	1	82
win City	5	25	61	74	7	0	î	173
	0	6	35	8	10	4	î	64
Vashington							1	
Vestern Michigan	3	30	38	153	4	4	1	233
Vestern New York	0	36	81	83	3	1	1	205
Visconsin	16	79	105	334	12	10	3	559
TOTAL IN REGULAR CHAPTERS STUDENT CHAPTERS	182	1190	2538	4554	281	147	72	8964
lassachusetts Institute of Technology	_	_	_	_	1	34	_	35
ichigan State College				_	_	29		29
issouri School of Mines					3	44	12	47
	_	_	_	-	3	12		12
orthwestern University		_	_	_		20	_	20
hio State University		-		-	_		-	
regon State College	_	-	_	-		22	_	22
nnsylvania State College	_	_	-	_	_	28	_	28
olytechnic Institute of Brooklyn	_	_	_	_	_	27	_	27
exas A & M	_	_	-	_		17	-	17
niversity of Alabama	_	_	_	_	_	23	-	23
niversity of Illinois	_	_	_	_	-	23	_	23
niversity of Michigan	_		_		-	72	-	72
niversity of Minnesota (Inactive)		_	_	_		1	_	1
TOTAL IN STUDENT CHAPTERS					4	352		356
TOTAL IN ALL CHAPTERS	182	1190	2538	4554	285	499	72	9320
					19	10	2	431
nternationalon-Chapter	9	23 69	367 118	10 50	29	2	5	282
	-					-		

REPORT OF THE TECHNICAL DIRECTOR

(Fiscal Year Ended June 30, 1952)

This progress report covers the period beginning July 1, 1951 and ending June 30, 1952:

Technical Activities

For reasons which probably are apparent, it not only becomes necessary in many instances, but also highly desirable in others, to modify the individual committee rosters so as to maintain committees having well qualified personnel and to eliminate those who, from past experience, are inactive and tend to hamper profitable activities. It is considered highly desirable, therefore, that a new National Committee Personnel Roster be prepared each year, rather than every other year, even though Divisional Officers are ordinarily only elected every two years. Furthermore, examination of the present roster makes it apparent that the Board of Directors, as well as the President and Vice-President changing each year, automatically make a roster obsolete. As a matter of interest, we now have 108 active technical committees. For the reasons enumerated, the Technical Director recommends that the Board of Directors authorize publication of a new roster again this Fall.

It is noteworthy that the Aluminum & Magnesium Division has recognized the importance of Titanium as a new engineering material of increasing prominence. Following a vote by the Executive Committee of that Division, they have requested that the Board of Directors authorize them to include the founding of Titanium and its alloys as part of its authorized activities. The Technical Director believes this would be a logical acquisition and that the Division is well qualified to undertake activities in this field since Titanium is in the category of a light-weight, high-strength structural material.

The Division also questions whether or not its name should be changed, in view of such additional activities. While the suggestion has merit, there are distinct advantages in changing the name of a Division. If it were to be changed, however, the Technical Director would recommend a name such as "Light Metals Division," so as to be more inclusive.

In view of the substantial activity in the field of Chemical Analysis, particularly relating to accepted and standardized procedures by the American Society for Testing Materials, and in view of the apparent lack of such interest by members of the Foundry Industry, it has been decided to abolish the A.F.S. General Interest Committee on Chemical Analysis.

A.F.S.-Sponsored Research

A complete and detailed report covering the present status of A.F.S.-sponsored research projects has been prepared and transmitted to the Research Committee of the Board of Directors, comprised of Messrs. Farquhar, Dietert and MacKenzie. The Chairman has reviewed the recommendations contained therein and will make some recommendations to the Board covering the appropriations for the coming fiscal year. The Staff is appreciative of the interest and helpful recommendations made by the Research Committee and the understanding attitude of the Board of Directors in this important field of technical activities of the Society.

1952 Convention Program

Little need be said as to the success of the International Foundry Congress, based upon the wide variety of technical sessions which were scheduled and the substantial number of papers which were involved. Of a total of 110 papers which were offered, 105 were approved by the Program & Papers Committees for presentation, the remaining ones having been rejected and returned to the authors as not conforming with the publication policy of the Society. Of these approved, 72 were preprinted and five were published in the April or pre-convention issue of American Foundryman.

A total of 68 official functions were scheduled during the

International Foundry Congress, exclusive of individual committee meetings. In excess of 4,000 requests were received from members of the Society for one or more preprints and if received before the specified deadline date, the preprints were mailed approximately six weeks in advance of the Congress, so as to provide members sufficient opportunity to review the papers and come prepared to participate in their discussion.

It should be noted that the number of requests for preprints has exceeded any previous year since 1946; likewise, the number of papers preprinted and made available on a gratis request basis, is almost twice as great. Both of these factors have contributed to unusually high costs for the production of the preprints for this year's Annual Meeting. In an ordinary year, seldom more than four or five papers are received from foreign sources as Official Exchange Papers; this year they totalled 15.

Preprinting of convention papers is an important service to the membership and should be carefully considered. However, it may be that this year we have over-extended ourselves in this particular activity and the Board may care to give consideration to limiting the number of papers which are preprinted so as to reduce the cost of this particular effort. The writer wishes to further emphasize that this is one of the few gratis services given to members of the Society and constitutes an important service in a technical organization such as ours. It therefore deserves careful evaluation before any specific policies are established.

For the first time in a number of years, we had two important Safety, Hygiene and Air Pollution sessions, a valuable step in our progress in this field. The papers were excellent and it is our intention to publish them as a separate booklet as evidence of our initial efforts in this field.

This is the first year in which we did not employ the services of professional sterotype recorders to make a record of the oral discussion following the presentation of the technical papers. Instead, each of the Program & Papers Committees were asked to appoint a recorder to screen the discussion, make a record of those individuals who made worthwhile contributions, and contact them later in order to gain the substance of their remarks, and preferably, to have them submitted in written form.

At this time it appears that this procedure has been sufficiently successful to justify its continuation for another year. Approximately only 12 to 15 recorders reported that either the discussion was not of sufficient importance, or irrelevant to the subject matter of the papers, and the balance of the written material submitted by the reporters is considered acceptable, as well as being concise and pointed. At this time all of these reports are in our hands thereby materially assisting in the production of the Annual Transactions, since in past years many of the participants did not return their edited discussions until late in August.

Safety, Hygiene and Air Pollution Program

As most of you probably are aware, we lost the services of K. M. Morse on March 1 of this year, to take a new position with the United States Steel Corp. at Pittsburgh. In his place, through good fortune we were able to obtain the services of Wm. N. Davis, who formerly was associated with the National Safety Council, and he is actively proceeding with the technical aspects of the program.

A complete report will be presented by F. W. Shipley, Chairman of the Safety, Hygiene and Air Pollution Committee of the Board of Directors, and therefore details relating to the program are not enumerated in this report.

Publications

The Annual Transactions of the Society for 1951 (vol. 59), comprising approximately 600 pages, was published and shipped during the month of December 1951 to all who had ordered it on a pre-publication basis. At that time a total of 1259 copies were shipped, consisting of 808 domestic, 152 foreign and

78 Mexican and Canadian, as well as 221 which were supplied gratis to Honorary and Sustaining Members of the Society who requested them. Since that time, of the total of 1500 copies produced, additional ones have been sold at the post-publication price of \$8.00 to members and \$15.00 to non-members, leaving a residual stock of only 95 copies, considered essential for future sales.

During the fiscal year the following publications were produced and made available for sale:

title and street and an arrangement and an arrangement				
			Non-	
	M	ember	Member	ľ
	Printing	Price	Price	
Cupola Research Committee Reports	1,000	\$1.00	\$1.50	
*Foundry Work (high school textbook)	10,000	1.76	1.76	
Foundry Cost Methods (reprinted in				
our office)	500	1.50	3.00	
Foundry Sand Handbook (6th Edition)	5,200	3.50	5.25	
Symposium on Principles of Gating	3,000	4.00	5.75	
Copper-Base Alloys Foundry Practices	3,000	3.75	5.75	
Fundamentals of Design, Construction				
and Maintenance of Exhaust Systems				
(reprinted)	600	3.00	5.00	
Permanent Mold Castings Bibliography				
(reprinted in our office)	250	1.50	3.00	
Gray Iron Research Progress Report No. 1				
(reprinted in our office)	250	1.00	2.00	
Annual Lecture for 1952 "Industry's				
Responsibility to Youth"	1,000	.25	.40	

^{*} Published under contract with John Wiley & Sons.

Unfortunately, it was necessary for Professor P. E. Kyle of Cornell University, for reasons beyond his control, to withdraw as the author of the College Foundry Textbook. Fortunately, we have been able to obtain Professors R. W. Heine and P. C. Rosenthal of the University of Wisconsin as joint authors of this text. They are energetically pursuing the preparation of the manuscript and it is planned that early in 1953 the manuscript will be available for bids by several of the prominent technical book publishers.

A considerable amount of work has been done on the Glossary of Foundry Terminology, but substantial additional work remains to be done, since we are completing it when time can be spared from more pressing activities. We recently cooperated with our European friends in arranging for a review of the glossary being prepared in French, by supplying the American equivalent of the French terms and transmitting same to them.

At the present time we have in process the preparation of an Index to A.F.S. Transactions for the 10-year period 1941-1950, inclusive, already two years past due, and should be published some time in the next few months.

The Patternmakers' Manual is in the final stages of editing prior to obtaining bids from the printers. Unless something unforeseen happens, we estimate that this book will be available for sale by approximately October 15.

The manuscript for "Foundry Applications of Statistical

The manuscript for "Foundry Applications of Statistical Quality Control Techniques" has been received and will be published as soon as previously mentioned books are completed.

The Committee on Gating & Risering of the Gray Iron Division has prepared an excellent chart containing accepted nomenclature and illustrations of the various types of gates and risers in general usage in the Industry. Subject to final review of the Committee, which should be in the very near future, we will undertake the publication of this work as a wall chart and also in 6x9-in. booklet form. It will be a long needed contribution in hopes of achieving uniformity of nomenclature and avoid misunderstanding between foundrymen.

Although it will not be included in the recommendations of the Publications Committee, the Society is in possession of a very extensive "Sand Abstracts" comprising all technical papers contained in the Annual Transactions of the Society up to 1950. These abstracts were prepared by Richard Voelz, serving under a Fellowship provided by the National Engineering Co. at the Northwestern Technological Institute of Northwestern University. The manuscript has been provided at no expense to the Society and because of the broad nature of it and interest on the part of foundrymen in this common problem, namely, Sand, it is believed that the Board of Directors should give additional consideration to the authorization for publishing this work, even

though the Publications Committee has not acted wholly favorably nor made a final decision.

Advance or pre-publication order forms for the Annual Transactions of the Society for 1952 (vol. 60) were sent to the entire membership about August 1. Subscriptions to Transactions again are on a prepaid pre-publication basis at a price of \$6.00 to members and \$15.00 to non-members of the Society, with gratis copies to all Sustaining and Honorary Members, on request. Subsequent to the close of the pre-publication offer, September 30, 1952, members are charged \$8.00 per copy and non-members \$15.00.

Additional publications whose production is contemplated, are

Symposium on Safety & Hygiene, No. 1.

Symposium on Safety & Hygiene, No. 2

Dust Control Manual Safety & Hygiene Recommended Practices

Symposium on Molding Methods

Technical Inquiries

Technical inquiries continue to be received at National Headquarters and when possible, are answered promptly reflecting the best available information at our disposal. During the past year a total of 466 individual requests were received. Unfortunately, because we lack a well qualified librarian, these replies are often difficult, time consuming, and in many instances, not nearly as comprehensive as they would be if such additional assistance were available to the Technical Director.

Apprentice Contest

The Society has again sponsored its Annual Apprentice Contest and for the third consecutive year the contest has included metal patternmaking as well as wood patternmaking, and molding in the fields of light metals, copper base, gray iron, and steel. A total of 268 entries were received from 94 companies and 10 Chapters. This represents an increase over last year and is nearly equal to the maximum participation in this activity in the last several years. Greater participation is probably possible by making the Chapters more conscious of the value of active participation by companies in their Chapter area.

Chapter Speakers List

As an aid to the Chapters in preparing a good technical program for the year's meetings, a new speaker list is made available each year to all Chapter representatives who attend the Chapter Officers Conference, and will be sent to others whenever requested. Each year the revision of this list is based in large part upon the Speaker Evaluation Reports received from the Chapters during the year, so as to avoid continued listing of speakers who have failed to perform acceptably at Chapter Meetings. New speakers of real merit are added when the list is being revised each year.

Attesting to the value of this list, it is noteworthy that only 21 per cent of the speakers named by the Chapter representatives as reflecting the two best they have had at their Chapter during the preceding year, were not included in the official Chapter Speaker List, as prepared by National Headquarters.

In addition, the Technical Director is attempting to arrange a series of first-class technical speakers to make presentations before several of the West Coast Chapters at some of their monthly meetings, as has been done for the past three years. This has materially aided these Chapters in obtaining good speakers on interesting subjects.

A.F.S. Film Directory

As an additional service to the Chapters we have again prepared a new and completely revised list of films covering subjects which may be of interest to the Chapters and which are known to be available.

The A.F.S. color-sound film on "Effects of Gating Design on Casting Quality," the third film resulting from research sponsored by the Society at Battelle Memorial Institute under direction of the Aluminum and Magnesium Research Committee, is available on a rental basis to our Chapters and any organization desiring it. As in the past, a rental fee of \$20 is charged for each showing, with the exception of educational institutions. This film had its premier showing at the International Foundry Con-

gress, and requests for its loan are being received at an excellent rate.

Interest in these research films has been beyond expectation, as evidenced by the fact that two European firms, one in Norway and one in Switzerland, purchased a copy of the latest film sight unseen, at a cost of \$400 each.

It is interesting to note that money expended for the production of the second film has been completely recovered through rentals received from loans.

Other Technical Activities

Close liaison with other technical societies continues, the Technical Director holding memberships in the following:

American Society of Mechanical Engineers

American Society for Metals

American Institute of Mining & Metallurgical Engineers

American Ordnance Association

American Society for Testing Materials

Society of Automotive Engineers

Society for Non-Destructive Testing Society for Advancement of Engineering Education

At the Annual Meeting of the American Society of Mechanical Engineers, the Technical Director was again elected Chairman of the Metals Engineering Division for the coming year, and also has continued as a member of the Professional Divisions Com-

mittee of ASME.

In closing, the writer wishes to express appreciation to the Board of Directors for their cooperation and understanding attitude in assisting to make our technical activities successful during the past fiscal year.

Respectfully submitted, S. C. Massari Technical Director

5-Year Comparative Membership Report

As of June 30, 1948-52, Inclusive

As of June 30, 1948-52, Inclusive						
	June 30, 1948	June 30, 1949	June 30, 1950	June 30, 1951	June 30, 1952	
Sustaining Members	199	221	197	186	191	
Company Members	1,525	1,380	1,229	1,218	1,259	
Personal Members	7,860	7,508	6,571	6,836	7,574	
Student & Apprentice Members	307	430	550	499	501	
Honorary Life Members	63	69	75	71	79	
International Members		455	424	396	429	
TOTAL		10,063	9,046	9,206	10,033	
New Members, 12 Months	1,979	1.755	1,787	2,316	1,850	
Resignations	286	303	484	319	193	
Delinquent Dropped		1,759	2,286	1,807	813	
Removed by Death	25	33	34	32	17	
Net Gain (Loss) for Year	720	(340)	(1,017)	158	827	
Members in Chapters	9.519	9,253	8,306	8,534	9,320	

Special Meeting 1951-52 Board of Directors

The Homestead, Hot Springs, Va.—Thursday, July 26, 1951

(1) ROLL CALL:

President Walter L. Seelbach, presiding Vice-President I. R. Wagner

(Terms expire 1953)

J. O. Ostergren

James Thompson

F. W. Shipley

E. C. Trov

Present:

Directors: (Terms expire 1952) T. E. Eagan L. C. Farquhar

V. J. Sedlon F. G. Sefing

W. L. Woody

(Terms expire 1954) H. W. Dietert J. T. MacKenzie M. J. O'Brien, Jr. A. M. Ondreyco Secretary-Treasurer Wm. W. Maloney

Technical Director S. C. Massari

Absent:

Directors: A. L. Hunt J. J. McFadyen L. D. Wright

(2) President Seelbach declared that this special meeting had

been called for organizing the Executive Committee in compliance with the Society's by-laws.

He announced that it was his intention as President to continue the policy of holding meetings of the Executive Committee only in cases of emergency requiring immediate decision, and that it would not be the intention to screen through the Executive Committee all subjects to be discussed in regular meetings of the Board.

The Directors voted by secret ballot to elect four members of the board to serve, together with the President and Vice-President, as an Executive Committee of six members, the 1951-52 Committee being elected as follows:

President W. L. Seelbach, Chairman; Vice-President I. R. Wagner, Director and Past-President Walton L. Woody, Director F. G. Sefing, Director F. W. Shipley, Director James Thomson. There being no further business to be considered, the meet-

ing was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTER L. SEELBACH, President

Special Meeting of 1951-52 Executive Committee

The Homestead, Hot Springs, Va.-Thursday, July 26, 1951

(1) ROLL CALL:

President Walter L. Seelbach, Presiding Vice-President I. R. Wagner

Past-President and Director Walton L. Woody

Director F. G. Sefing Director F. W. Shipley

Director James Thomson

Secretary-Treasurer Wm. W. Maloney Technical Director S. C. Massari

(2) The By-Laws of the Society (Art. X, Secs. 1-9) were read, describing the method by which the Nominating Committee shall be appointed in order to elect a Vice-President and new Directors of the Society at the 1952 Convention.

(3) Names of various candidates were presented from lists submitted by eligible Chapters. In accordance with the By-Laws, the following seven members were appointed by the Executive Committee to form, together with Past-Presidents Walton L. Woody and E. W. Horlebein, the 1951-52 Nominating Committee of nine:

Chairman, Past-President Walton L. Woody, Vice-President, National Malleable & Steel Castings Co., Cleveland.

Past-President E. W. Horlebein, President, Gibson & Kirk Co.,

Baltimore, Md.

M. E. Brooks, Fdry. Engr., Dow Chemical Co., Bay City, Mich. (Rep. Saginaw Valley Chapter and Light Metals)

A. E. Hageboeck, Exec. Vice-President, Frank Foundries Corp., Moline, Ill. (Rep. Quad City Chapter and Gray Iron)

M. E. Rollman, Engr., Cincinnati Milling Machine Co., Cincinnati. (Rep. Cincinnati District Chapter and Gray Iron) V. S. Spears, Sales Engr., American Wheelabrator & Equipment

Corp., Mishawaka, Ind. (Rep. Michiana Chapter and Equipment)

H. F. Taylor, Assoc. Professor, Massachusetts Inst. of Technology, Cambridge, Mass. (Rep. New England and Education)

F. M. Wittlinger, Secretary, Texas Electric Steel Casting Co., Houston, Texas (Rep. Texas Chapter and Steel)

R. C. Woodward, Chief. Met., Bucyrus Erie Co., South Milwaukee, Wis. (Rep. Wisconsin Chapter and Steel)

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTER L. SEELBACH, President

First Meeting of 1951-52 Board of Directors

The Homestead, Hot Springs, Va.-Friday, July 27, 1951

(1) ROLL CALL:

President Walter L. Seelbach, *Presiding* Vice-President I. R. Wagner

Present:

Directors:

(Terms expire 1952)
T. E. Eagan
L. C. Farquhar
V. J. Sedlon
F. G. Sefing
W. L. Woody

(Terms expire 1953)
J. O. Ostergren
F. W. Shipley
James Thomson
E. C. Troy

(Terms expire 1954)
H. W. Dietert
J. T. MacKenzie
M. J. O'Brien, Jr.
A. M. Ondreyco

Secretary-Treasurer Wm. W. Maloney Technical Director S. C. Massari

Absent.

Directors: A. L. Hunt J. J. McFadyen L. D. Wright

Election of Secretary-Treasurer

President Seelbach appointed a Nominating Committee consisting of Vice-President Wagner as Chairman, with Directors Woody and Ondreyco, to place in nomination the names of candidates for Secretary and Treasurer. The Committee placed in nomination the name of the incumbent Secretary-Treasurer, and on motion duly made, seconded and unanimously carried, Secretary-Treasurer Wm. W. Maloney was declared re-elected for the fiscal year 1951-52.

Fixing of Salaries and Compensations

In executive session with members of the Staff excluded, the Board of Directors established salaries and compensations for various members of the National Office Staff for the fiscal year 1951-52.

Budget of Estimated Income and Expense

The Treasurer was requested to present the budget of estimated income and expense as a two-year budget, covering the two fiscal years July 1, 1951-June 30, 1953, as recommended by the Finance Committee. President Seelbach stated that the Finance Committee was still endeavoring to build up a reserve approximately equal to an average year's expense, and that the budget had been recommended partly with this policy in mind. He also pointed out that the Finance Committee desired Board approval for setting a two-year budget for the first time, in recognition of the fact that the Society's financial affairs must consider two years at a time.

Each item of income and expense recommended was discussed in detail. On motion by Director Shipley, seconded by Director Sefing and carried, a two-year budget of estimated income and expense for the fiscal years 1951-1953 accepted, corrected in certain details by the Board, with provision for review and possible revision in July 1952.

Standing and Special Committees

At the request of the President, motion was duly made, seconded and carried, authorizing the President to appoint Standing and Special Committees in accordance with the By-Laws. President Seebach announced that the list of Committee appointments would be sent to all members of the Board at a later date.

It was suggested that an editorial in AMERICAN FOUNDRYMAN over the signature of the Chairman of the Finance Committee would be worthwhile in stressing the financial operations of the Society. It was also suggested that the Secretary should prepare talks for the Directors to present to the Chapters stressing the value of membership.

National Castings Council Representatives

The President recommended, and on motion duly made, seconded and carried, the Board approved the appointment of Vice-President Wagner to serve with President Seelbach as A.F.S. representatives to the National Castings Council.

Recommendations of 1950-51 Board of Directors

(a) Safety & Hygiene. President Seelbach introduced Kenneth M. Morse as the new full-time Staff Director of the Safety, Hygiene, and Air Pollution Program and called upon Mr. Morse to present briefly his thoughts on the need for such a program in the foundry industry. The President stated that Mr. Morse's first activities would be devoted to preparation of a brochure for the solicitation of funds, following the appointment of the full Safety & Hygiene Committee. He pointed out that establishment of bases for solicitation of funds had been delegated to the Safety & Hygiene Committee of the Board, comprising President Seelbach as Chairman, with Directors Shipley and Thomson.

(b) Headquarters Building. On motion duly made, seconded and carried, the new Board accepted the recommendations of the old Board pertaining to progression of the project to acquire and construct a new permanent headquarters for the Society.

(c) Educational Division. The new Board approved recommendations of the old Board toward arranging a 1952 Convention session of the Educational Division with Chairmen of Chapter Educational Committees. The new Board also approved recommendation that A.F.S. confine its educational activities to the secondary level of schools in 1952.

In connection with recommendation of the Educational Division that some form of contest on educational activities be initiated for the Chapters, the Board recommended that the Division endeavor to develop some definite plan to be submitted for Board approval at an early date.

(d) FEF Convention Session. The new Board accepted the recommendation of the old Board that A.F.S. should decline the request for an all-day session on education during the 1952 Convention under auspices of Foundry Educational Foundation.

1952 Convention Registration Fees

The Secretary requested special action by the Board on registration fees at the International Foundry Congress in 1952, and on motion duly made, seconded and carried, the following fees were approved: Members \$2.00, non-members \$5.00, International Visitors free, Students free, wives of International visitors free. No fee on the "free day" of Sunday, May 4.

It was the consensus that the \$2.00 and \$5.00 fees should be charged at non-exhibit as well as at exhibit Conventions.

Approval of Resolutions

The Secretary requested Board approval of the following resolutions required for conducting financial and various other affairs of the Society, and on motion duly made, seconded and carried, the resolutions were approved, as follows:

- (a) Resolved that resolutions required by the Harris Trust & Savings Bank of Chicago, authorizing the withdrawal of funds, are hereby approved and the Secretary authorized to certify thereto.
- (b) Resolved that checks for the withdrawal of funds deposited in the name of the Society with depository banks, including all General Checking accounts and Interest Savings Accounts, and for the withdrawal of all securities held in the various funds of the Society by the Trust Department of the Harris Trust & Savings Bank of Chicago, shall require the signatures of any two of the following Officers: President, Vice-President, Secretary-Treasurer, Technical Director.
- (c) RESOLVED that the Secretary be authorized to rent a safety deposit box at the Harris Trust & Savings Bank of Chicago for the safekeeping of Society securities, and that any two of the following have authority to obtain access

to such safety box: President, Vice-President, Secretary-

Treasurer, Technical Director.

(d) RESOLVED that resolutions required by the Royal Bank of Canada, Montreal, Quebec, and by its subsidiary banks at Toronto, Ontario, and Vancouver, B. C., establishing checking accounts in the name of the Society and authorizing the withdrawal of funds therefrom, are hereby approved and the Secretary authorized to certify thereto.

(e) RESOLVED that the Treasurers of the Eastern Canada, Ontario and British Columbia Chapters of the Society are authorized to make deposits, in the Royal Bank of Canada and its subsidiary banks, of funds received from Canadian members in payment solely of membership dues in the Society, and the Secretary is authorized to make all necessary accounting arrangements.

(f) RESOLVED that checks for the withdrawal of funds deposited in the name of the Society with Canadian depository banks, shall require the signatures of any two of the following Officers: President, Vice-President, Secretary-

Treasurer, Technical Director.

RESOLVED that the Treasurers of the Eastern Canada, Ontario and British Columbia Chapters of the Society shall each give evidence to the Secretary of the existence of Indemnity Bonds covering each of them individually and separately, in amounts deemed satisfactory to the Society, premiums thereon to be paid by the respective Chapters.

(h) RESOLVED that the Secretary be authorized to execute all contracts for the administration of Society affairs, subject to specific approval by the Board of Directors. In the case of A.F.S.-sponsored research projects, approval of the project by the Board of Directors includes the authority for the Secretary to execute contracts for performance of

such projects on a bid basis.

(i) RESOLVED that the Secretary be authorized to reimburse traveling expenses for members in attendance at any regularly called Board of Directors, Executive Committee, or Technical Committee meeting, with the following exceptions: No expense shall be paid to Directors or Committee members for attendance at meetings held during the week of the Annual Convention of the Society, unless specifically authorized by the Board of Directors. When meetings are held in conjunction with other committees or associations, the Secretary is authorized to determine what portion of the expense of such attendance shall be paid by the Society.

RESOLVED that the Secretary be authorized to negotiate the compensations of necessary Staff employees below \$3600.00 per year, and that Finance Committee approval be required on all compensations of \$3600.00 per year

In connection with Blanket Indemnity Bonds for the National Office Staff, the following resolution was also approved as revised below:

(k) RESOLVED that the present Blanket Indemnity Bond be renewed covering all Staff members for Five Thousand Dollars (\$5,000.00), except that Indemnity Bonds be purchased covering the Secretary-Treasurer and Technical Director for Two Hundred Thousand dollars (\$200,000.00) each and covering the bookkeepers at Twenty Thousand dollars (\$20,000.00) each, premiums to be paid by the Society.

The Treasurer, having questioned the amount of premiums required by this resolution, was directed to investigate and report back to the Finance Committee at an early date.

A.F.S. Sponsored Research Projects

Research projects of the Society as recommended by the Research Committee of the Board, Director Dunbeck, Chairman, were presented in detail. On motion duly made, seconded and

carried, the recommendations were approved for continued research in 1951-52.

1952 Chapter Officers Conference

At the request of the Secretary and on motion duly made, seconded and carried, the Board approved the holding of the 9th Annual Chapter Officers Conference in June 1952 on the same basis as the 1951 Conference.

1952 Exhibit Space Rental Fees

At the request of the Secretary for separate action, the Finance Committee recommended that exhibit space rental fees for 1952 be established as follows: Exhibit space Permit Fee, required of all exhibitors, \$25.00; Space Rentals, \$3.00 per sq. ft. It was pointed out that the increase in rental fees would in large part offset any lesser income from registration fees, established at a rate less than fees charged at the 1950 Convention.

Certificates

Director Eagan suggested that retired Directors of the Society be presented with certificates in recognition of their conscientious service and that they be made retroactive for all living Directors. On motion by Director Eagan, seconded and carried, the Board voted to prepare a suitable certificate for living past Directors in recognition of their service to the Society, and that the certificates be awarded on completion of terms of service.

Director Sefing suggested that some form of honorary certificate should be prepared for past Division Chairmen as well as for past Chairmen of General Interest Committees, and that such a certificate also be made available to the Chapters for presentation to retiring Chapter Chairmen as of 1951-52. On motion by Director Eagan, the question of honorary certificates for Chairmen of Divisions and of Chapters was tabled until the next Board meeting in order for the Staff to study the matter more

1952 Nominating Committee

For the information of the Board, President Seelbach reported that the Executive Committee, in special session the afternoon of July 26, had appointed seven members of the Society to serve with the two immediate past Presidents to form the 1952 Nominating Committee, as follows:

Chairman, Past-President Walton L. Woody, Vice-President of Operation, National Malleable & Steel Castings Co., Cleve-

Past-President E. W. Horlebein, President, Gibson & Kirk Co., Baltimore, Md.

M. E. Brooks, Fdry. Engr., Dow Chemical Co., Bay City, Mich. (Rep. Saginaw Valley Chapter and Light Metals)

A. E. Hageboeck, Exec. Vice-President, Frank Foundries Corp., Moline, Ill. (Rep. Quad City Chapter and Gray Iron)

M. E. Rollman, Engr., Cincinnati Milling Machine Co., Cincinnati. (Rep. Cincinnati District Chapter and Gray Iron)

V. S. Spears, Sales Engr., American Wheelabrator & Equip. Corp., Mishawaka, Ind. (Rep. Michiana Chapter and Equipment)

H. S. Taylor, Assoc. Professor, Massachusetts Inst. of Technology, Cambridge, Mass. (Rep. New England and Educational)

F. M. Wittlinger, Secretary, Texas Electric Steel Casting Co., Houston, Texas (Rep. Texas Chapter and Steel)
R. C. Woodward, Chief Met., Bucyrus Erie Co., South Milwau-

kee, Wis. (Rep. Wisconsin Chapter and Steel)

There being no further business to come before the Board, the meeting was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved: WALTER L. SEELBACH, President

Minutes

Meeting of the 1951-52 Nominating Committee

Stevens Hotel, Chicago-December 7, 1951

Presiding, Walton L. Woody, Chairman, as the immediate Past-President A.F.S.

E. W. Horlebein, Past-President A.F.S.

(1) ROLL CALL:

M. E. Brooks (Rep. Saginaw Valley Chapter and Light Metals)

A. E. Hageboeck (Rep. Quad City Chapter and Gray Iron)

M. E. Rollman (Rep. Cincinnati District Chapter and Gray Iron)

V. S. Spears (Rep. Michiana Chapter and Equipment)

H. S. Taylor (Rep. New England and Education)

F. M. Wittlinger (Rep. Texas Chapter and Steel)
R. C. Woodward (Rep. Wisconsin Chapter and Steel)

Nomination of President

On motion duly made, seconded and unanimously carried, the incumbent A.F.S. Vice-President, I. R. Wagner, was nominated as President of the Society for the year 1952-53.

Nomination of Vice-President

Following submission of names of candidates, on motion duly made and carried, Collins L. Carter, President of Albion Malleable Iron Co., Albion, Mich., was nominated as Vice-President of the Society for the year 1952-53.

Nomination of Directors

Following submission of names of all candidates, the Nominat-

ing Committee selected the following Directors to serve terms of three years each:

Martin A. Fladoes, President, Sivyer Steel Castings Co., Milwaukee, Wis.-representing Steel and Wisconsin Chapter.

Walter J. Klayer, Works Mgr., Aluminum Industries, Inc., Cincinnati, Ohio-representing Light Metals and Cincinnati

J. O. Klein, Vice-Pres. & Secy., Texas Foundries, Inc., Lufkin, Texas-representing Malleable Iron and Texas Chapter.

A. D. Matheson, Gen. Mgr., French & Hecht Div., Kelsey-Hayes Wheel Co., Davenport, Iowa-representing Gray Iron and Quad City Chapter.

H. G. Robertson, Works Mgr., American Steel Foundries, Alliance, Ohio-representing Steel and Canton District Chapter. Motion to nominate the above Directors was duly made, seconded, and unanimously carried.

Past-President as Director

It was pointed out that President Walter L. Seelbach, on completion of his term of office, would automatically qualify to serve one additional year as a National Director in accordance with the by-laws of the Society and without nomination.

There being no further business to be presented, the meeting was declared adjourned.

Respectfully submitted, 1951-52 NOMINATING COMMITTEE AMERICAN FOUNDRYMEN'S SOCIETY WALTON L. WOODY, Chairman

Mid-Year Meeting of the A. F. S. Board of Directors

Drake Hotel, Chicago-December 13-14, 1951

(1) ROLL CALL

President Walter L. Seelbach, presiding Vice-President I. R. Wagner

Directors:

(Terms expire 1952) (Terms expire 1953) T. E. Eagan J. J. McFadyen L. C. Farquhar F. W. Shipley V. J. Sedlon James Thomson W. L. Woody E. C. Troy

L. D. Wright

(Terms expire 1954) H. W. Dietert A. L. Hunt M. J. O'Brien, Jr. A. M. Ondrevco

Secretary-Treasurer Wm. W. Maloney Technical Director S. C. Massari

Absent Directors:

J. T. MacKenzie, J. O. Ostergren and F. G. Sefing

Reading of Minutes

A quorum having been established, the Secretary announced that minutes of the following meetings had previously been approved by letter ballot of the Board; (a) Special Meeting of the 1951-52 Board of Directors, July 26, 1951; (b) Meeting of the Executive Committee, July 26, 1951; (c) Meeting of the Board of Directors, July 27. The Secretary then read such actions and background material on these meetings as might apply to the present meeting.

Report of the Secretary

The mid-year report of the Secretary, covering the period July 1-December 10, 1951, was read and discussed, and is made a part of these minutes.

The Secretary pointed out that the Society's membership had

increased from 9206 on July 1 to 9436 on November 30, 1951, an increase of 230 members; that 12 of the 40 Chapters showed a loss of members, 3 showed neither a loss nor a gain, and 25 a gain. He again stressed the importance of concentrating on membership during the balance of the fiscal year.

The Secretary reported no new Chapters formed to date but indicated that a number of Chapters were considering or had already determined on a broadening of their sponsorship of,

Regional Conferences.

A progress report on the International Foundry Congress and Show was presented, covering the exhibits, free space for members of the National Castings Council, the mailing of housing applications, with special emphasis on sessions of particular interest to members of the Board. The latter includes the projected Defense Production sessions, the A.F.S. Annual Business Meeting and Lecture, International Committee meetings, the Society's Annual Banquet, the Canadian Dinner, International Education Dinner, Alumni Dinner, International Reception, and matters concerning registration.

In conclusion, the Secretary reported three Staff additions since last July: K. M. Morse as Director of the Safety & Hygiene and Air Pollution Program; Wm. J. Mulhall as Head Bookkeeper, and Horace Hardy as Editorial Asistant on AMERICAN

FOUNDRYMAN.

Report on American Foundryman

A more detailed report on American Foundryman advertising and editorial progress was presented by the Secretary. He stated that gross revenue for the six-month period July-December 1951. totaled \$73,900 or over \$9,000 in excess of forecast. It was stated that the December 1951 issue grossed \$13,303, or \$4,000 higher than any previous December issue.

He reported a December issue of 481/3 pages, or 12 pages more than any previous December; also a total of 272% pages for the six issues, July-December, or 53 pages greater than in any previous comparable six-month period; and that the forecast for the first six months was 278 pages, or 51% pages above the actual.

With AMERICAN FOUNDRYMAN now running close to 50 pages of advertising monthly, the Secretary stated that more total pages per issue would now have to be run so as to maintain a proper balance between editorial and advertising. At present the balance, as approved by the Publications Committee, calls for not more than 50 per cent advertising. He also stated that the entire magazine was being restyled editorially, effective March 1952.

In summing up, the Secretary stated that, barring unforeseen circumstances, the advertising budget of American Foundryman for the current fiscal year should be met. President Seelbach offered, as a consensus of the Board, congratulations to the Staff of American Foundryman for progress made.

Report of the Treasurer

The Treasurer presented a statement of Income and Expense for the five-month period July to November 1951, inclusive, showing the following: five-month forecast of Income \$238,673, actual Income, \$255,220; five-month forecast of Expense \$201,147, actual Expense \$200,805; forecast of Excess Income for the five-month period \$37,526, actual Excess Income \$54,414, or \$16,888 more than forecast.

Various items of income and expense were discussed, and in accepting the report of the Treasurer, the Board expressed satisfaction with the progress being made in the Society's finances.

Report of the Technical Director

The semi-annual report of technical activities was presented by the Technical Director covering A.F.S. sponsored Research projects, Convention program, General and Special Publications and over-all Technical Activities. The report of the Technical Director is made a part of these minutes.

The report on A.F.S.-sponsored research was supported by Director Farquhar as Chairman of the Research Committee of the Board. The Technical Director reported publication of the 1951 volume of Transactions, the high school foundry text entitled "Foundry Work," a compilation of Cupola Research reports, and progress reported on the "Foundry Sand Testing "Handbook," the "Foundry Glossary," "Recommended Practices for Casting of Copper-Base Alloys," a symposium on "Principles of Gating," a "Patternmaker's Manual," and a reprint of the "Foundry Cost Methods" booklet.

The Technical Director requested action on his previous proposal to eliminate the use of stenotype reporting at the 1952 Convention sessions, and stated that the several Technical Divisions and General Interest Committees had been sounded out on the proposal. He reported that 78 per cent of those replying had expressed themselves in favor of elimination of stenotype reporting, to be replaced by reports of representative technologists in the industry.

Motion was made by Director Eagan, seconded and carried, that an written discussion on 1952 Convention papers should be published but that oral discussion should be reported by means of technical reporters as arranged by the Technical Director. It was understood that Board approval of the proposal applied only to the 1952 Convention, subject to review thereafter.

Reports of Board Committees

(a) National Castings Council. President Seelbach and Vice-President Wagner reported on meeting of the National Castings Council November 29, with no recommendations requiring action by the A.F.S. Board. One Director asked for clarification of a general proposal for better representation of the castings industry in Washington; it appeared to be the consensus of the Board, without vote, that such representation or its consideration was beyond the province of A.F.S.

(b) Housing Committee. President Seelbach presented in digest a lengthy report from R. J. Teetor, Chairman of the A.F.S. Housing Committee, requesting stronger action in pursuing the housing project prior to expiration of the Society's present office lease on September 30, 1953. Vice-President Wagner reported on the latest meeting of the Action Subcommittee of the Housing Committee, held in Chicago November 30, called mainly to consider additional sites for the proposed Headquarters Building.

In order to advance the program more vigorously, the Secretary was authorized to act promptly for the employment of an administrative assistant, with the Finance Committee acting for

the Board of Directors in the interim. It was understood that the Secretary would seek immediate full authorization from the Housing Committee to proceed with necessary details for the taking of options, rezoning and purchase of a property.

(c) Safety, Hygiene, and Air Pollution Committee. The report of Director Shipley as Chairman of the Safety, Hygiene, and Air Pollution Committee of the Board was presented. The report covered organization of the Steering Committee, development of two technical subcommittees (on Welding, and on Foundry Dust Control and Ventilation), prospective formation of two additional committees on Air Pollution and on Foundry Safety, development of a program on safety, hygiene and air pollution for the International Foundry Congress, cooperation with governmental and professional groups, and miscellaneous activities of K. M. Morse. The report indicated that real progress has been made to date.

In the discussion that followed, the Secretary indicated that approximately \$1,000 had been contributed to date to the program, without solicitation. President Seelbach reported that, at a meeting of the National Castings Council on November 29, the members present had supported the program enthusiastically and had indicated a desire to make the original solicitation to their memberships and to those segments of the Foundry Industry which they represent. He therefore stated that, while A.F.S. would go ahead with special solicitations, it is intended to withhold general solicitation until after the original approach by the trade associations member to the National Castings Council. The new Safety, Hygiene, and Air Pollution brochure was presented to all Directors with the request that they take up at this time the question of donations by their individual companies.

On motion made, seconded and carried, the Board instructed the Secretary to express to the National Castings Council the sincere appreciation of the A.F.S. Board for the expressed willingness of the Council members to cooperate with A.F.S. toward the successful conclusion of the solicitation of funds for the program.

(d) Chapter Contacts Committee. Vice-President Wagner reported as Chairman of the Chapter Contacts Committee, and asked each Director to report on Chapters visited and still to be visited. He urged the members of the Committee to act as spark plugs for the effort to increase membership to 10,000 by next June 30, and to make their Chapter visits as early as possible.

Vice-President Wagner stated that some thought had been given to the possibility of staging the presently annual Chapter Officers Conference at less frequent intervals, and requested that all Directors evaluate the possibility and attend the next Board meeting prepared to discuss the matter fully. He questioned whether, after holding eight annual meetings, it was longer necessary to continue the conference on an annual basis.

(e) Research Committee. The report on A.F.S. sponsored research projects was supported by Director Farquhar as Chairman of the Research Committee of the Board.

In response to a question by one Director as to the status of the Cupola Research Project, it was stated that, in the opinion of the previous Board of Directors, any intention to raise new or additional funds for this project should hinge upon a full report and recommendations by the Steering Committee of the Cupola Research Project. Such a report not having been received to date, no action for additional funds is at present contemplated.

(f) Finance Committee. President Seelbach, as Chairman of the Finance Committee, stated that no separate report of his Committee would be presented.

(g) Publications Committee. The Secretary stated that no separate report of the Publications Committee was available, but digested a letter from Chairman H. M. St. John dated November 23, 1951, presenting certain recommendations from the June 29 meeting of the Publications Committee.

The Technical Director pointed out that, while the Special Publications Production budget for 1951-52 is \$10,000, in all probability the expense of productions pending would exceed that amount by an estimated \$5,000. It was understood that the Staff would obtain Finance Committee approval for the cost of any Special Publications produced during the current fiscal year beyond the present budget.

(h) Board Policy Committees. President Seelbach read a letter from Director Sefing dated December 10, and his reply of December 12, urging more progress in developing Board policies through the Board Policy Committees. After full discussion the following was agreed to: Director Woody as Chairman of the Finance Policy Committee, and Director Eagan as Chairman of the Technical Policy Committee, will hold early meetings. The Secretary was instructed to send both Chairmen in the near future a series of questions on which Board policies are most needed.

Recommendations of the Board of Awards

The following recommendations of the Board of Awards were presented for approval by the Board of Directors of the Society, in accordance with A.F.S. By-Laws:

(a) That the John A. Penton Gold Medal of A.F.S. be awarded to Frank G. Steinebach "for outstanding contributions to the Foundry Industry and the Society."

(b) That the William H. McFadden Gold Medal of A.F.S. be awarded to H. Morrogh of England "for outstanding work and development in the field of spheroidal cast iron."

(c) That the Joseph S. Seaman Gold Medal of A.F.S. be awarded to Albert Portevin of France "for technical contributions over many years to the arts and sciences of the Foundry Industry."

(d) That the Peter L. Simpson Gold Medal of A.F.S. be awarded to A. P. Gagnebin and K. D. Millis jointly "for outstanding work and development in the field of spheroidal cast iron."

Following discussion, motion was made, seconded and carried, approving the recommendations of the Board of Awards for the awarding of Gold Medals of the Society in 1952.

Recommendations of the Board of Awards for the awarding of Honorary Life Memberships in the Society then were presented for approval by the Board of Directors, as follows:

(a) That Honorary Life Membership be awarded to W. G. Mixer of Flint, Mich. "for a lifetime of service and application of engineering to the Foundry Industry."

(b) That Honorary Life Membership be awarded to A. I. Krynitsky of Washington, D. C. "for a lifetime of fundamental research in the field of Metal Casting."

(c) That Honorary Life Membership be awarded to Walter L. Seelbach on completion of his present term of office as President of A.F.S.

On motion made, seconded and carried, the Board of Directors accepted recommendations of the Board of Awards for the awarding of Honorary Life Memberships in 1952.

The procedures adopted by the Board of Awards for the presentation of awards at the 1952 Convention were approved by the Board of Directors, the Secretary being requested to convey to the Board of Awards the desirability of limiting all acceptance speeches to a maximum of three minutes each.

International Award. Rules and regulations covering the Award of Honor of the International Committee of Foundry Technical Associations were presented. The Secretary stated that the President of A.F.S. had requested the Board of Awards to recommend the 1952 recipient, for approval by the Board of Directors, which in turn would make the Society's official recommendation to the International Committee.

Recommendation of the Board of Awards was presented that A.F.S. recommend that the Award of Honor of the International Committee be made in 1952 to the late Carlo Vanzetti of Italy, and that presentation of this award be made by the President of the American Foundrymen's Society. On motion made, seconded and carried, the recommendation of the Board of Awards was approved, and the Secretary instructed to transmit the recommendation to the International Committee on behalf of the American Foundrymen's Society.

General Recommendations. On behalf of the Board of Awards, the Secretary then placed before the Board of Directors, for the purpose of obtaining their reactions, certain revisions of the Awards Manual as proposed by a committee of the Board of Awards. He emphasized the point that the revisions involved were not presented as recommendations of the Board of Awards, and that they concern two major considerations: (a) The suggestion that the presentation of names for consideration by the Board of Awards be classed as "suggestions" or "recommendations" rather than "nominations"; and (b) that steps be __en to revise the present A.F.S. By-Laws so as to eliminate the requirements that actions of the Board of Awards are subject to approval by the A.F.S. Board of Directors.

Following full discussion, and on motion duly made, seconded

and unanimously carried, the Board of Directors expressed a negative reaction to both proposals for revision of the Awards Manual. The Secretary was instructed to inform the Board of Awards of the reaction of the Board of Directors.

Recommendation of the Board of Awards was considered, suggesting the preparation of a standard type of honorarium for the Charles Edgar Hoyt Annual Lecturer. Without vote, the matter was referred to the Annual Lecture Committee for consideration, as a province of that committee.

A.F.S. Trustees to FEF

A letter from FEF to the President of A.F.S. was presented, requesting A.F.S. to name two trustees to the Foundry Educational Foundation, to take office at the annual meeting of FEF in May, 1952. The letter requested that one trustee be a present member of the A.F.S. Board of Directors, preferably also a member of the Executive Committee. On motion made, seconded and carried, President Seelbach was authorized to make the necessary appointments and inform FEF accordingly.

Vice-President of International Committee

The President stated that the Society had been honored by the International Committee of Foundry Technical Associations with an invitation to select and recommend the Vice-President of the International Committee for 1952, with the understanding that such nominee, when approved by the International Committee, would succeed to the Presidency in 1953. The necessary qualifications and considerations were pointed out, including personal attendance at the International Foundry Congress scheduled for Paris in 1953.

On motion made, seconded and unanimously carried, the Board of Directors voted to recommend to the International Committee the name of Past-President L. N. Shannon of Birmingham, Ala., for the Vice-Presidency of the International Committee in 1952. The Secretary was instructed to transmit the Society's recommendation promptly.

Vice-Chairman of International Committee on Testing Cast Iron

Fresident Scelbach stated that the International Committee on Testing Cast Iron had invited A.F.S. to make the necessary recommendation for the election of a Vice-Chairman in 1952. On motion by Director Eagan, seconded by Director Shipley and unanimously carried, the Board of Directors voted to recommend to the International Committee on Testing Cast Iron the name of H. Bornstein of Deere & Co., Moline, Ill., for the Vice-Chairmanship of said committee. The Secretary was instructed to transmit this recommendation promptly.

Chapter Invitations

President Seelbach pointed out that some eight separate non-A.F.S. foundry groups have been formed in the Eastern states without A.F.S. Chapter status, and suggested that certain of these groups might possibly consider Chapter status if officially invited by the Board of Directors to do so. He pointed out that most, if not all of these groups were in the nature of management groups rather than technical organizations, and invited the comments of the Directors. Following discussion, motion was made, seconded and carried that the Secretary should investigate the possibilities of additional A.F.S. Chapters in the Eastern states, to the end of issuing official invitations from the A.F.S. Board that such groups consider A.F.S. Chapter status.

Better Chapter Guidance

Vice-President Wagner indicated the need for closer relations with the Chapters and better guidance of their activities, particularly in such matters as membership work and the proper staging of technical meetings. He stated, however, that since the Board had authorized the Secretary to obtain additional administrative assistance in the near future, and because of the active work of the Directors in visiting Chapters, he desired to make no further recommendations at this time.

Foundry Management Meetings

Vice-President Wagner proposed the holding of a series of meetings with foundry top management in various areas so as to emphasize the importance of technical activities in foundry operations. He stated that he might call on various Directors for their assistance in organizing such meetings, and urged that the meetings be organized on a definitely informal plane, on a personal basis between the Directors and foundry management. Without vote, the Board approved the holding of such meetings.

Student Delegates

The Secretary presented a plan, tentatively concurred in by FEF, for bringing two student delegates to the 1952 International Foundry Congress from each FEF school and each A.F.S. Student Chapter school. The Secretary stated that the details of the plan were being worked out with FEF and the various schools, but requested support of the Board in carrying the plan forward.

On motion made, seconded and carried, the Board approved the student delegate plan proposed and instructed the Secretary to work out the details with the Foundry Educational Foundation for implementing at the 1952 Congress.

1954 Convention City

The Secretary requested consideration of the 1954 Convention and Exhibit city, stating that it was necessary to make such determinations well in advance in order to assure the Society of proper dates and facilities. He pointed out that two cities have the combined facilities necessary for such events, and reported recent discussions with the Cleveland Convention Bureau on the possibility of holding the 1954 Convention and Exhibit in that city. A letter from the Cleveland Convention Bureau was presented indicating the possibility of obtaining room guarantees sufficient for housing the event, and recommended favorable consideration for Cleveland in 1954.

Following discussion, on motion made, seconded and carried, the Board of Directors voted willingness to hold the 1954 Convention and Exhibit in Cleveland, if agreeable to the Committee of Exhibitors.

1955 Convention City

The Secretary presented an official letter from the Texas Chapter inviting A.F.S. to hold its non-exhibit Convention in 1955 at Houston, Texas. It was pointed out that a prior but informal invitation for 1955 had also been received from the New England Foundrymen's Association during the Buffalo Convention in 1951, and that Vice-President Wagner had held conversations since with the New England group. Without vote, it was the consensus of the Board that full consideration should be given both invitations, but that it was at this time too early to make a final decision. The Secretary was instructed to convey to the Texas Chapter the appreciation of the Society for the invitation extended.

New Student Chapter

The Secretary presented a letter and petition dated December 10 for the formation of a Student Chapter at the University of Michigan, Ann Arbor, Mich., signed by 31 students. In view of the fact that all requirements for formation of a Student Chapter

had been fulfilled, with the exception of the receipt of student member applications and the payment of the first year's dues, he recommended that the petition be provisionally accepted.

On motion made, seconded and carried, petition for formation of a new Student Chapter at the University of Michigan was declared accepted, contingent upon the fulfillment of all requirements for Student Chapter status. The Secretary was instructed to proceed accordingly.

Saugus Iron Works Memorial

The Secretary presented a letter from past-Director B. L. Simpson urging the American Foundrymen's Society to take an active part in the reconstruction of the Saugus Iron Works at Saugus, Mass., where the first casting in America was made. As an expression of the interest of the Board of Directors, motion was made, seconded and carried, instructing the Secretary to investigate the possibility of A.F.S. cooperation in the project, to report back to the Board at a later date.

European Bank Accounts

Letter from Director MacKenzie was presented, suggesting that consideration be given the setting up of A.F.S. bank accounts in Europe, thereby enabling European members of the Society to pay their dues in their own currencies, the accumulated funds to be made available to Americans traveling abroad. It was pointed out that the American Society for Metals was considering a similar step, and that A.F.S. has already set up bank accounts for the same purpose in Canada. Without vote, the Board suggested that the Secretary investigate the matter further through personal conversations and discussions during the coming International Foundry Congress.

Request of Time Study and Methods Committee

The Technical Director stated that the Time Study and Methods Committee, in meeting October 29, 1951, had requested that the Board of Directors consider authorization for the payment of travel expenses of Prof. M. E. Mundel of Purdue University to the International Foundry Congress in order to present a film on the "Use of Motion Pictures for Foundry Motion and Time Study." It was pointed out that, by resolution of the Board, the Staff was unable to reimburse the Convention expenses of individuals without specific authorization by the Board.

Without vote, it was the consensus of the Board that there should be no deviation from the existing Board resolution. However, Vice-President Wagner and Past-President Woody indicated a willingness to personally make it possible for Prof. Mundel to attend the International Congress.

There being no further business to come before the Board of Directors, the meeting was declared adjourned.

> Respectfully submitted, Wm. W. MALONEY Secretary-Treasurer

Approved: WALTER L. SEELBACH, President

Minutes Meeting of the A. F. S. Board of Directors Atlantic City, N. J.—May 4, 1952

(1) ROLL CALL:

President Walter L. Seelbach, presiding Vice-President I. R. Wagner

Directors:

(Terms expire 1952)
T. E. Eagan
L. C. Farquhar
V. J. Sedlon
F. G. Sefing
W. L. Woody
L. D. Wright

J. J. McFadyen J. O. Ostergren F. W. Shipley James Thomson E. C. Troy (Terms expire 1954)
H. W. Dietert
A. L. Hunt
J. T. MacKenzie
M. J. O'Brien, Jr.
A. M. Ondreyco
Secretary-Treasurer Wm. W. Maloney

Secretary-Treasurer Wm. W. Male Technical Director S. C. Massari Present by Invitation:

R. J. Teetor—Chairman, A.F.S. Housing Committee Incoming Officer and Directors as Observers:

Collins L. Carter, Vice-Pres.-elect.
Walter J. Klayer
J. O. Klein
A. D. Matheson
H. G. Robertson
Absent:
Martin A. Fladoes

Building Project

A quorum having been established, President Seelbach stated that, in deference to Mr. Teetor, Chairman of the A.F.S. Housing Committee, a progress report on the A.F.S. Headquarters Building project would be presented first by the Chairman. Mr. Teetor stated that an option to purchase a site in DesPlaines, Ill., had been signed and that four of the five necessary steps for rezoning the property for use by A.F.S. had been accomplished to date, with final approval by the City Council of DesPlaines to take place May 5. He stated that the projected site, selected after considerable survey of available properties by members of the Action Subcommittee and the Staff, consisted of 2½ acres of land, approximately 20 per cent of which would be occupied by the proposed building. Plans for the exterior and interior were available and discussed at some length.

The Chairman pointed out that a total of \$142,000 had been contributed (cash and pledged) for the building, but that estimates of building costs plus land costs, etc. indicated that the total contributed would be insufficient to build the contemplated structure. President Seelbach then called for discussion by the Board on the additional financing necessary.

A number of points raised by the Directors regarding building costs were answered by Chairman Teetor, President Seelbach and the Staff. Motion was then made, seconded and unanimously carried

That the Board of Directors authorizes, for completion of the A.F.S. Permanent Headquarters Building project, the expenditure of A.F.S. Reserve Funds necessary beyond those monies contributed by the industry, up to a maximum overall cost of \$200,000; it being understood that, if the maximum overall cost as stated proves insufficient, the Housing Committee must make specific request upon the Board of Directors for approval of additional funds.

At the conclusion of the housing discussion, Chairman Teetor asked to be excused and left the meeting with the expressed thanks of the Board for his work on behalf of the project.

Membership

The Secretary presented the report of the membership as of April 24, 1952, showing a total membership of 9931. He stated that, for the first time, it appeared that the Society would make its target of 10,000 members by June 30, 1952, due to the active membership work of the Chapters in the last four or five months.

The report showed a total of 2377 members added to the rolls from July 1, 1951, to April 24, 1952, inclusive, as against 1652 dropped for all causes, or a net gain of 725 members. The report also showed a total circulation of 10,669 for AMERICAN FOUNDRY-

The Secretary stated that the Society had received a petition from Omaha, Neb., signed by 67 persons desiring the establishment of a new Chapter in that area, each of the 67 agreeing to become a member in the event a Chapter be approved by the Board of Directors. He stated that a Steering Committee had been formed, of which J. M. Bruer, Paxton-Mitchell Co., Omaha, was serving as Chairman, and that in terms of petitioners and companies involved, the Secretary recommended approval of the petition by the Board. In response to questions the Secretary stated that formation of the proposed Chapter was being actively promoted by the Mo-Kan Chapter at Kansas City, Mo., and that the Timberline Chapter was not involved because of distance.

On motion made, seconded and unanimously carried, the Omaha petition for formation of the 41st regular Chapter of the Society was unanimously approved, and the Secretary instructed to notify the Steering Committee of Board action.

Vice-President Wagner and Directors Troy and Sefing reported on visits to the Conestoga, Lehigh Valley and Reading Foundrymen's groups in Pennsylvania, and Vice-President Wagner also reported on a visit with the New England Foundrymen's Association and their discussions concerning Chapters at those points. Secretary Maloney reported on cooperation with the

Pittsburgh Foundrymen's Association in connection with the International Study Tours and stated that he was visiting that Chapter on May 9 for the Study Tour Reception there.

Secretary Maloney also reported a conversation on May 3 with Henry Stenberg of Draper Corp., Hopedale, Mass., a member of the Executive Committee of the New England Foundrymen's Association, and three other New England foundrymen concerning a possible A.F.S. Chapter in New England. The Secretary had finally asked the group if they would like to receive an official invitation from the A.F.S. Board to consider Chapter status, and the group had very strongly suggested this be done. Vice-President Wagner recommended that official invitations

Vice-President Wagner recommended that official invitations be issued by the A.F.S. Board to the three Pennsylvania foundry groups and the New England foundry group to consider A.F.S. Chapter status, and on motion made, seconded and unanimously carried, the issuance of official Board invitations was approved. The Secretary also was instructed to invite one representative from each of the four groups to the Chapter Officers' Conference in Iune, at A.F.S. expense.

Director McFadyen reported on interest of a group of foundrymen in Calgary, Alberta, in holding technical meetings. It was the Board consensus that it would be difficult to support a Chapter in Calgary, but that effort should be made toward setting up a Calgary section of the British Columbia Chapter. Director McFadyen and Secretary Maloney indicated that they would cooperate with this movement as much as possible.

Finances

The Treasurer reported on finances of the Society as of March 31, presenting an Income & Expense Statement and on Operational Statement for the first nine months of the year. He reported total income of \$616,169 against the year's Income Budget of \$657,400; total Expense \$378,824 against the year's Budget of \$521,050, resulting in an earned Excess Income (over Expense) of \$237,345 against a budgeted \$136,350 for the entire fiscal year. The Treasurer presented an Operational Statement showing net Income of \$436,428 from three activities (Membership Dues \$178,994, American Foundryman \$20,561, Exhibit \$236,872), and net Expense of \$199,083.

A Balance Sheet as of March 31 was offered showing total Cash and Investments of \$479,615 (Cash on hand \$259,015, Investments \$220,600), exclusive of the Building Fund, Safety, Hygiene, and Air Pollution Fund, and Awards Fund.

The report on Advertising in American Foundryman to and including the May issue showed a total of \$165,023 gross from 606% pages of Advertising, against the previous high gross of \$115,963 from 439% pages in 1950-51. It was pointed out that the 11-month gross of \$165,000 had already exceeded the year's budget of \$162,000.

The report also showed that both the April and May issues had far exceeded any previous April and May numbers: April 1952—\$29,711 gross, May 1952—\$20,991 gross revenue; previous best April issue—\$21,766 in 1946; previous best May issue—\$12,982 in 1948. The Secretary stated that the pre-Convention issue had included 107 pages of advertising for the first time since 1946.

Safety, Hygiene, and Air Pollution

The Secretary reported a total of \$30,000 pledged to the Safety, Hygiene, and Air Pollution Fund as the result of original solicitation efforts by the trade association members of the National Castings Council. He stated that urgent business of the International Congress had prevented more concentrated Staff effort to follow up on the work of the NCC members, but that this was slated for activity at the earliest possible date following the Congress. He announced that Wm. N. Davis, formerly Senior Engineer with the National Safety Council, had been hired as the Director of the program as of May 1, 1952, following the resignation on March 1 of K. M. Morse.

President Seelbach and Vice-President Wagner reported briefly on the importance of the program and urged the Directors to follow up with their own companies toward financial support.

International Foundry Congress

Secretary Maloney and Technical Director Massari reported on various phases of the International Congress, including the pre-Congress Study Tours for International Visitors, the presentation of 11 International exchange papers, a sellout of all Exhibit space at the Foundry Show, a total of 107 technical papers at the various sessions. Various Directors commented favorably on different aspects of the Congress to date, including the highly successful International Banquet on May 3.

There being no further business to be presented, the meeting

was declared adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTER L. SEELBACH, President

Minutes Annual Meeting 1951-52 Board of Directors Edgewater Beach Hotel, Chicago—July 28, 1952

(1) ROLL CALL:

President Walter L. Seelbach, presiding Vice-President I. R. Wagner

Present:

Directors:

(Terms expire 1952) T. E. Eagan L. C. Farquhar V. J. Sedlon

L. C. Farquhar V. J. Sedlon F. G. Sefing L. D. Wright (Terms expire 1953)
J. J. McFadyen
F. W. Shipley
James Thomson
E. C. Troy

(Terms expire 1954)
H. W. Dietert
J. T. MacKenzie
M. J. O'Brien, Jr.

Secretary-Treasurer Wm. W. Maloney Technical Director S. C. Massari

Absent:

Directors: A. L. Hunt
A. M. Ondreyco
J. O. Ostergren
Walton L. Woody

Present (as observers):

Vice-President-elect Collins L. Carter Directors-elect: M. A. Fladoes

W. J. Klayer J. O. Klein A. D. Matheson A. G. Robertson

Reading of Minutes

Minutes of the meeting of the Board of Directors held May 4, 1952 at Atlantic City, N. J., were read and, on motion duly made, seconded and carried, approved.

Annual Reports of Staff Officers

(a) Report of the Secretary

The annual report of the Secretary covered the following major activities: Membership, Chapters, Convention and Exhibit, International Foundry Congress, American Foundryman, Building Project, and General Administration.

Membership.

The Secretary announced that A.F.S. membership had reached the goal of "10,000 members by June 30" with a total of 10,033, representing a net gain of 827, or approximately 8.9 per cent over the total of 9,340 on June 30, 1951.

The report showed that the membership gain was reflected in all classes. It was pointed out also that the rate of membership turnover during 1951-52 (20.4 per cent) was the lowest in four years, that the total membership gain was the largest in five years, and that the net gain was also the best in five years.

The Secretary stated that 18 Chapters made their individual membership targets during the year, against only 8 the previous year. He recommended that membership targets be continued on the basis of a goal of 11,000 total by June 30, 1953.

Chapters.

The Secretary reported the addition of one new regular Chapter (Corn Belt Chapter at Omaha, Neb.) and two new Student Chapters (University of Michigan at Ann Arbor, Mich., and Brooklyn Polytechnic Institute at Brooklyn, N. Y.) during 1951-52. A number of other Chapter possibilities were discussed, and

announcement was made that official Board invitations had been issued during the year to the New England Foundrymen's Association, the Lehigh Valley Foundrymen's Association, the Conestoga Foundrymen's Group, and the Reading Foundrymen's Association. He reported that no definite response had been received to date. Other active Chapter possibilities in which interest is being shown include Memphis, Tenn., Norfolk, Va., and a foundry group in Calgary, Alta.

The Secretary reported a successful 9th Annual Officers Conference in Chicago, June 12-13, with a total attendance of 107, including 82 Chapter delegates from 41 Chapters, plus 11 National Officers and Directors.

Student Delegate Plan.

The Secretary reported that 22 Students had attended the International Foundry Congress under the Student Delegate Plan sponsored jointly by A.F.S. and F.E.F., and that 17 had submitted essays in the contest for \$500 in prizes donated by C. B. Schneible, Chairman of F.E.F. Trustees. He announced the following winners in this contest: Ray Decker, University of Michigan; Ernest Frens, Michigan State College; Jack N. Wheeler, Missouri School of Mines; Joseph Alber, Northwestern Technological Institute; and Arthur Zrimsek, University of Wisconsin. Convention.

The Secretary reported that a total of 13,303 persons attended the International Foundry Congress & Show and 56th Annual Convention of the Society in Atlantic City, May 1-7. He reported general acceptance of the registration fees of \$2.00 for members and \$5.00 for non-members, few complaints on housing from those in attendance, and splendid cooperation from the Atlantic City Convention Bureau, Convention Hall and service organizations.

The report pointed out that the Foundry Show was a sellout with 243 paid exhibitors occupying 83,760 sq. ft., the largest exhibit since 1946. A total of 68 technical sessions, 30 Committee meetings, and 25 other events were staged during Congress Week. The International Banquet was attended by 1175 people and proved a highlight of the Congress, and the Alumni Dinner, which was open to all this year, was addressed by W. J. Grede, President of the National Association of Manufacturers.

The Secretary reported that nearly 275 foundrymen from 31 different countries had attended the International Congress and that 113 had participated in the two pre-Congress Study Tours arranged in conjunction with the Mutual Security Agency of the government. A total of 145 foundry and allied plants were made available for visits by foundrymen from abroad due to the cooperation of A.F.S. Chapters in Buffalo, Cleveland, Detroit, Chicago, Cincinnati, Philadelphia and New York, and the Pittsburgh Foundrymen's Association in Pittsburgh, Pa. The Secretary reported receipt of three congratulatory communications on the entire International Congress & Show from the National Castings Council, the Mutual Security Agency and the 1952 Committee of Exhibitors.

American Foundryman.

Reporting on American Foundryman, the Secretary stated that for the second straight year the budget for display advertising had been surpassed with a total of 664 advertising pages, a new high over the previous high of 488 pages in 1950-51.

Editorially, the Secretary pointed out that the magazine had been completely redesigned in March 1952, and that more practical articles would be sought during 1952-53.

Technical Activities.

The Secretary reported briefly the replacement of K. M. Morse by W. N. Davis as Director of the Safety, Hygiene, and Air Pollution Program, effective May 1, 1952. He stated that a total of \$42,000 had been contributed to the program to date, and that greater impetus would have to be given program financing during the coming year.

Building Project.

The Secretary announced that the projected site in DesPlaines, Ill., for a permanent Headquarters Building of A.F.S. under option to purchase since January 11, 1952, had been dropped due to the fact that the Scciety's petition for rezoning had been refused on May 5 by the DesPlaines City Council. He stated that every effort was being made by the Staff to obtain a new and satisfactory site in the immediate future.

General Administration.

The Secretary announced various changes in the Executive Staff during the fiscal year 1951-52, and the issuance of 6 issues of The BULLETIN BOARD to the Board of Directors and to Past Presidents during the year.

Following presentation of the report, Director Dietert, in commenting on the value of the Student Delegate Plan in effect at the 1952 Convention, urged adoption of the plan for the next Exhibit year and offered to contribute \$500 in prize money for the student essays. Director Sefing urged that essays for the Student Delegates be kept on general subjects, such as, "Impressions of the Foundry Congress and Exhibit."

Director Eagan expressed opposition to the Annual Lectures being of too general a nature, offering his opinion that these lectures should be technically broad in nature, that selection of lecturers should be confined to the A.F.S. membership, and that many qualified men of the industry are capable of presenting the type of lecture which he felt should be given. Motion was made, seconded and carried that rules and regulations for an Annual Lecture of a broad technical nature, with the lecturer selected from the Society's membership, be established. Action was referred to the new Board.

The Report of the Secretary, as accepted, is made a part of these Minutes.

(b) Report of the Treasurer.

The Treasurer reported that the Exhibit year 1951-52 produced a total Income of \$734,778 and Excess Income over Expense of \$182,613, both greater than for any previous 12-month period in the history of the Society. Income per member was \$73.24 against \$63.98 in the previous Exhibit year 1949-50, and expense per member of \$55.03 against \$50.66 in 1949-50.

The report indicated total value of Society Fund Principals as of June 30, 1952, was \$628,208, including all Award, Building and Safety & Hygiene Funds, against \$412,718 in 1951, \$348,302 in 1950, and \$219,590 in 1949. The Treasurer reported the Society's finances in satisfactory shape with expenses under close control of the Finance Committee and Treasurer. The official audited report of finances for the fiscal year ended June 30, as prepared by George V. Rountree in Chicago, was presented and discussed in detail, and the auditors commended for an excellent report. The Treasurer went over in detail the various items of Income and Expense for the year, pointing out major variations from the budget.

The Report of the Treasurer, on motion duly made, seconded and carried, was accepted and the Staff commended for the excellent financial showing during 1951-52. The Report of the Treasurer is made a part of these Minutes.

(c) Report of the Technical Director

The Technical Director presented a detailed report covering technical Committees, A.F.S.-sponsored Research projects, the 1952 Convention program, the Safety, Hygiene, and Air Pollution Program, Publications of the Society, the 1952 Apprentice Contests, and other technical activities.

Recommendations for continuation of Research projects during 1952-53 were tabled for consideration by the new Board.

During the year 1951-52, it was reported that seven new special publications had been produced, including Cupola Research Committee Reports, Foundry Work, (high school textbook), Symposium on Principles of Gating, Copper-Base Alloys Foundry Practices, and several existing publications reprinted. He also reported that 1500 copies of the Transactions for 1951 (Vol. 59) were published, and that complete manuscripts have been received for a Patternmakers' Manual, a Symposium on Quality Control, and a Glossary of Foundry Terms, with an Index to Transactions 1941-1951 now in process.

The Technical Director reported a total of 260 entries from 94 companies and 10 Chapters in the 1952 Apprentice Contests, compared with 246 entries from 76 companies and 8 Chapters

the previous year.

He reported that a new Chapter Speaker List and Film Directory had been developed, and that many showings of the A.F.S. color-sound film on "Effects of Gating Design on Casting Quality," the third film developed under direction of the Aluminum & Magnesium Division, had been made during 1951-52. He pointed out that money expended for production of the film had been completely recovered through rentals of the film and sale of prints both here and abroad.

The Technical Director reported that the use of recorders instead of stenotype reporters at the 1952 Convention technical sessions had been sufficiently successful to justify its continuation for another year. A considerable amount of acceptable material has been received from the recorders and will appear in Transactions.

The report of the Technical Director, on motion duly made, seconded and carried, was accepted and is made a part of these Minutes.

Reports of Board and Appointive Committees

(a) Board Policy Steering Committee

The Secretary read a letter from Walton L. Woody, as Chairman of the Board Policy Steering Committee, urging the adoption of Board policies at the earliest possible date. The Secretary then presented reports of the Finance Policy Committee and the Technical Policy Committee for consideration by the Board. Reports of these two Committees, on motion duly made, seconded and carried, were accepted, subject to approval by the new Board of Directors.

Chairman Woody as Chairman of the Finance Committee and Director Eagan as Chairman of the Technical Policy Committee tendered their resignations in consideration of new Committee appointments by President-elect I. R. Wagner.

(b) Chapter Contacts Committee

Vice-President Wagner, as Chairman of the Chapter Contacts Committee for 1951-52, presented a statement showing visits by National Officers and Directors and Staff members, and urged that the 1952-53 Committee, under leadership of Vice-President-elect Carter, continue to concentrate on membership activities in Chapter visits.

It was the consensus that A.F.S. and F.E.F. educational activities at the college level should be mutually discussed and worked out to better advantage of all concerned and with greater understanding than at present of their mutual fields of activity.

Director Sefing urged that the A.F.S. Officers should make a direct annual contact with non-A.F.S. groups in the interest of future Chapter possibilities. Director Troy requested that copies of the A.F.S. Chapter Manual be sent to non-A.F.S. foundry groups for their information. Director Dietert urged the assignment of A.F.S. Directors or members to contact colleges and universities not affiliated with F.E.F. and where no A.F.S. Student Chapter now exists.

(c) 1952 Committee on Exhibits

The Secretary presented the minutes of the meeting of the Exhibits Committee held in Chicago, June 6, 1952, at which time the group expressed unanimous agreement that the 1952 International Foundry Congress & Show in Atlantic City had been highly successful in every way, but also expressed the opinion that foundry shows in Atlantic City should not be held more frequently than every 6 to 10 years. The report discussed various items concerning the 1952 Show and offered a number of recommendations for the 1954 Exhibit, subject to approval by the 1954 Committee of Exhibitors.

(d) Finance Committee

In the absence of Chairman Woody, no report was presented, the financial report being embodied in the report of the Treasurer.

(e) Housing Committee

In the absence of Chairman R. J. Teetor, the Secretary presented a report on the A.F.S. Housing Project from its inception in 1950 to date, reporting on unsuccessful efforts to date to obtain a building site, stating that strong effort would now be made to find a proper location. He also reported that a total of \$142,924.25 had been contributed to date by member companies, individuals and Chapters. He stated that the Housing Committee was being notified of steps taken thus far and that the membership as a whole would also be kept informed.

(f) National Castings Council

Vice-President Wagner read the minutes of the annual meet-

ing of the National Castings Council held in Atlantic City, May 5, 1952, reporting on the continued interest of the Council in the A.F.S. Safety & Hygiene and Air Pollution Program and a resolution of congratulation to A.F.S. on a successful International Foundry Congress & Show.

(g) Publications Committee

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In the absence of Chairman H. M. St. John, the Secretary read the minutes of the A.F.S. Publications Committee meeting held in Chicago, July 7, 1952, detailing information on new publications during the past fiscal year and publications proposed for 1952-53. Director Eagan urged publication of a revised edition of the "Cupola Handbook" at the earliest possible date.

(h) Research Committee of the Board

Director Farquhar, as Chairman of the Research Committee of the Board, presented a report on activities during the past year and a recommended budget for the coming year. The latter was referred to the new Board for action.

(i) Retirement Plan Trustees

In the absence of Chairman B. L. Simpson, the Secretary presented a statement of receipts and disbursements for the fiscal year 1951-52 through July 31, 1952. He also read a letter from Chairman Simpson recommending that the A.F.S. Secretary-Treasurer be made a signator for the withdrawal of checks from the Retirement Trust Account along with the A.F.S. Trustees. Action deferred to the new Board.

(k) Safety, Hygiene, and Air Pollution Committee

Director Shipley, as Chairman of this Board Committee, presented a lengthy report on progress during 1951-52 and plans for the coming year. Director Eagan recommended that the report be revised for publication in American Foundryman and that a copy be placed in the hands of contributors and prospective contributors.

(5) At the conclusion of the Board meeting, President Seelbach reviewed steps leading to the present method of completely open discussion by members of the Board and urged that this method of operation be continued. He stated that he still believes the Society could and should obtain a total of 15,000 members within the next several years. He suggested that the Board Policy Committees hold their meetings in the near future, in the interest of better Board guidance, and also urged prompt completion of the Building Project. In conclusion he expressed his thanks to the Directors and to the members of the National Office Staff for their cooperation during the year.

(6) There being no further business to be presented, the Annual Meeting of the 1951-52 Board of Directors was declared

adjourned.

Respectfully submitted, WM. W. MALONEY Secretary-Treasurer

Approved:

WALTER L. SEELBACH, President

Ninth Annual Chapter Officers Conference June 12 - 13, 1952 - Sherman Hotel - Chicago

PROGRAM

	Thursday, June 12
9:30 AM	Welcome to Chapter Officers Chairman I. R. Wagner Introductions
	A.F.S. BackgroundA.F.S. Pres. W. L. Seelbach
9:52 AM	How a Technical Society Works
	Tech. Dir. S. C. Massari
10:20 AM	Program Chairman's JobConferees
12:00 PM	Luncheon
1:30 PM	Here is Your National OfficeSec'y, W. W. Maloney
2:00 рм	"The Foundrymen's Own Magazine"
	Editor H. F. Scobie
	Adv. Mgr. T. B. Koeller
2:20 PM	A.F.S. Building ProjectA.F.S. Pres. W. L. Seelbach
2:50 PM	Report on F.E.F
	(Tech. Dir. S. C. Massari
9.05 pm	Chapter Educational Programs E. M. Strick R. W. Schroeder
3:03 PM	
	(Tech. Asst. J. E. Foster
3:50 рм	Nomination of National Officers, Secy. W. W. Maloney
4:00 PM	Coming Regional ConferencesSec. W. W. Maloney

4:30 PM	"Chapter Contacts" Plan. Chairman I. R. Wagner Vice-PresElect C. L. Carter
6:30 PM	DinnerGuest Speaker: Henry Pildner
8:30 рм	Special Meetings

Friday, June 13

	Building Your MembershipConferees
10:40 am	Chapter Chairman's Job Secy. W. W. Maloney
	The New S. H. & A. P. Program
	Pres. W. L. Seelbach
	Program Dir. W. N. Davis
11:30 AM	Finances
12:00 PM	Luncheon
1:15 PM	Summing up the ConferenceSecy. W. W. Maloney
1:35 PM	The Shakeout
2:35 РМ	ResolutionsConferees

ATTENDANCE

2:45 PM Adjournment

Directors and Guests

Participating in the 9th Annual Chapter Officers Conference in addition to those mentioned, were the following national directors: Martin A. Fladoes, Sivyer Steel Casting Co., Milwaukee; J. O. Klein, Texas Foundries, Inc., Lufkin, Texas; A. D. Matheson, French & Hecht Div., Kelsey-Hayes Wheel Co., Davenport, Iowa; M. J. O'Brien, Jr., Symington-Gould Corp., Depew, N. Y.; H. G. Robertson, American Steel Foundries, Alliance, Ohio; F. W. Shipley, Caterpillar Tractor Co., Peoria, Ill.; James Thomson, Continental Foundry & Machine Co., East Chicago, Ind.; and R. E. Kennedy, A.F.S. Secretary Emeritus, University of Illinois, Chicago.

Guests included Marion Allen, Pres., Foundry Educational Foundation, Chicago; George K. Dreher, Exec. Dir., FEF; R. W. Schroeder, University of Illinois, Chicago; Earl M. Strick, Erie Malleable Iron Co., Erie, Pa.; and Wm. G. Gude, Managing Editor, Foundry, Penton Publishing Co., Cleveland.

Chapter Officers

BIRMINGHAM—Chairman Fred K. Brown, Pres., Fred K. Brown Co., Inc., Birmingham, and Program Chairman Biddle W. Worthington, Dir. of Res., McWane Cast Iron Pipe Co., Birmingham. (Also Vice-Chairman.)

BRITISH COLUMBIA—Chairman Herbert Heaton, Fdy. Sls. Engr., Letson & Burpee, Ltd., Vancouver, B. C., and Program Chairman W. R. Holeton, Asst. Prof. Met., University of British

Columbia, Vancouver, B. C. (Also Vice-Chmn.)

CANTON DISTRICT—Chairman Alexander Prentice, Mgr., Stark

Foundry Co., Canton, Ohio, and Program Chairman Robert A. Epps, Sales Rep., Stoller Chemical Co., Akron, Ohio. (Also Vice-Chairman.)

CENTRAL ILLINOIS—Chairman G. H. Rockwell, Supt. Fdy. Planning, Caterpillar Tractor Co., Peoria, Illinois, and Program Chairman Henry Felten, Plant Supt., Peoria Malleable Castings Co., Peoria, Illinois. (Also Vice-Chairman.)

CENTRAL INDIANA—Chairman Carl O. Schopp, Asst. Gen. Supt., Link-Belt Co., Indianapolis, and Treasurer C. D. Sears, Met., International Harvester Co., Indianapolis.

CENTRAL MICHIGAN-Chairman David W. Boyd, Secy., Engineering Castings, Inc., Marshall, and Vice-Chairman John Wolf,

Mgr., Midwest Foundry Co., Coldwater, Mich.

CENTRAL NEW YORK-Chairman D. J. Merwin, Vice-Pres., Oriskany Malleable Iron Co., Inc., Oriskany, N. Y., and Vice-Chairman John A. Feola, Fmn. Nonferrous Fdy., Crouse-Hinds Co., Syracuse, N. Y.

CENTRAL OHIO-Chairman E. M. Durstine, Secy.-Treas., The Keener Sand & Clay Co., Columbus, Ohio., Vice-Chairman C. W. Gilchrist, Fdy. Supt., Cooper-Bessemer Corp., Mount Vernon, Ohio, and Secretary N. H. Keyser, Rsch. Engr., Battelle Memorial Institute, Columbus, Ohio.

CHESAPEAKE-Program Chairman Wm. H. Baer, U. S. Navy Dept., Bureau of Ships, Washington, D. C. (Also Vice-Chair-

CHICAGO-Chairman John H. Owen, Dist. Sls. Mgr., Harbison-Walker Refractories Co., Chicago, and Vice-Chairman J. A. Rassenfoss, Rsch. Met., American Steel Foundries, East Chi-

CINCINNATI DISTRICT-Program Chairman W. L. Oberhelman, Secy., Oberhelman-Ritter Foundry Co., Cincinnati (Also Vice-Chairman), and Secretary Harry F. Greek, Pres., The Hill & Griffith Co., Cincinnati.

CORN BELT-Chairman Earl White, Fdy. Consultant, Paxton-Mitchell Co., Omaha, and Vice-Chairman Bert J. Baines, Fdy.

Supt., Omaha Steel Works, Omaha.

DETROIT-Chairman Michael Warchol, Met., Atlas Foundry Co., Detroit, and Vice-Chairman Harry E. Gravlin, Asst. Plant Mgr., Ford Motor Co., Dearborn, Mich.

EASTERN CANADA-Program Chairman John G. Hunt, Asst. Fdy. Supt., Dominion Engineering Works, Ltd., Montreal, Que. (Also Vice-Chmn), and Secretary A. H. Lewis, Met., Dominion Engineering Works, Ltd., Montreal, Que.

EASTERN NEW YORK-Chairman Leigh M. Townley, Asst. Met., Adirondack Foundries & Steel, Inc., Watervliet, N. Y., and Program Chairman E. S. Lawrence, Fdy. Met., General Electric Co., Schenectady, N. Y. (Also Vice-Chairman).

METROPOLITAN-Chairman J. S. Vanick, Met., International Nickel Co., Inc., New York, and Vice-Chairman Bernard N. Ames, Sr. Met., Materials Lab., U. S. Naval Shipyard, Brooklyn, N. Y.

MEXICO_Chairman Juan Latapi, Prod. Engr., Fundiciones de Hierro y Aciero, S.A., Mexico City, D. F., and Secy.-Treas. Luis Delgado-Vega, Mgr., Casco, S. De R.L., Mexico D. F., Mexico

MICHIANA-Chairman A. E. Peterson, Gen. Fmn., Oliver Corp., South Bend, Ind., Vice-Chairman Leslie Pugh, Fdy. Supt., Casting Service Corp., LaPorte, Ind., and Secretary-Treasurer V. C. Bruce, Slsmn., F. B. Stevens, Detroit.

Mo-Kan-Chairman J. F. Redman, Jr., Partner, Redman Pattern Works, Kansas City, and Vice-Chairman Wm. N. Chivvis, Plant Mgr., Magnes Metal Div., National Lead Co., Mission,

NORTHEASTERN OHIO-Chairman Frank C. Cech, Head of Patternmaking Div., Cleveland Trade School, Cleveland, and Program Chairman Stephen E. Kelly, Asst. Gen. Mgr., Eberhard Mfg. Div., Eastern Malleable Iron Co., Cleveland (also Vice-

NORTHERN CALIF.—Chairman George W. Stewart, Owner, East Bay Brass Foundry, Richmond, Program Chairman W. S. Gibbons, Partner, Ridge Foundry, San Leandro, and Harold Henderson, Chairman 1952 Regional Conference, H. C. Macaulay Foundry Co., Oakland.

No. ILL. So. Wis. Vice-Chairman Chas. N. Deubner, Fdy. Supt., Yates American Machine Co., Beloit, Wis.

NORTHWESTERN PA.—Chairman Fred J. Carlson, Asst. Fdy. Supt., Weil-McLain Co., Erie, Pa., and Program Chairman C. F. Gottschalk, Vice-Pres., Cascade Foundry Co., Erie, Pa. (Also Vice-Chairman.)

ONTARIO-Chairman Andrew Reyburn, Fdy. Supt., Cockshutt Farm Equipment, Ltd., Brantford, Ont., and Chapter Director W. A. Jones, Foreman, Canadian Westinghouse Co., Dundas.

Ont., Canada.

OREGON-Chairman Wm. M. Halverson, Indl. Engr., Electric Steel Foundry Co., Portland, and Program Chairman Henry C. Weiss, Fdy. Supt., Mobilift Corp., Portland. (Also Vice-Chairman.)

PHILADELPHIA—Chairman Arnold N. Kraft, Fdy. Mgr., Wilkening Mfg. Co., Philadelphia, and Vice-Chairman W. D. Bryden, Pres., Philadelphia Bronze & Brass Corp., Philadelphia.

QUAD CITY-Chairman Boyd Hays, Mgr., John Deere Malleable Works, East Moline, Ill., Vice-Chairman Eric Welander, Met., John Deere Malleable Works, E. Moline, Ill., and Secretary-Treasurer Robert E. Miller, Fdy. Engr., John Deere Planter Works, Moline, Ill.

ROCHESTER-Chairman Leon C. Kimpal, Indl. Engr., Rochester

Gas & Elec. Corp., Rochester, N. Y

SAGINAW VALLEY-Chairman K. H. Priestley, Pres. & Chief Met., Vassar Electroloy Products, Inc., Vassar, Mich., and Vice-Chairman F. J. McDonald, Processing Supt., Saginaw Malleable Iron Plant, Central Fdy. Div., GMC, Saginaw, Mich. St. Louis District—Chairman Henry W. Meyer, Sand Tech.,

General Steel Castings Corp., Granite City, Ill., and Program Chairman Webb L. Kammerer, Pres., Midvale Mining & Mfg.

Co., St. Louis. (Also Vice-Chairman.)

SOUTHERN CALIF.-Chairman Harold G. Pagenkopp, Partner, Angelus Pattern Works, Huntington Park, Calif., and Program Chairman Hubert Chappie, Fdy. Supt., National Supply Co., Torrance, Calif. (Also Vice-Chairman). TENNESSEE-Chairman W. M. Hamilton, Mgr., Crane Co., Chat-

tanooga, Tenn.

TEXAS-Chairman P. B. Croom, Owner, Houston Pattern Works, Houston, and Program Chairman Israel Smith, Pres., Western Foundry Co., Tyler, Texas. (Also Vice-Chairman.)

TIMBERLINE-Chairman James E. Schmuck, Supt., Rotary Steel Casting Co., Denver.

TOLEDO-Chairman John G. Blake, Supt., Alloy Founders, Inc., Toledo, Ohio, and Vice-Chairman Bernard J. Beierla, Chief Met., E. W. Bliss Co., Toledo, Ohio.

TRI-STATE—Chairman Clyde C. Beagle, Purch. Agt., The Webb Corp., Webb City, Mo., and Vice-Chairman D. W. Harris, Fdy. Supt., Frank Wheatley Pump & Valve Co., Tulsa, Okla.

Twin City-Chairman Joseph W. Costello, Pattern Supv., American Hoist & Derrick Co., St. Paul, and Vice-Chairman O. Jay Myers, Tech. Dir., Fdy. Prod. Div., Archer-Daniels-Midland Co., Minneapolis.

WASHINGTON-Chairman J. F. Dolansky, Supt., Griffin Wheel Co., So. Tacoma, Wash., and Program Chairman James N. Wessel, Materials Engr., Puget Sound Naval Shipyard, Bremerton,

Wash. (Also Vice-Chairman.)

WESTERN MICHIGAN-Chairman Wm. J. Cannon, Secy.-Treas., Nugent Sand Co., Muskegon, and Program Chairman Fred DeHudy, Met., Centrifugal Foundry Co., Muskegon Heights, Mich. (Also Vice-Chairman.)

WESTERN NEW YORK-Chairman Carl A. Harmon, Chief Met., Hanna Furnace Corp. Div. National Steel Corp., Buffalo, and Program Chairman J. M. Clifford, Works Mgr., Bison Castings, Inc., Buffalo. (Also Vice-Chairman.)

WISCONSIN—Chairman J. G. Risney, Pres., Risney Foundry Equipment Co., Wauwatosa, Wis., Program Chairman P. C. Fuerst, Asst. Fdy. Supt., The Falk Corp., Milwaukee, and Vice-Chairman A. F. Pfeiffer, Fdy. Pattern Dept., Allis-Chalmers Mfg. Co., Milwaukee.

INDUSTRY'S RESPONSIBILITY TO YOUTH

By

John S. Bugas*

One evening quite recently my 12-year-old daughter looked up from her homework and asked me: "Why do people in our country believe in Communism and Socialism? Why doesn't everybody—at least everybody in our country—believe in the way we do things in America?"

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In trying to answer her questions, I found myself putting a lot more emphasis on what is wrong with Communism and with all forms of Socialism than on what is right with our democracy. I found myself speaking defensively.

It has since occurred to me that my reaction was probably typical of a great many businessmen, and that the reaction is something less than perfect. On other questions, I would have reacted more positively.

If, for example, I wanted to convince her of the superiority of one brand of automobile over another I would suggest a simple procedure: take a long ride in each brand and you be the judge for yourself.

We know from experience that the test of performance is more convincing than sales talk. Here the automobile dealer has an advantage that we do not enjoy when we want to discuss with curious youth the advantages of our economic and political society as compared to any other in the world.

True, like the car salesman with his sales talk to the buyer, we can try through lectures and other educational devices to convince the young people of America that our system is superior, but we can not ask our youth to try out the ways of life offered in other countries and compare these systems with our own.

In fact, our bitterest rivals, the little gang of men inside the Kremlin, who claim to have the best "model" system of all, are most averse to letting anyone lift the hood and take a close look to see whether their model really works. Millions of hapless people who have been incorporated into the Russian system in the last few years can not testify to us about their new way of life for the simple reason that it is im-

possible to testify to anything from the absolute security of the Russian prison camps.

This lack of opportunity to see other systems at first hand surely means we must use our ingenuities with a new vigor and with an aggressive positive approach so that youth can satisfy itself that our own system, even with its imperfections, is the best that mankind has yet devised for us. We may change design from time to time-but the important fact here is that we are free to change as experience and education point to new requirements, and our changes never interfere with the essential basic concepts that distinguish our nation, the freedom and dignity of the individual, his right to seek happiness in his own ways, the right to ask questions and to arrive at good answers, andwhat is perhaps the main business of life-the freedom to learn and enrich our experience by the things we do.

So let me start by saying that the young people of our country should know that we think the promise of our industrial society is greater today than at any time in our history. The notion that the days of American progress are behind us is nonsense. As we get older, we may be tempted to sigh about the good old days. But down in our hearts we know, as our fathers knew before us, that this is only nostalgia. As that 12-year-old I mentioned a moment ago sometimes reminds me, "Today is yesterday tomorrow."

Promises to Our Young People

As we fill in the gap between our fathers and our children, I think that we who earn our living in the management of business and industry must keep vigorously alive three promises to the young people of our land—promises that help importantly to shape the legacy we must leave—promises expressed not soley in words but written in deeds. They are:

- We shall keep open the doors of opportunity so that the young people of tomorrow will have the chance their fathers had to use all their ability, courage, willingness to take chances.
- We shall provide conditions in which youth can move forward in their constant reaching out for something better in life.

Vice President, Industrial Relations, Ford Motor Company, Dearborn, Mich.

3. We shall set an example of citizenship by continually acting with regard to the rights of others-with an awareness of the public interest.

These promises must be written in the language of deeds. If, as an Industrial Relations man, I were to attempt a report to youth, I think I could make the case for our system of democratic capitalism. And, to keep my argument from being muddied up by abstractions, I would talk in specific terms to those of our young people who might get the idea that this country, like the old gray mare, ain't what she used to be.

For example, is industry keeping open the door of opportunity?

The answer is yes, in more ways than ever before.

Let me cite you a few examples.

Virtually all the great discoveries of resources and techniques in American economic and industrial history illustrate the individual's right to persist "in being wrong"-provided he is prepared to accept the consequences.

Examples of American Ingenuity

In 1903, as you know, a farmer's son with no money at all developed in a small shop in Detroit a flimsy machine which was later to become the foundation of one of the great industrial organizations of our time. That young man was Henry Ford. His story, although it turned out to be one of the most spectacular incidents of its kind, was not by any means unique.

The great oil fields of Kansas were discovered by a driller who was supposedly so ignorant of geological principles that he did not know the geologists had "proved"—at least to their own satisfaction—that no oil was to be found in the state of Kansas.

While Thomas Edison was persisting in his attempts to discover a filament for an electric light, internationally-known experts in the electrical field attributed his persistence to what they called his profound ignorance of electrical principles. They had already proved, through logical deductions from natural laws, that Edison could not do what he was shortly about to do.

Largely as a result of the continuing freedom of opportunity, industry expanded in this country at a rate exceeded nowhere else on earth. And as it expanded, it began to see the dangers of over-centralization-dangers which, by the way, are at the heart of Socialism and Communism.

American business and industry learned with experience that management, in large organizations, is not a function which can be held tightly by some small group at the top. No system of central planning-in industry or anywhere else-can match the initiative and resourcefulness of a great many people who are seeking constantly to work out better ideas, better methods and better products. Industry has grown more efficient and more productive because it has spread the management function widely, drawing upon as many individual centers of initiative as pos-

This decentralization of responsibility and author-

ity is a dramatic move in the direction of increasing opportunity for the individual, and probably no movement in recent years is worthy of more attention.

Industry believes strongly in the value of keeping the door of opportunity wide open. Why? Because we have found it works.

Industry An Educational Force

Let us take another example. Industry has become one of the greatest educational forces in our country today. For we have learned that opportunity alonethe mere fact that the chances for better jobs existis often not enough; often it is necessary to help a man through training to equip himself for those positions. If we were able to add together all of the money spent for training employees for better jobs, and educating them for higher responsibilities, I am sure we would discover that industry is in effect a huge university, with an enrollment that would compare respectably with the enrollment in our professional institutions of higher learning. Last year, for example, at Ford Motor Company alone more than 645,000 hours were devoted to various kinds of training pro-

No discussion of this particular point should omit giving attention to the great amount of time, energy and money spent today in the selection of individuals for the various jobs and responsibilities. We at Ford, like a great many companies, give much attention to matching abilities with opportunities. I cannot think of any other better way to insure an employee the greatest chance for taking advantage of whatever tal-

ents he may have or acquire.

Here is another instance. There was a time when communication between the heads of great business organizations and employees in the offices and shops was primitive to say the least-even non-existent. Successful communication in large organizations that have 100,000 or more people on the payroll is not easily achieved. But there is no question that industry has made great strides in its responsibility of helping employees to get a better understanding of their company and their jobs and thus get a measure of satisfaction and a sense of participation that they could not have had even ten years ago. Here again all the drive is away from the authoritarian concept toward the democratic notion that everybody counts.

I have been making the case here largely on the basis of relations between management and employees in the modern business and industrial enterprise. I think I could make as good a case by reviewing the efforts of most modern corporations toward stimulating other business enterprises to seize their opportunities-new enterprises and old, small ones and big ones.

Big Business Aids Small Business

Indeed, what may appear to be "big business" is really a vast number of small ones working together and continually creating opportunities for others with initiative and energy. More than 6,000 companies supply the Ford Motor Company, for example, and our own Purchasing department works closely with

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men in these companies, trying to help them help us. Ideas are exchanged, new methods worked out, and opportunities open to supplier organizations are thus expanded.

Why? Simply because it works best that way—for us, as well as for them.

At the other end of the process we are doing a similar job with another group of small businessmen—our 7,600 dealers, each one a small independent business in itself. The extent to which any American industrial organization of today devotes time, energy and money to the education and training of dealers would amaze most people.

I doubt whether any other section of our American society has more vigorously fought to keep open the door of opportunity and to encourage individual initiative. Industry really believes in these things and is not merely repeating pious phrases.

I said that business and industry in this country also felt a responsibility never to be satisfied with the present and that this, too, was a promise to the youth of America.

Do we need evidence on this point? The competitive drive for better products at lower costs is evident all around us. Even in these days when costs and prices have been pushed forward by powerful inflationary forces, we nevertheless find a great many products which either do not cost more, relatively speaking, or do not cost very much more—especially when we examine the improvements which have been made.

A 1952 automobile, for example, contains at least 50 items of standard equipment that in 1925 were either unknown or were extra-cost accessories. A car built in 1925 had an average lifetime mileage of 50,000 miles. Today's cars last at least twice that long.

A typical low-priced car in 1925 had a 20-hp engine, weighed 1,500 lb and was 11 ft long. Today the same make of car has a 110-hp engine, weighs about 3,200 lb and is more than 16 ft long. The increase in weight alone is due to new features which add to the comfort and durability of the car.

Industry Flexible to Change

But industry's case under this heading can be made in many other ways. It is one of the amazing characteristics of the American industrial system that management will scrap old methods without the slightest hesitation if a change promises a better product or lower costs. I find it hard to think of a man less interested in the status quo than the competitive American businessman. Why? Because he has to be that way unless he wants to "lose his shirt." The American industrialist who gives only lip service to the idea of progress soon vanishes from the scene. He can never afford to pass up a chance to substitute the improved new for the old. It is interesting to note that foreign industrialists who came to the United States a few years ago to study American methods, found our willingness to change one of the most astonishing characteristics of our technological system.

The head of one of the largest producers of start-

ing, lighting and ignition equipment outside the United States comes to visit us every few years. On his latest visit, he told us he had been half inclined not to make the trip as he did not expect there would have been enough change in our methods to make it worthwhile. But when he arrived and was shown around, he said he was amazed. "Your methods of two years ago have been almost completely scrapped," he said. He pointed to a photo mural on one of our office walls, and said, "That photo of a production operation on ignition equipment has become entirely out of date since you put it there a year or so ago. You don't do it that way any more."

Here again the promise of American industry to the oncoming generation is not words but deeds.

One of the most interesting evidences of this view-point in my opinion is the effort continually made in so many corporations to get good suggestions or new ideas both from men in the shops and from management. Last year, the Ford Motor Company received over 28,400 acceptable suggestions from eligible employees and nearly 9,000 acceptable proposals from members of management. More than \$366,000 was paid to employees for their suggestions, making a total of well over a million dollars paid out since the plan was inaugurated in 1947.

It may interest you to know that no compensation is paid out to managers for their suggestions. Both they and we consider it as a part of their responsibilities.

Industry Seeks Better Methods

But the important fact here is not the number of suggestions, or the amount paid out for them, but the insistent urge in the American industrial system to find better methods and better products and never to be satisfied with the present. On a much larger scale, we see that business was willing to invest in new ideas and new methods to the extent of 23 billion dollars laid out for new or improved plants, equipment and facilities in 1951.

Since the end of World War II, Ford Motor Company alone has spent or committed over a billion dollars for new plants and equipment.

Perhaps industry's most significant efforts in recent years have been to find better ways for people to work together. Very substantial gains have been made, I believe, in the field of labor relations. A vast amount of research has been done in the field of human relations—much of it by industry.

I have suggested that there is a third field in which industry recognizes a responsibility to youth and is fulfilling its promise by action. I refer to a responsibility to set an example of good citizenship.

Many corporations have come to the conclusion, I believe, that citizenship is a responsibility not only of the individual but of institutions. A corporation has duties and responsibilities to match the privileges it enjoys in doing business under a system of keen competition. It must act continually with an awareness of the public interest. It has a responsibility not only to produce a good product at a fair price but to be-

have in such a way that it contributes to the good of the nation and each of the communities in which it operates.

The extent to which business and industry generally accept this point of view is, I think, reflected in the increasing public esteem granted to American industry today and reflected in the public opinion polls.

Industry Is Community Minded

I believe there is every good reason why this should be so. All of the constructive activities of business and industry which I have been reviewing should contribute to public confidence. But in addition, corporations like Ford Motor Company, for example, are taking many other and more specific steps to fulfill their responsibilities of citizenship. In all of the areas where we have plants, we have set up committees to deal with our community responsibilities. We accept our proper share of the social obligations of the community.

We encourage our plant managers to take an active part in local affairs. We have a continuing plan of Open House programs so that the public in each community can see at firsthand what we are doing and how we operate. There are no "Iron Curtains"

around American industry.

Each community relations committee is authorized to ensure that Ford Motor Company carries out its full citizenship responsibilities in the communities where it operates. Here again we are not expressing merely a pious sentiment but we are acting on our beliefs and have set up organizations to deal with our

responsibilities.

There is another example I would like to mention to you. Three years ago, Ford Motor Company joined in a new plan to raise funds for 126 public service and health organizations in the Detroit area. We agreed that we would permit a company canvass of all our employees, provided all organizations would join in the single campaign so that there would be only one solicitation. Our own people, with the assistance of the union, took over the job of organizing the campaign within the company. As a result, Ford employees gave \$807,000 to these organizations as compared with \$490,000 in the preceding year. I offer this as evidence of the viewpoint of many industrial managers today.

If I were making a report to youth, I think I would offer one more piece of evidence with regard to in-

dustry's interest in corporate citizenship.

We hear a great deal today about the subject of public relations. To some people this represents an interest on the part of business and industry in communicating its own views to others. I believe that this is no more than half the truth. An interest in public relations represents not only a desire to be understood but also to understand.

Many of you will remember the story of the Indian chief who went to the bank to borrow a large sum of money and was asked for collateral—more specifically—how many horses he could put up to secure the loan. This was a new idea to the Indian chief but the rela-

tionship between money and horses gradually became clear to him. Some years later when he returned to the bank with a large sum of *money* which he wanted to deposit for safekeeping, it did not seem at all strange to him to look the banker in the eye and ask him how many *horses* he had to secure the deposit.

It seems to me now that when you and I look at the record not only of yesteday but of today and study the case for American industry and our capitalistic system, we are bound to be very much impressed by the number of horses we have. As a matter of fact, I think it is very much worth our while to take a look at those horses and to count them. Furthermore, it seems to me that when anybody proclaims the virtues of some other method—Socialism, Communism, government ownership, or what you will—we have every right to ask "How many horses have you got?"

I think it is a question that must embarrass the proponents of these reactionary systems. The devotion to the common man which we see so warmly expressed by so many systems antagonistic to democracy is a very touching devotion. But there seems to be very little pay-off. The common man receives a great deal of sympathy—which is a wonderful thing to have—but it seems there is a great gap between the

promise and the performance.

It is well to remember that the greatest contribution which has ever been made in the history of the world to the welfare of the common man was made by a thing called the capitalist system. Karl Marx himself, more than 100 years ago, paid tribute to what he called its "wonders" of production. The trouble with Marx was that he then went on to make the same mistake as the experts who thought they had proved that Edison could not do what he did. Marx thought the laws governing the capitalist system were as inflexible as the machines it used—and he made his predictions accordingly. He did not realize that the system, with all its faults, has an amazing capacity to correct those faults—simply because it is a system based upon the individual.

That is still the trouble with the Communists and Socialists who persist in following his out-moded theories today. They think, for example, that they have only to acquire machines like ours in order to equal our miracles of production. They are baffled by the simple fact that behind those machines are the efforts, brains and skill of millions of free individuals, each one spurred on by the knowledge that every miracle is achieved only to be surpassed.

If there is one modern, popular phrase which substantially describes this process, I think it may be

"Don't Fence Me In."

American Way Appeals to Courage

One thing more: perhaps as we think about these things, we should remember that our way must appeal to our courage. If we over-emphasize the importance of protection and security—if we adopt the fateful psychology of the Maginot Line, we are licked. There is no security except in taking chances, in moving ahead, in facing the problems of change, in

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trying to do new things. The Great Wall of China did not work either.

If I am asked again why people get taken in by Communism—why everybody does not believe in Democracy—I think maybe I can do a much better job of answering. I think I would say something like this:

"Democracy is a great idea. It is not an easy idea. Democracy carries with it a lot of responsibility. It is not pie in the sky or a free ride or an easy life. It is hard work. It takes gumption and courage. You have to take the hard knocks—the good with the bad. You have to have faith in people. You have to fight not only for your own rights but for the other fellow's

rights. To a lot of people this looks like a hard row to hoe—more weight than they want to carry. Things like Communism look easier.

"But the simple truth of the matter is that Democracy pays. It does more for you than any other system men have yet been able to devise. Maybe it is a reminder that there is no easy way."

I think our responsibility to youth—and industry's responsibility to youth—is to keep on promoting those beliefs among young people not simply by words but by deeds—keeping open the doors of opportunity—providing conditions in which youth of today and the men of tomorrow can move forward toward ever better things—setting an example of citizenship.

ROLE OF THE RESEARCH FOUNDRY UNIT

By

G. A. Lillieqvist*

Lower cost and improved performance, achieved through research, are the parents of the American living standard. Every improvement contributing to the comfort, ease and convenience of daily life has its origin in invention and development. There is no alternative, and this holds true for all industries.

The world's progress will depend to a great extent on developments in research laboratories. Advances in living will come through new processes, new products and new opportunities achieved through research. Without freedom all research, fundamental or applied, will falter and progress toward greater things will stop. Many words have been used to define research, to differentiate between fundamental, scientific, applied and industrial research.

It is not the intention to discuss the merits of the several kinds of research. In our industry the methods and procedures used to solve problems are not of primary concern. The end result is all important. However, it must be recognized that all applied or industrial research projects originate from fundamental or scientific research. Therefore, both types of research are of extreme importance for the continued development of new processes and new products.

In general, fundamental research is carried out by universities and various research foundations, while applied industry concentrates upon applied research. And now the question arises as to what contributions research has made to the foundry industry, and answering this question leads to our subject "The Role of the Research Foundry Unit."

Organization

The officials of the author's company visualized the necessity and the great importance of research in 1926 when they made the decision to organize a research laboratory. This laboratory was to be utilized mainly as a tool for applied research to assist in the advancement of the steel casting industry from an art to a science.

The organization was very small in the beginning, consisting of two technically trained college men and a machinist. Today, the organization has grown to 30 men, of which 11 are college men from various universities; and 19 non-technical people, such as machinists, foundry helpers, chemists and induction furnace melters. There is no doubt that the organization will continue to expand.

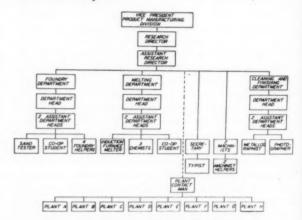


Fig. 1-Foundry research organization chart.

The present organization is illustrated in Fig. 1. It is of interest to note that the research director reports to the vice-president in charge of the manufacturing division, who reports directly to the president of the company. The organization is divided into three sections; foundry, melting, and cleaning and finishing. Each section is directed by a department head and two assistants, both of whom are university graduates either in chemical or metallurgical engineering.

Why is it necessary to have two assistants in such

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Official exchange paper from the American Foundrymen's Society to I.B.F. presented at the 1952 Annual Conference of the Institute of British Foundrymen at Buxton, England, June 10-13, 1952.

a small organization? The answer is obvious. Today in our industry there is a very real and very serious shortage of a precious national resource-researchers. To obtain more of them we must train them. To maintain leadership no phase of research can be of greater importance than the training of young re-

Another important phase of research is "How to utilize research and how the information is given to our manufacturing plants." True enough, reports are written and transmitted to the works, but in the author's opinion this is not sufficient. New methods, new processes, new applications must be discussed pro and con and, as shown in Fig. 1, the organization includes a contact man at each of the company plants with whom the various subjects can be discussed. Several of these contact men have had some of their training at the research laboratory.

The contact man in turn reports the findings to the respective works managers and, if favorably received, plant experiments are outlined. The contact man coordinates the various experiments and reports the results to the respective works manager and the research director, who in turn reports the findings to the vice-president in charge of the manufacturing division. A decision is then made as to whether or not the new process should be adopted as standard

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The duties of the department heads of the various sections are assigned as follows:

Foundry Department

- 1. New materials for patterns
- 2. Molding sand mixes
- 3. Core sand mixes
- 4. Reclamation of sand
- 5. Core blowers
- 6. New binding materials
- 7. Mold and core washes
- 8. Sand testing
- 9. Acceptance tests for various materials
- 10. Heading and gating
- 11. Use of chills, cracking strips, etc.
- 12. Pouring of molds.
- 13. Setting up quality control charts

Melting Department

- 1. Principles of induction furnace operation
- 2. Principles of electric furnace operation
- 3. Principles of open hearth furnace operation
- 4. Methods of charging
- 5. Temperature control
- 6. Deoxidizing practice
- 7. Ladle practice
- 8. Pouring temperature control
- 9. Chemical analysis of steel and other materials
- 10. Specifications
- 11. Setting up quality control charts

Cleaning Department

- 1. Removal of gates and heads
- 2. Removal of defects
- 3. Blasting

- 4. Welding
- 5. Grinding
- 6. All types of heat treatment
- 7. Physical testing
- 8. Non-destructive testing
- 9. Straightening
- 10. Investigation of casting failures
- 11. Specifications
- 12. Setting up quality control charts

The list is quite extensive, and it is, of course, humanly impossible to work on all of these subjects at any one time.

Furthermore, many problems arise in a steel casting foundry when experiments will have to be made at the plant and cannot be carried out in the laboratory; i.e., certain melting and deoxidizing practices in larger melting units; gating and heading practices for larger complicated castings; and removal of larger defects. The personnel of the research laboratory must be acquainted with these methods and procedures in order to be of assistance and make recommendations to the several plants.

The various projects are divided into four classifications. Methods of keeping records of the various experiments and of reporting results to the parties concerned are shown in the following examples:

Research Laboratory Projects

Project I-Process development

A—Steel castings

1-Melting

- 2—Foundry and core
- 3—Cleaning and finishing
- 4—Inspection

Project II-Product development

A-Steel castings

Project III—Service work for plants

A—Steel castings

Project IV—Service failure investigations

A—Steel castings

Preparation of Reports

- 1. The heading of each report will include the project number, the title of the project and the specific work being reported.
- 2. Abstract.
- 3. Object of investigation.
- 4. Summary and conclusions.
- 5. Recommendations.
- 6. Details of investigation including discussion of the results obtained.

These reports have been well received because a busy man needs only to read the abstracts in order to be informed as to what has been accomplished. Of course, the contact man and other parties especially interested in a particular field read and study the complete report.

During the fiscal year of 1951, from Sept. 1950 through Sept. 1951, 196 reports were submitted by the company's research laboratory: 109 reports or 55 per cent for process development; 55 reports or 28 per cent for product development; 24 reports or 12 per cent for service work for the various plants; and 8

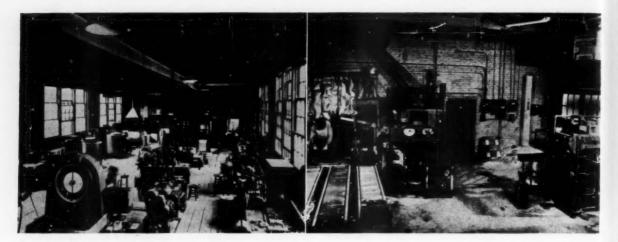


Fig. 2-One view of research foundry.

Fig. 3-Another view of research foundry.

reports or 5 per cent for service failure investigations.

The research laboratory is not operated under a fixed budget. Apparently, the company management believes that the cost of operating the research laboratory is more than offset by the benefits obtained. Additional evidence of willingness to absorb research cost is indicated by the substantial moneys recently allotted to further expand research facilities.

One other point which might be of interest, although only indirectly related to research, is the committee work within the company. Besides the various activities of societies such as the American Foundrymen's Society, the Steel Founders' Society of America, American Society for Metals, and American Institute of Mining and Metallurgical Engineers, the company officials have organized various committees within the company. Such committees are the Open Hearth Committee; Electric Furnace Committee; Pattern Standardization Committee; Core Blower Committee; and the General Inspection committee, with various sub-committees such as x-ray, magnetic powder inspection, etc. At least once a year the interested parties at the several plants meet and discuss problems of mutual interest, relate their experiences, discuss research laboratory projects and outline future experiments. This work has been of tremendous benefit to the company. Top management is very encouraged and is comtemplating additional committees.

Research Laboratory Layout

Figures 2 and 3 illustrate part of the present research laboratory layout and equipment. Figure 2 shows the main shop, in which is located two induction furnaces, heat treating furnaces, and a completely equipped machine shop for the preparation of test specimens. Figure 3 is a partial view of the experimental foundry showing a molding machine, shotblasting equipment, a burning-off machine, and a core blower.

In addition, there is a fully equipped chemical laboratory, including the latest type of spectroscope, physical testing laboratory, microscopic laboratory, and sand testing laboratory. The library includes the transactions of the important societies and the various trade magazines.

Process Development

Temperature Measurement and Fluidity

In the past few years a tremendous amount of research work has been done in the United States and Great Britain on temperature measurement and its influence on the fluidity of steel. It is impossible in this discussion to make reference to all the valuable papers which have been published, such as that of J. F. B. Jackson, Director of Research, British Steel Founders' Association Research and Development Division.¹

All foundries of the author's company—basic open hearths, acid or basic electric furnace shops—are equipped with immersion types of instruments, platinum-platinum 10 per cent rhodium, rayotube, or the photo-electronic equipment. Superintendents and melters rely on such tools to assist in producing a quality product.

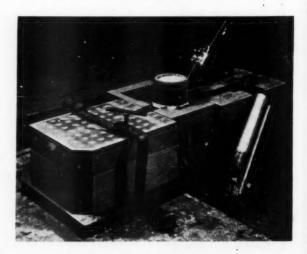


Fig. 4—Fluidity test mold assembly.

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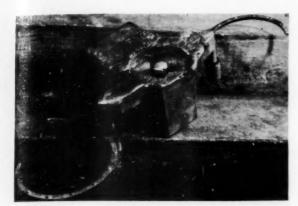


Fig. 5-Fluidity test casting.

Test procedures and findings for a research laboratory project conducted on temperature measurement and steel fluidity will be described. The fluidity test mold assembly, made of core molds, is shown in Fig. 4.

A 32 gage platinum-platinum 10 per cent rhodium thermocouple is inserted at an angle of 45 degrees into the pouring cup, and is shielded by a fused quartz tube 6 in. long, 4 mm o.p., and ½ mm wall thickness. Accurate pouring temperatures can be obtained by this method if the pouring of the steel is continuous for a period of 7 to 12 sec. The resulting fluidity test casting is shown in Fig. 5.

The photograph shows the large reservoir and how the steel enters into the downgate and hence into the spiral on both sides. This casting weighs approximately 90 lb, and the large reservoir has been provided in order to control the rate of feeding the spirals. Very satisfactory results have been obtained at various mold entering temperatures from induction furnace, acid electric furnace, and basic open hearth furnace steels of the plain carbon steel composition, as shown in Fig. 6.

The highest fluidity was obtained from the acid electric furnace steel which had the highest sulphur content, namely, 0.040 per cent. In the author's opinion, it is clearly demonstrated that, from a practical viewpoint, the mold-entering temperature is the most important factor in the control of the property "fluidity." Despite all the experiments which have

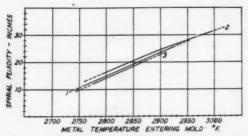


Fig. 6—Effect of melting unit on fluidity of Grade B steel.

1. Induction furnace steel; 2. Acid electric furnace steel;
3. Basic open hearth steel.

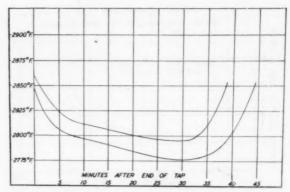


Fig. 7—Temperature of metal in ladle during pouring of two Grade B heats with tap temperatures of 2935 F.

been conducted, it is well known that the foundrymen disagree with the metallurgists. The foundrymen place emphasis on other factors, the most important of which is the rate of pouring the various molds. It may be said at this point that the proper pouring of a mold is a difficult and most important task, and should be given more study. Perhaps some method to reduce the human element may be found.

Some interesting and enlightening experiments have been made at one of the basic open hearth plants. They determined the temperature of the liquid steel in a 30-ton ladle approximately 9 in. above the nozzle from the beginning to the end of the pouring operation. The curves of Fig. 7 show that the temperature of the liquid steel increases at the end of the pour.

This gave valuable information as it was always claimed in the past that light-section castings should be poured in the beginning of the heat. Fluidity tests as described have also been made at various positions of the pouring operation and found to coincide with temperature.

Other valuable information was obtained from accurate reliable temperature measurements. Figure 8 illustrates the influence of temperature on misruns and cold shuts, and Fig. 9 the effect on sand inclusions. As expected a low temperature will result in misrun castings, and a high temperature in erosion and sand inclusions.

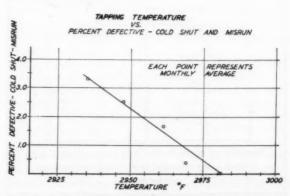


Fig. 8—Tapping temperature vs percent cold shut and misrun castings.

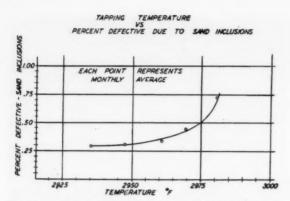


Fig. 9—Tapping temperature vs percent defective castings due to sand inclusions.

Factors Affecting Hot Tears in Steel Castings

Ever since steel castings have been made the problem of hot tearing has been encountered, and many investigations have been conducted to better understand and eliminate this defect. Many valuable papers have been written on this subject—Hall,² Middleton and Protheroe,³ and Briggs and Gezelius,^{4,5} to name a few.

The phenomenon of hot tearing in stee! castings is indeed very complex as so many factors enter into the picture, such as the chemical composition of the steel, the mold-entering temperature of the steel and, probably most important, the proper adjustment of the sand and core mixes. At the present time we have studied only the occurrence of external hot tears as we believe that they are of greater importance than internal hot tears.

After many investigations to find a suitable test casting to evaluate the sensitivity and reproducibility of hot tearing, the core assembly shown in Fig. 10

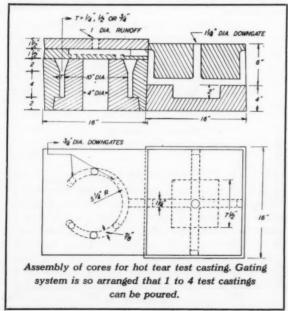


Fig. 10—Sketch of mold for casting used to evaluate the sensitivity and reproducibility of hot tearing.

was selected. This casting has a ½-in. wall thickness, and an approximate inside diameter of 10 in. Other satisfactory test castings may be designed, but for our purpose the design selected gave valuable information.

The as-poured casting weighing approximately 95 lb is shown in Fig. 11. The casting is poured through the downgate into a fairly large reservoir, the liquid steel rising into another gate and then into the ring gate, which has four small finger gates which feed the casting.

To obtain the mold entering temperature, a platinum-platinum 10 per cent rhodium thermocouple was inserted into the pouring cup in a manner similar to that in the fluidity test experiments. The pads on the outside of the casting are 2 in. wide, 1/8 in. thick, and running the entire height of the casting. These pads are for the purpose of creating a hot spot at this particular location, and are in part responsible for the occurrence of hot tears.

Now the problem of how to evaluate the severity of hot tears arises. If, after a severe hot hydrochloric acid etching test for 20 min, the casting shows no evidence of hot tears or external cracks, the hot-tear rating is designated as No. 0. If the casting shows slight cracks, as in Fig. 12, the hot tear rating is No. 1.

The hot-tear rating number is progressively increased as the cracks become more severe. In other words, a hot-tear rating standard has been established and serves as a basis of comparison for future evaluation. This method of evaluation has been extremely helpful. With very severe cracks, as shown in Fig. 13, the rating is No. 8.

The important question now arises: "What has been accomplished with this test casting?" (Fig. 11).

During the 2-yr period preceding the writing of this report, 60 progress reports have been prepared on such factors as chemical composition, deoxidation practice, pouring rate, mold-entering temperature, variation in core sand mixes, variation in grain size, grain shape and grain distribution, and many others. The effects of entering temperature and sulphur content, and sand mixtures have been selected for discussion.

It is a well-known fact that sulphur in steel increases the occurrence of hot tears, and this is shown



Fig. 11—Casting used to evaluate sensitivity and reproducibility of hot tearing.

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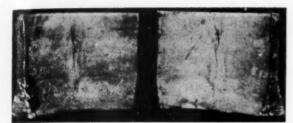


Fig. 12-Castings showing hot tear rating 1.



Fig. 13—Casting showing hot tear rating 8.

in Fig. 14. At a mold entering temperature of 2800 F and a sulphur level of 0.025 per cent, the hot-tear rating number is 0 to 1; at a sulphur level of 0.035 per cent, it is No. 1. At a level of 0.045 per cent, rating number is 3 to 4 and, finally, at the sulphur level of 0.060 per cent the rating is No. 7. It is of interest to note that at an approximate mold-entering temperature of 2950 F the hot-tear occurrence decreases very sharply, no doubt due to the fact that the cores collapse very rapidly at these high temperatures. All the experiments were made from induction furnace heats of plain medium carbon steel of the following nominal composition: C, 0.27 per cent; Mn, 0.70; Si, 0.40; P, 0.015; S, as indicated. The deoxidation practice was the same in all cases, consisting of 21/2 lb of aluminum and 3 lb of calcium-manganese-silicon per ton of charge.

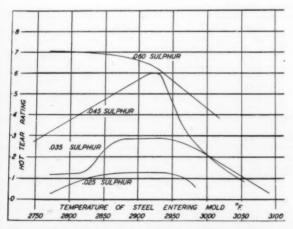


Fig. 14—Relationship between hot tear rating, steel entering temperature, and sulphur content.

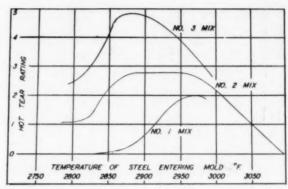


Fig. 15—Effect of three sand mixtures on hot tear occurrence in Grade B steel.

The other factor, the effect of core sand mixes, on the hot-tear occurrence is illustrated in Fig. 15. Coresand mix No. 1 has the greatest collapsibility and, consequently, the lowest hot-tear rating. Core mix No. 2 has an intermediate collapsibility and intermediate hot-tear rating, while mix No. 3, the strongest of the three mixes, has the highest hot-tear rating.

For comparative purposes core mix No. 2 was used for all subsequent work. Such metallurgical and foundry factors as are listed below were investigated:

Metallurgical Factors

- 1. Influence of mold-entering temperature.
- 2. Influence of C, Mn, Si, Cr, Ni, Mo and combinations thereof.
 - 3. Influence of phosphorus and sulphur.
- Influence of deoxidation practices with small and large amounts of aluminum, titanium, selenium, and misch metal.

Foundry Factors

- 1. Influence of grain fineness.
- 2. Influence of grain shape.
- 3. Influence of grain distribution.
- 4. Influence of various binder additions.

Use of Proper Welding Electrodes

In recent years considerable pressure has been exerted on the steel foundries to use better welding electrodes. Many customers' specifications require that the weld deposit have approximately the same chemical analysis and mechanical properties as the parent metal.

It is a difficult matter to match the chemistry of the weld deposit with the parent metal, especially in the higher carbon ranges. However, satisfactory welding electrodes have now been developed which will give approximately the same physical properties in the weld deposit as in the parent metal. This holds true for various heat-treating operations such as normalizing and tempering, quenching in liquid media and tempering. In order to evaluate various welding electrodes, a test casting was designed as shown in Fig. 16.

Tensile test bars and Charpy impact specimens are removed as indicated after the appropriate heat treatment. If the welding electrode meets the properties

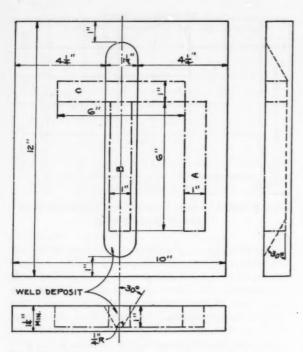


Fig. 16—Sketch of casting used to evaluate various welding electrodes.

with a sufficiently large factor of safety, it is recommended to the foundries where further experiments are carried out. Should the welding rod be satisfactory at all the plants, this rod will be adopted as standard. Only lime-ferritic or low hydrogen-containing welding electrodes are being considered, both for the D.C. and A.C. welding machines.

Incidentally, the same test casting is being used for qualification tests of individual welders. Before a welder is permitted to work on critical sections of a casting, he must produce a weld in the test casting which meets a radiographic standard as well as the required physical properties. Should he fail to meet the requirements, he will be given another test. Fail-

ure to qualify necessitates the man's removal from welding work.

Many other experiments are being conducted both in the foundries and in the research laboratory, especially with regard to the removal of gates, risers, and various foundry defects. This problem is nationwide, and the Research Committee of the Steel Founders' Society of America undertook an excellent project on this subject.⁶

Inspection

In cooperation with the foundries the research laboratory has set up radiographic and magnetic powder inspection standards for the most important castings, such as railroad side frames, couplers, yokes, brake beams and bolsters. Such standards are only applicable to long run repetitive work. The castings are divided into various zones (zone 1 is a highly stressed area, zone 2 a medium stressed area, and zone 3 a low stressed area). Zone 1 must meet the highest standard, must be entirely free of internal and external cracks, and only small amounts of lack of fusion, porosity, sand inclusions and shrinkage are permitted.

These standards have been set up by correlating data obtained through many years' experience in testing castings statically, dynamically and in service.

The research laboratory is searching for a more positive and quicker method of magnetic powder inspection of unmachined castings. The use of copper prods for magnetic powder inspection is being discouraged due to the fact that at the point of contact of the copper prod a small hardened area will be obtained which might develop into small cracks.

Other methods of inspection are being investigated, such as the magnetic coil method, fluorescent penetrant, and others. The use of dye check, etching tests and ultrasonic testing methods are continuously under consideration.

New Development

Cast Steels with Higher Physical Properties

In the past few years British research has done a great deal of work in the use of cerium in order to

TABLE 1-NORMALIZED RESULTS

	Special.	Yield.	Tensile.	Elong.,	Red, of	Charpy,	Ft. lb.	Composition, %				
Steel	Addition	psi	psi	%	Area %	Rt	_40 F	C	Mn	Si Si	P	S
1	No Rare Earth	45300	75800	30.4	52.7	35.9	11.8	.26	.70	.45	.015	.032
1	Rare Earth	45900	76100	30.6	56.3	48.0	15.8					
2	No Rare Earth	55800	93600	24.1	37.8	19.5	7.8	.43	.70	.45	.015	.032
2	Rare Earth	58000	94200	25.4	45.4	23.4	9.4					
3	No Rare Earth	63800	93400	27.9	55.3	46.8	18.1	.30	1.60	.45	.015	.032
3	Rare Earth	64100	95400	28.9	60.6	52.5	23.1					1

TABLE 2—QUENCHED AND TEMPERED RESULTS

	Special	Yield,	Tensile,	Elong., Red. of		Charpy, Ft. lb.		Composition, %				
Steel	Addition	psi	psi	%	Area %	Rt	_40 F	C	Mn	Si	P	S
1	No Rare Earth	81200	104900	18.3	46.5	33.3	25.7	.26	.70	.45	.015	.032
1	Rare Earth	81000	103900	20.2	50.8	42.5	34.7					
3	No Rare Earth	118200	126500	16.2	38.2	27.8	23.8	.30	1.60	.45	.015	.032
3	Rare Earth	117300	125800	18.2	47.1	42.5	34.8					
2	No Rare Earth	136800	154800	11.5	30.5	17.0	16.0	.43	.70	.45	.015	.032
2	Rare Earth	140800	158000	15.2	44.6	27.8	23.7					

improve the properties of cast iron, and has succeeded in developing the ductile or spheroidal cast iron. This work has been followed with great interest in the United States, and is now in limited commercial use, except that magnesium and various magnesium alloys have been substituted for cerium.

The steel casting industry became very much interested in these magnesium alloy additions for deoxidation purposes and started some experiments. It soon became apparent that these alloys due to excessive "fireworks" were too hazardous in the steel foundry. The magnesium content in the alloys had to be lowered to a maximum of 6 per cent, and no beneficial effects were obtained. Therefore, it was decided to experiment with cerium and misch metal, generally known as rare earth elements.

In Table 1 the physical properties and the Charpy impact values at room temperature and at —40 F are tabulated in the normalized condition with and without the rare earth element additions to a (1) low plain carbon steel (C, 0.26; Mn, 0.70), (2) a medium carbon steel (C, 0.40; Mn, 0.70), and (3) an intermediate manganese steel (C, 0.30; Mn, 1.60).

Ductility and Charpy impact values of these steels were improved both at room and subnormal temperatures.

Table 2 shows the same steels in the quenched and tempered condition. Again the same properties are increased to a considerable degree. The steels were made in a basic lined induction furnace, deoxidized in the regular manner with aluminum and calcium-silicon, followed with additions of 2 lb of misch metal per ton of charge. Misch metals contain 15 different rare earth elements such as cerium, lanthanum, prase-odymium, neodymium, samarium and all the other rare earth elements which are listed in the chemical periodicals.

Certain changes can be observed under the microscope which, no doubt, are partly responsible for the superior physical properties. In Fig. 17 the type of the inclusions are shown from a steel made with and without the additions of rare earth elements, photo-

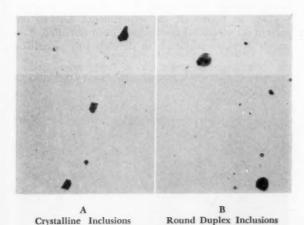
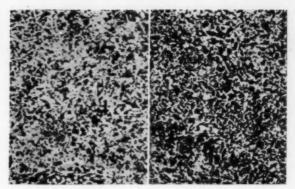


Fig. 17—Illustration of two types of inclusions. 250x.



No rare earths added.

Rare earths added.

Normalized 1.60 Mn Steel

Fig. 18—Photomicrographs of normalized steel castings.

micrograph "A" representing the steel without the special additions, and "B" with the additions.

Steel "A" contains Type III or the crystalline inclusions, while steel "B" has Type I or the round inclusions. It is well known that if cast steels have predominately Type I inclusions superior properties are obtained. However, we do not believe that this change in non-metallics is entirely responsible for the remarkable increase in the ductility and impact values.

In Fig. 18 the same steels, after a normalizing treatment, are shown in the etched condition. No significant difference man be observed, and this holds true for the normalized and quenched and tempered steels.

Further research studies were made with these steels, especially to determine their relative influence on fluidity and hot tearing. Figure 19 shows the fluidity curves at various mold-entering temperatures. The fluidity of the steel treated with rare earth elements is increased, especially at the low mold-entering temperatures.

Figure 20 shows the influence of rare earth elements on the occurrence of hot tears as determined by the specially designed test casting. The lower hot tear rating with the addition of rare earth elements probably is due in part to the change from Type III to Type I inclusions.

It is indeed a very promising and interesting field

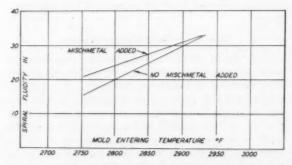


Fig. 19—Effect of mischmetal on fluidity of intermediate manganese steel.

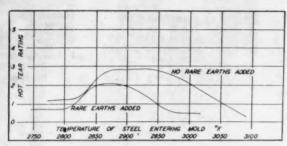


Fig. 20—Hot tear occurrence in Grade B steel containing 0.035 pct sulphur with and without rare earths added.

for further research. What changes occur to the gases in the steel with these additions? What changes occur to the sulphur in the steel? While all the reasons are not yet established, several processing applications for cast steel treated with such additives are known.

New Deoxidizers and Desulphurizers

It would be of great benefit to the steel casting industry and especially to the acid open hearth and the acid electric furnace steel producers if some alloy could be developed which would act both as an efficient deoxidizer and desulphurizer. This problem has been under consideration for many years, but no entirely satisfactory solution has been found. However, progress has been made and there is no doubt that such an alloy for the purpose intended will be commercially available to the industry in the not too distant future. Various encouraging experiments have been made with sodium additions, but due to the hazard involved in introducing it into the liquid steel more experimental work is necessary. Another alloy in various combinations which has exhibited great promise is calcium-manganese-aluminum. A typical alloy with which we had some success both in basic and acid practices had the following approximate chemical analysis: Ca, 20 per cent; Si, 45; Al, 12. A decrease of approximately 5 points in the sulphur content is being obtained with this alloy combination. Perhaps by increasing the calcium and lowering the silicon content a greater sulphur elimination can be obtained. In cooperation with the alloy producers further work is contemplated, and it is hoped that by greater sulphur reduction higher physical properties and fewer occurrences of hot tears may result.

The "C" or Shell Molding Process

Since World War II considerable experimental work has been done to develop the "C" process named after Croning of Hamburg, Germany. There is no doubt that this process has tremendous merits, not only from the standpoint of economy, but also from that of casting quality, and every effort should be made to increase the research activities on it. It has been reported that in the non-ferrous as well as in the gray iron and high alloy casting industry, considerable progress has been made.



Fig. 21-Photo of shell core.

The research laboratory has experimented with this process for approximately 21/2 years. All of the experiments were made with induction furnace heats of plain medium carbon steels, deoxidized in the standard manner with aluminum and calcium-manganese-silicon. The shell core used in the experiments is shown in Fig. 21.

This core is the same as that being used for the center core in the hot-tear experiments. No difficulties were encountered in producing the cores when approximately 6 per cent of the special phenolic resins were used. The cores were smooth and strong and had sufficient permeability.

However, after the casting had been produced sections were removed and etched in hot muriatic acid. Upon examination small defects were encountered, as illustrated in Fig. 22. These defects are small blowholes and are surrounded by steel high in carbon content as detected by a microscopic examination. This carbon "pick up" is, no doubt, due to a hydrocarbon evolution from the phenolic resin which will carburize the steel. This undesirable condition is shown in Fig. 23.

Many different phenolic resins have been tried, but in all cases the same results were obtained. It is impossible to relate all the experiments, but to the present no successful results have been obtained. The search of other materials to develop this promising



Fig. 22-Photo showing defects on surface of casting.

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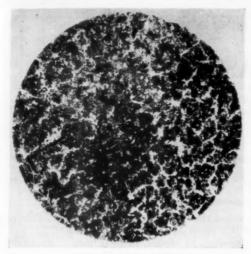


Fig. 23-Mag. 100x.

method of producing cores is continued. In this connection the author was greatly interested in the report and recommendations of Sub-Committee T.S. 30 of I.B.F. Technical Council on the subject of "Synthetic Resins in the Foundry."

During the past few years many experiments on the use of phenolic resins have been conducted in the company foundries. Few difficulties have been encountered, and approximately 75 per cent of the cores are made by using phenolic resins. It is more economical because the baking cycles have been shortened and casting quality has not suffered. It is simple to determine if a core has been underbaked as it is only necessary to drive out the moisture during the baking operation. In oil-bonded cores it is impossible to determine when the core is properly baked. Even though the moisture of the core has been eliminated, the oil must be polymerized or oxidized to a certain degree in order to obtain a satisfactory core.

Service Work for Plants

Carbon "Pick Up" from Cores

As in all steel foundries, certain difficulties suddenly arise, in many cases for unexplainable reasons. Management, foundrymen and metallurgists strive to find the cause and to correct the conditions. Occasionally a man from the research laboratory organization is delegated to determine the cause and remedy. This is termed "service work for plants."

Not long ago machining difficulties were encountered due to hard spots on the castings. The castings were made in a low carbon steel and should have been entirely satisfactory. A section containing the hard spots was removed and microscopically investigated. Surprisingly, a small layer of higher carbon content was observed directly at the surface, and this was considered as responsible for the machining dif-

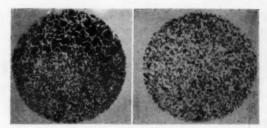


Fig. 24 (Left)—Carbon pickup, and Fig. 25 (Right)—No carbon pickup. 100x.

ficulties. This carbon "pick up" at the surface is shown in Fig. 24.

The core mix in this area was then investigated and found to be responsible for the increase in the carbon content. The core mix was changed by eliminating some of the organic binders and the undesirable condition was corrected, as shown in Fig. 25.

Service Failure Investigations

It is well known that in all foundries, and in fact in all industries, something occasionally goes wrong. In this company service failures are minimized as much as possible through better design, service testing, and the use of statistical quality control. Should a casting failure occur an investigation is welcomed as it often presents new ideas for further research work.

Acknowledgment

The author wishes to express his deepest appreciation to C. G. Mickelson, J. A. Rassenfoss and J. D. Wozny for their splendid assistance in the preparation of this paper, as well as to the officials of the American Steel Foundries for giving their permission to publish this exchange paper.

References

- 1. F. F. B. Jackson, "Temperature Measurement and Temperature Control in Steel Foundries," *Proceedings of Electric Furnace Steel Conference* (1949).
- 2. H. F. Hall, "The Strength and Ductility of Cast Steel during Cooling from the Liquid State in Sand Molds," Second Report of the Steel Castings Research Committee, British Iron and Steel Institute (1936).
- 3. J. M. Middleton and H. T. Protheroe, "The Hot Tearing of Steel," *Journal*, Iron and Steel Institute, vol. 168, August 1951.
- 4. C. W. Beiggs and B. Gezelius, "Studies on Solidification and Contraction in Steel Castings—II—Free and Hindered Contractions of Cast Carbon Steel," Transactions, American Foundrymen's Society, vol. 42, pp. 449-476 (1934).
- 5. C. W. Briggs and B. Gezelius, "Studies on Solidification and Contraction in Steel Castings—IV—Free and Hindered Contraction of Alloy Steels," Transactions, American Foundrymen's Society, vol. 44, pp. 1-32 (1936).

 6. C. W. Briggs, "Metal Removal in the Cleaning of Steel
- C. W. Briggs, "Metal Removal in the Cleaning of Steel Castings," Technical Research Committee, Steel Founders' Society of America (Sept. 1951).
- 7. "Synthetic Resins in the Foundry," Report and Recommendations of Sub-Committee T.S. 30 of the Technical Council, Foundry Trade Journal, July 5, 1951.

RISERING CASTINGS

A PROGRESS REPORT FOR 1948-1951

By

J. B. Caine*

Summarizing recent work on scientific risering, this Official Exchange Paper from A.F.S. to the 1951 International Foundry Congress applies to all metals even though the illustrations are for steel. Difficulties with the use of closely spaced risers required to meet feeding distances are discussed. Methods are advanced to overcome these difficulties and to enable the foundry to produce castings free from dispersed shrinkage with the minimum number of reasonably sized risers and an acceptable yield.

Three years ago the writer presented an exchange paper before the French Foundryman's Association on risering¹, in which he attempted to formulize and collect all available data on the science of risering. It was significant that the science of risering was less than five years old. Until that time, risering of castings was an art. A study of the earlier literature will uncover many papers and articles describing risering of individual castings and risering "tricks", but none on the fundamentals of risering.

The last three years have been even more productive and, therefore, a progress report is in order. Many of the blanks have been filled, especially those concerning riser spacing and temperature gradients required to obtain solidity in uniform sections of cast steel. In the previous paper the writer was only able to

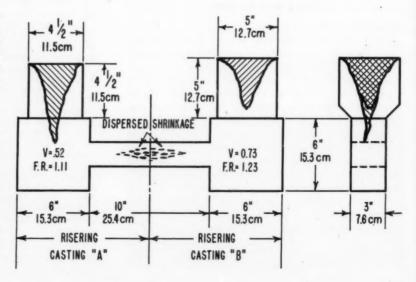
formulize riser positioning. Now, it is possible to discuss this phase of risering quantitatively.

The work of Messrs. Pellini, Bishop, Brandt, Dunphy, Myskowski and staff of the United States Naval Research Laboratory has, during the past few years, done much to advance the science of risering. This work is so outstanding as to warrant special mention. Fortunately this work is basic, so it can be applied to all types and sizes of castings. Because it is basic it cannot be applied directly to production castings but must be interpreted. The writer hopes to do some of this interpretation in this paper.

At present this work is farther advanced for steel than for other metals. It is only a question of time before similar data are forthcoming for the others.

As the paper referred to in the first paragraph has been published in France, England, and the United States, little space will be devoted to its review. The two most important points can be summarized by Fig. 1 for riser positioning, Equation 1 and Fig. 2 for riser dimensioning. It must be emphasized and reemphasized that all risering must be divided into these two phases, positioning and dimensioning. An equally important corollary is that shrinkage at the riser-casting junction must be differentiated from shrinkage in the

Fig. 1... Basic risering system showing shrinkage under the riser due to insufficient riser, and solidity with correct riser size. Since heavy sections are joined by lighter sections that freeze off and isolate the heavy sections at an early stage of solidification, each heavy section can be risered as an individual casting. Two types of shrinkage—at and away from the riser—are shown here.



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Official exchange paper, from the American Foundrymen's Society to the Belgian Foundry Technical Association, presented at the International Foundry Congress, Brussels, Belgium, Sept. 10-14, 1951 by W. S. Pellini, Naval Research Laboratory, Washington, D. C.

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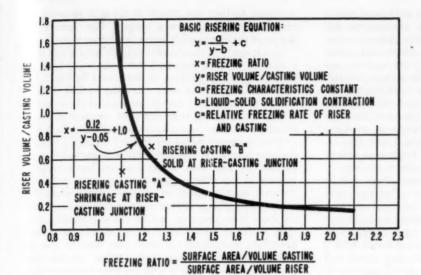


Fig. 2 . . . Basic risering equation for dimensioning risers. Curve does not include any factor of safety; factor can be included by adding whatever amount is desired to constant c in the equation. For example, a factor of 5 per cent should be enough for steel. Curve does compensate for chills. When chills are used, the surface area in contact with them is multiplied by a factor which approaches 5 as full coverage of the section is approached. This will decrease riser size.

casting away from this junction. Unless these two phases of risering and modes of shrinkage are differentiated, there will be continual confusion of thought.

Figure 1 formulizes the basic risering system, a number of relatively heavy sections joined by lighter sections that freeze off and isolate the heavy sections at an early stage of solidification. Therefore, each of the heavy sections can be risered as an individual casting, according to Fig. 2. The two types of shrinkage, at and away from the riser, are illustrated in Fig. 1. Steel has been selected for illustrative purposes and clarity. It cannot be emphasized too strongly that all metals behave the same on solidification: they contract. The difference is one of degree, not kind.

This point is not appreciated by many foundrymen, especially those in the iron and non-ferrous fields. Voids caused by solidification contraction are, in many of these alloys, dispersed to almost microscopic size. So called "open grain" in the cast iron is a typical example. True, many times this "open grain" is due to too high carbon content. Many times it is due to the same type of shrinkage shown as a definite shrinkage cavity in Fig. 1, except that the shrinkage cavity is dispersed into many millions of tiny cavities.

The reader can imagine the confusion resulting from trying to eliminate "open grain" due to solidification shrinkage by minor changes in metal composition. An actual experience may be illustrative. A certain iron foundry had been making a bushing-like casting for years, when all of a sudden the majority of castings were rejected for open grain. Frantic efforts changing the metal analysis and melting practice were unavailing. Sand and gating changes did not eliminate the trouble.

Finally, in desperation an adequate riser was used and the trouble disappeared. Why did the trouble show up after thousands of castings had been satisfactory with the old risering set up? After the reason was found the explanation was simple. The user of the castings had installed new high speed machining equipment for higher production. The smoother, almost polished machine surfaces resulting from these higher machining speeds made the small shrinkage voids vis-

ible. The older machines gave such a rough machined surface that it obscured these tiny voids.

A minor mystery that has intrigued the iron foundryman for years can possibly be explained by this concept of dispersed shrinkage. Womochel and Sigerfoos² reported years ago the difference in shrinkage characteristics of iron as influenced by the molding sand. Taylor and Schmidt³ have checked these findings and in doing so made a significant observation. The density of the iron in the sound areas of those castings showing definite shrinkage voids was higher than the density of the iron in those castings that were visually sound. As was discussed in reference 1, metals with low solidification shrinkage such as iron are quite sensitive to minor variations in rates of heat extraction and thermal gradients. Could it be that a change in rate of heat extraction and skin formation due to sand would cause dispersed shrinkage in one case, concentrated shrinkage in another?

Therefore, the reader should not dismiss the overall picture of solidification shrinkage, simply because the particular cavities illustrated in this paper for steel do not look exactly like those shown by his metals. A similar warning should be made regarding the risers shown. Top open risers are shown simply for clarity of presentation. Side blind, top blind, side open, either with or without necked down cores can be used. There is no change in fundamentals, only a difference in such details as shape and location of the shrinkage cavity or cavities (if any), change in riser size, etc.

Space does not permit a detailed discussion of Fig. 2, or details as to the derivation of Equation 1. This has been covered in reference 1.

Risering Uniform Sections

The uncertainty in Fig. 1 lies in the dispersed shrinkage in the uniform section and in riser positioning to eliminate it. This type of shrinkage seems to be common to all metals. Even steel that shows definite shrinkage voids at hot spots and junctions of sections is prone to this dispersed shrinkage in uniform sections. The approach of the steel foundryman has been to promote directional solidification, or use

closely spaced risers, not to blame it onto sand, metal analysis, gas, and other mysterious phenomena. All foundrymen would do well to consider this approach.

Tapering the section to promote directional solidification as discussed in reference 1 for steel is one method of eliminating dispersed shrinkage. Usually the tapering or padding required is prohibitively expensive, especially if the padding must be removed. Therefore, the usual approach of the steel foundryman to eliminate dispersed shrinkage in uniform sections has been closely spaced risers. Until very recently this type of riser positioning has been a matter of guesswork and has caused quite a bit of trouble.

The uniform single risered sections illustrated in Fig. 3 are simply variants of the basic risering system of Fig. 1 in which one heavy section has been removed, the other replaced and changed slightly to make it a riser. The question, unanswered until recently, is: How far can a given steel section be fed sound with a single riser? Figure 3 summarizes the work of the staff of the Naval Research Laboratory on this subject 4.5. The criterion of soundness used by the Naval Research Laboratory is radiographic soundness to 1½ per cent sensitivity of ½ and ¼ in. (1.28 and 0.65 cm) sections taken from the center of the cast section. This insures no voids in the metal whose greatest dimension exceeds 0.0075 in. (0.2 mm).

It was found that feeding distance is a function of section thickness (T). This function varies with the geometry of the section. Semicircular and circular plate-like castings were studied in ½, 1, 1½ and 2 in. (1.28 to 5.1 cm) thicknesses. Square bars investigated varied in size from 2 to 8 in. (5.1 to 20.3 cm). The dimensions of the rectangular sections are in Fig. 3.

Figure 4 summarizes the findings of the Naval Research Laboratory on the effect of gate position and multiple risers. The criterion of soundness is the same as for the single risered castings in Fig. 3. It is quite surprising that the position of the gate is not important. However, it should be emphasized that this statement may only apply to short feeding distances and closely spaced risers. If feeding distances are increased above those shown in Fig. 4, it is quite probable that gate position and metal flow will become important. Temperature gradients seem to determine solidity as far as dispersed shrinkage in uniform sections is concerned. Therefore, the position should be highly influential. This phase of risering will be discussed later.

The temperature gradients influential in Fig. 3 and 4 are those of the end of the casting and the riser. It was found that the casting end was sound for a distance of $2\frac{1}{2}$ times the section thickness $(2\frac{1}{2}T)$. The temperature gradient due to the riser contributed soundness for a distance of 2T, when the section was sound thruout. Therefore a single risered section, whether fed from the end or from the center, can be fed sound in steel for a distance of $4\frac{1}{2}T$. However, if the section is fed by two risers, one at each end of the casting section, there is no end effect and the distance between risers can only be 4T (2T for each riser).

The feeding distances shown in Fig. 3 and 4 are the maximum sizes or lengths that can be fed sound in steel. If these dimensions are exceeded, dispersed shrinkage will first be found close to the riser, as has been shown in reference 4. As the feeding distance is

increased further, the length of the area showing dispersed shrinkage becomes longer. The size of the dispersed shrinkage voids seems to increase more slowly.

May it again be emphasized that all metals will show a relation between dispersed shrinkage away from the riser and feeding distance as illustrated in Fig. 3 and 4 for steel. Perhaps in some cases the feeding distance will be much greater, but in most metal there will be a relation, differing from the steel relation in degree certainly, but just as certainly not differing in kind.

This dispersed shrinkage in steel occurring close to the riser when the feeding distance is exceeded is not affected by riser size, at least under the conditions studied so far. The volume-freezing ratio points for the individual castings in reference 4 are plotted in Fig. 5. The points differentiate shrinkage and solidity away from the riser, not at the riser-casting junction. As can be seen, there is no relation at all between shrinkage away from the riser, riser size, or the line differentiating shrinkage and solidity at the riser-casting junction. This is true for even extreme differences in riser size and freezing ratio.

For example, one casting was solid with a riser volume of only 24 per cent of the casting volume (V, 0.24; FR, 1.48). Another showed dispersed shrinkage with a riser volume 320 per cent of the casting (V, 3.20; FR, 2.76). Each group of ascending points in Fig. 5 represents a given section made with the same riser, or risers, varying the length, or feeding distance. Each group shows both solid castings and those with dispersed shrinkage regardless of its position.

It must be emphasized that the feeding distances plotted in Fig. 5 are mostly for risers with a riser contact equal to the full diameter of the riser. There are two exceptions, the semicircular plates in Fig. 3 and the casting fed vertically in Fig. 4. All the castings in these two groups were fed with risers whose volume exceeded 100 per cent of the casting volume (riser volume/casting volume over 1.0). This point may be important.

Brinson and Duma⁶ have shown that when steel castings are fed through necked down cores, the occur-

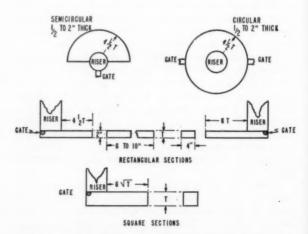


Fig. 3... Diagrams show maximum distances that can be fed with one riser without dispersed shrinkage. If these dimensions are exceeded, dispersed shrinkage will first be found close to the riser. (From Bishop & Pellini)

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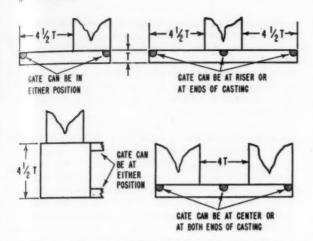


Fig. 4... Effect of gate and riser positions on feeding distances of 2 x 8-in. sections. Riser spacing much in excess results in scrap. (From Bishop & Pellini)

rence of a type of shrinkage resembling centerline dispersed shrinkage is dependent on the length of section to be fed. At a feeding distance of 1T, the opening in the necked down core between the riser and casting can be relatively small and the core thickness can exceed ½ in. (1.28 cm). An increase in feeding distance to only 2T will result in shrinkage immediately below the reduced neck between casting and riser. Increasing the feeding distance accentuates this shrinkage.

Perhaps the type of shrinkage studied in reference 6 is not related to dispersed shrinkage as discussed in this paper. Perhaps it is. If it is only remotely related to dispersed shrinkage, the wide variation in shrinkage encountered by Brinson and Duma points to contact area of the riser as being a major variable in feeding distance. This phase of risering has not been investigated. It is so important that it should be placed first on the agenda of future investigations.

If the points discussed in the previous paragraphs are accepted, they must change the thinking of the majority of steel foundrymen on two important points:

(1) Increasing the riser size with an adequate contact will eliminate or decrease dispersed shrinkage away from the riser.

(2) There is an area of sound metal close to the riser, regardless of length and size of casting. Both of these misconceptions have led to instances of extreme over-risering and very low yields.

The feeding distances of steel risers shown in Fig. 3 and 4 are disappointingly short, especially when more than one riser must be used to feed a uniform section. This brings up the question: Is maximum solidity, as determined by riser spacing as in Fig. 3 and 4, required? On highly machined, highly stressed castings such as cut tooth gears, the answer must be yes, for such castings are being made today with riser spacing approaching those of Fig. 3 and 4. Riser spacing much in excess of that shown in Fig. 4 will result in scrap castings.

However, there have been too many millions of steel castings that have performed satisfactorily in the past, fed with more widely spaced risers, to discard past risering practices. The question boils down to: What mechanical properties can be expected with various degrees of solidity as far as dispersed shrinkage is concerned? It will be many years before this question can be answered with any degree of certainty. The first point that must be settled is what tests are to be used to evaluate servicability. The test now used almost universally is the tension test. This test is probably the most sensitive test for dispersed shrinkage. An appreciable decrease in reduction of area is experienced in 1 in. (2.5 cm) steel sections when the feeding distance is as short at 3T.

There are many objections to the tension test as an evaluating test from the designer's point of view. The writer has conducted what might be called a one-man crusade about these objections ^{7,8,9} and has proven, at least to his satisfaction, that this test is meaningless as far as evaluating serviceability. This is particularly true for the elongation and reduction of area components. The two tensile properties that do have meaning in design are yield point, or elastic limit, and modulus of elasticity. Both of these tensile properties are not affected by even gross rounded discontinuities⁸ and all evidence to date shows that they are not affected by dispersed shrinkage in steel to a degree found in even unfed sections.

There are no formal data on the effect of dispersed shrinkage on notched impact strength or fatigue strength, two tests that do have meaning in evaluating servicability. There are service data. What better test for resistance to shock loading can be found than the railroad coupler? Consider the impact load that these castings must transmit when a 100-car train starts up by "taking the slack out of the couplers" and jerking the cars forward. Many thousands of these castings have performed satisfactorily at temperatures as low as -40 F. Such castings contain dispersed shrinkage and perform satisfactorily without excessive safety factor.

At least one important blank needs filling—the effect of dispersed shrinkage on fatigue strength. The only published work on this subject that the writer can find is that of Sims¹⁰ in which no difference could be found in the fatigue strength of specimens cut from forged and centrifugally cast airplane cylinder barrels. Service evidence indicates no great effect, for again there have been too many cast parts that have performed satisfactorily under high fatigue stresses. Formal proof would be advantageous.

The feeding distance ratios given in Fig. 3 and 4 are important, regardless of the values given them. Even though greater distances are used for most castings, why not rationalize these feeding distances? If 4T between multiple risers is not necessary, why not use 6T, 8T, 10T, or whatever value produces acceptable castings? Such a system will eliminate many of the production difficulties that are now thought to be mysteries. If it is safe to extrapolate the values in Fig. 3 and 4 to sections larger than 2 in. (5.1 cm), it will be found that many thick sections are now over-risered and that fewer risers than are now being used will feed these sections sound.

Temperature Gradients

Even though present evidence indicates that dispersed shrinkage is not too harmful when present in

castings subject to normal design stresses, there are classes of castings that need to be made with closely spaced risers, if for no other reason than appearance. On the other hand, there is evidence that the critical areas of steel castings designed to low factors of safety and used at tensile strengths exceeding 120,000 psi must be cast at least as sound as the castings shown in Fig. 3 and 4. The question is how to obtain this de-

gree of solidity economically.

Additional work by the Naval Research Laboratory points to a solution. The investigations of references 4 and 5 show that if the temperature gradients in plate-like sections of steel exceed 1-2 F per inch (0.22 -0.44 C per cm) no dispersed shrinkage is found. The temperature gradients of square sections need to be greater, 7-12 F per inch (1.54-2.64 C per cm) to obtain the same degree of solidity. Here is the first quantitative data as to what temperature gradients are required to obtain this much-discussed directional solidification

that is required.

This same group of investigators has also contributed information on the effect of chills on temperature gradients on solidifying cast steel11. They find that an adequate chill, one of sufficient thickness to extract heat until final solidification, will extend thermal gradients for longer distances into the casting (unpublished work by N.R.L.). If a temperature gradient of only 1-2 F per inch is required to produce sound platelike sections in steel, it should not be difficult to set up this temperature gradient for longer distances by the use of strategically placed chills. For that matter, it should be possible to extend such a small gradient by gating and proper metal flow. A gradient of 7-12 F per inch will be more difficult to achieve, but is within the possibilities of chills. An important gap in this knowledge is the temperature gradient required for solidity at junctions of sections.

Much of the basic information is now at hand to not only riser uniform sections of cast steel to the solidity required for the most severe inspection, but also to produce this solidity without the use of excessive risers. A start has been made on other metals. The thermal behavior of cast iron has been studied12 and these data can be applied to dispersed shrinkage problems in iron, just as similar work has been applied for steel.

Riser Dimensions

The use of closely spaced risers to obtain the minimum amount of dispersed shrinkage creates problems in riser dimensioning that were not discussed in reference 1. In fact, they were not even thought of at that time. Probably the greatest difficulty is the seeming inability to obtain proper feeding efficiency when risering with feeding distances approaching those shown in Fig. 3 and 4. The reason is not any mysterious change in solidification characteristics, but a decided change in the surface area-volume relation as the length decreases. This is particularly important with multiple risered castings, especially circular shapes and rings. There is no end surface area to promote gradients with these shapes. When the length dimension is decreased, the decrease in surface area through which heat is dissipated is surprisingly large.

This effect is shown in Fig. 6. In this chart the riser volume required to feed the casting, or part of the casting, is plotted against length (expressed as a function of thickness for various section ratios from 1:1 to 1:3. The riser volume is expressed as a fraction of the casting volume. Two series are plotted, one for single risered castings when the end surface areas are effective in dissipating heat, one for multiple risered castings when there is no end surface area effective insofar as heat dissipation is concerned.

The expected increase in riser size required due to increased mass of section to be fed (width times thickness) is overshadowed by the increase in riser size required as the length decreases, as it must with closely spaced risers. Quite unexpected is the sharper rate of increase as the section size decreases. Small sections need larger closely spaced risers in relation to the casting volume than do the larger sections. When the distance between risers becomes less than 4T it will be found that the only riser that will feed a ring-like section is a continuous riser.

Complicating Factor

Another complicating factor when risering casting shapes such as gear rims is the restriction in heat dissipation at the T-shaped junction of web and rim. This restriction to heat flow into the sand increases as the distance from the hub to the rim decreases, until finally the surface area of the inner surface of the rim is no longer effective and cannot be included in the surface area-volume calculations. Nothing quantitative is known at present regarding this important phase of riser dimensioning. Experience has shown that a small core surrounded by a large mass of metal extracts so

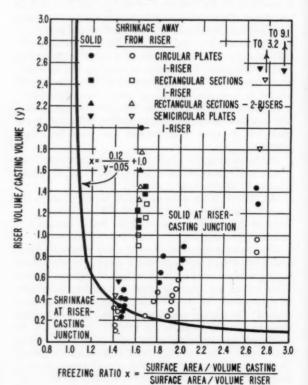


Fig. 5 . . . This graph clearly demonstrates the independence of dispersed shrinkage to riser size.

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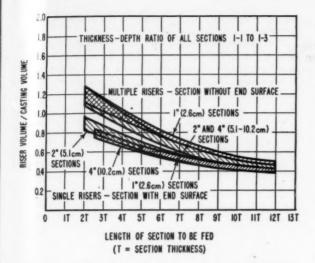


Fig. 6 . . . This graph shows the effect of end surface area on the required riser volume.

little heat that the surface area of such cores should not be included in the surface area-volume calculations.

In the case of the gear rim, only a fraction of the surface area of the inner surface of the rim will be omitted as ineffective. What this fraction should be cannot be estimated at present. However, it is important. If conditions are such that only half the inner surface is effective in heat transfer, it will be found that the riser size need be about doubled for closely spaced risers, spaced 4T to 5T apart.

Two possible solutions are available to the problem of producing sound uniform sections with a minimum number of risers and high yield. Although the following discussion applies only to steel from a quantitative standpoint, the same principles apply to all cast metals. The only reason for limiting the discussion to steel is that the constants are not available for other metals.

The first solution involves again the use of external chills. As was discussed previously, it should be possible to increase the feeding distance of risers by the proper use of chills. This automatically means fewer risers. All that is necessary for the practical application of this principle for steel is some experimental work in translating temperature gradients as determined in references 4 and 5 into feeding distances.

The use of external chills will decrease riser size in two ways. More widely spaced risers will decrease riser size according to Fig. 6. Secondly, the effective surface area of that geometric surface area in contact with the chills is increased by a factor proportional to the rate of heat extraction of a chilled surface versus a sand surface. This means for steel that the surface area in contact with the chills is multiplied by a factor which approaches 5 as full coverage of the section is approached. The effect of this increase in surface area in decreasing riser size is automatically taken care of in the surface area-volume calculations according to Fig. 2.

Chills therefore have three advantages. They permit an increase in riser spacing with fewer risers to cut off and grind. At the same time they make possible a substantial decrease in riser size due to this longer riser spacing as well as an additional decrease in riser size because of an increase in cooling rate of the casting. Further investigational work on the proper application of external chills will be very worthwhile.

Chills should be used wherever possible. If for any good reason they cannot be used, there is another method of decreasing riser size: decreasing the rate of solidification of the riser by means of insulators with a lower thermal diffusivity than sand, or addition of heat to the riser. It is probable but not proven that riser spacing will not be affected by these means.

Any change in the relative solidification rate of the riser can easily be incorporated in the basic risering equation. Constant c in this equation:

$$(1) x = \frac{a}{y-b} + c$$

measures the relative freezing rate of the casting and riser. If the casting and riser are in contact with the same sand, the relative freezing rate is unity and c=1.0. If any thing is done to decrease the freezing rate of the riser, the value for c becomes less than unity.

Curve A in Fig. 7 is drawn for steel with c=1.0 and duplicates the curves in Fig. 2 and 5. Again, steel is used for an example only because the constants a and b are known for steel. If an insulating or exothermic material is used on or around the riser so that the freezing rate of the riser is decreased 10 per cent, constant c becomes 0.90 and the curve denoting the change from shrinkage to solidity at the casting-riser junction shifts to the position of curve B. If the freezing rate of the riser can be decreased 50 per cent, the resulting curve for c=0.50 is shown by curve c.

Effect of Casting Shape

As can be seen in Fig. 7, if the casting is "rangy," when the freezing ratio is greater than 1.3, the decrease in riser size (even with a 50 per cent decrease in riser freezing rate) is not great. The lines are close together. With "chunkier" castings, when the freezing ratio is less than 1.3, the lines separate and the decrease in riser size possible by decreasing the relative freezing rate of the riser becomes quite large. No general relation can be given, for the riser size depends on the exact geometry of the casting section to be fed. The diagonal straight lines in Fig. 7 illustrate two examples, one a rangy casting made with one riser, one a chunky section requiring multiple risers.

The rangy 1 x 3 x 20 in. (2.5 x 7.5 x 50.8 cm) section requires a riser 33 per cent of the casting volume (the intersection of the diagonal straight line with curve A) for solidity at the riser-casting junction if the riser freezes at the same rate as the casting. A 10 per cent decrease in freezing rate of the riser will only allow a decrease in riser size to 28 per cent of the casting volume. A 50 per cent decrease will allow a decrease in riser volume to 20 per cent of the casting volume. It is doubtful if this small decrease will justify the time and expense of application of insulators or strongly exothermic pipe eliminators.

The preceding paragraph should not be misconstrued that the writer does not advocate the use of pipe eliminators for such castings. A prime function of a pipe eliminator is to keep the riser open to at mospheric pressure. This is necessary for all risers. The preceding remarks apply only to those materials and processes that are used primarily to reduce riser

When the casting becomes chunky, as illustrated by the 3 x 9 x 12 in. (7.5 x 22.9 x 31.2 cm) section in Fig. 7, even a slight decrease in relative freezing rate of the riser permits a significant decrease in riser size. A decrease in freezing rate of only 10 per cent allows the riser to be decreased from 100 to 78 per cent, increasing the yield from 50 to 56 per cent. If the relative freezing rate can be decreased by 50 per cent, the riser volume need be only 33 per cent and a yield of 75 per cent is possible.

It should at least be mentioned that the foregoing values and the position of the curves in Fig. 2, 5, and 7 do not include any factor of safety. Production risering curves should include some small factor of safety. This can be accomplished by adding whatever amount is decided to constant c in equation 1. A factor of safety of 5 per cent should be sufficient for steel.

There is still much to be learned. The following points are, in the writer's estimation, the most im-

portant.

1. Determination of the effect of riser contact on feeding distance.

2. Determination of constants a and b in equation 1 for metals other than steel.

3. Feeding distances in junctions of sections.

4. Determination of the restriction to heat flow into the sand at junctions of sections and its correlation to effective surface area.

5. Determination of the correlation of tempera-

ture gradients set up by external chills and feeding

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References

1. J. B. Caine, "Risering Castings," French Foundrymen's Association Transactions (1948); published in England by Enfield Rolling Mills Ltd., Enfield; A.F.S. Transactions, vol. 57, p. 66

2. H. L. Womochel and C. C. Sigerfoos, "Influence of Mold on Shrinkage in Ferrous Castings," A.F.S. Transactions, vol. 48,

p. 591 (1941)

3. H. F. Taylor and W. A. Schmidt, "Risering of Gray Iron Castings, Research Report No. 3," (to be published by A.F.S.).
4. H. F. Bishop and W. S. Pellini, "The Contribution of Riser

and Chill-Edge Effects to Soundness of Cast Steel Plates," A.F.S. Transactions, vol. 58, p. 185 (1950).
5. H. F. Bishop, E. T. Myskowski, and W. S. Pellini, "The

Contribution of Riser and Casting End Effects to the Soundness of Cast Steel Bars," A.F.S. Transactions, vol. 59, p. 171 (1951).

6. S. W. Brinson and J. A. Duma, "Observations on Knock Off Risers as Applied to Steel Castings," A.F.S. TRANSACTIONS, vol. 56, p. 586 (1948).

7. J. B. Caine, "What is Strength?" Foundry, July 1948. 8. J. B. Caine, "Radiography as a Foundry Tool," Foundry,

Jan. 1949.

9. Caine and Haught, "Slack Quenching vs Quench Cracking," Metal Progress, Sept. 1949. 10. C. E. Sims, "Comparative Properties of Cast and Forged

Steel," Foundry, May 1945.

11. H. F. Bishop, F. A. Brandt, and W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," A.F.S. Transactions, vol. 59, p. 435 (1951).

12. R. P. Dunphy and W. S. Pellini, "Solidification of Gray Iron in Sand Molds," A.F.S. Transactions, vol. 59, p. 425 (1951).

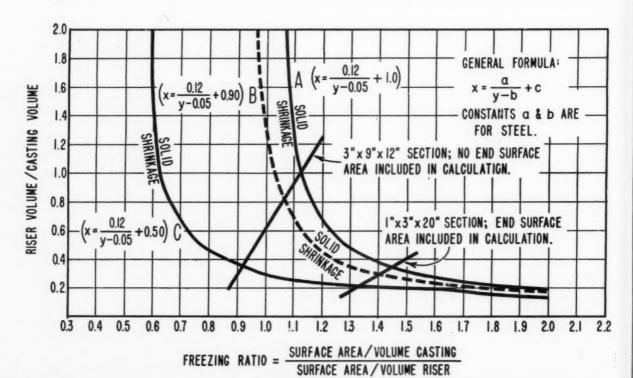


Fig. 7 . . . Effect of change in constant c on riser size. Like Fig. 2, the curves do not include any safety factor.

SAND CONTROL WITH PARTICULAR REFERENCE TO THE PREVENTION OF SCABBING

By

W. B. Parkes *

Introduction

Of the defects which may be found in castings, the scab is frequently the most difficult to associate with any particular property of the molding sand used, and is therefore the one most difficult to remove.

Most other defects which are directly due to the mold can be related to some particular property, or combination of properties, of the molding sand. The appropriate tests were originally devised by the American Foundrymen's Society and are used almost exclusively.

The tests most generally employed for routine control are those for moisture content, permeability and green and dry strength, with sometimes non-standard tests such as those for strength at high temperatures and volume of gas evolved on heating.

Provided that the green strength test is associated with a second test such as the deformation test, 1 or the more recently devised "shatter" test, 2 the results of these tests can be directly related with the behavior of the sand in the foundry—blowholes or mis-runs with high moisture content or low permeability, poor moldability with low green strength, cuts and washes with low dry or hot strength, and so on. This is not true of the scab, and most attempts to investigate this defect have had to depend on observations on test castings.

Formation of Scabs

The term scab is quite often applied to any defect on a casting which is the result of displacement of part of the sand mold by metal, and although an attempt may be made to obtain greater precision by the use of such terms as "blind scab," "dumb scab," "rammer scab" and many others, there is so little agreement as to their meaning that they are almost worthless.

The type of scab to be considered here is that fre-

quently known as an "expansion scab," and there is general agreement that this defect is the result of expansion of the surface layers of the sand mold. It may occur on almost any part of a mold, but is most frequently found where the sand is strongly heated before it is completely covered by metal at rest, i.e., on the flat top face of a mold heated by radiation from the metal while the mold is being filled, or an irregular bottom face over which there is a flow of molten metal, or on the side near the ingate.

The mode of formation of the defect is shown in

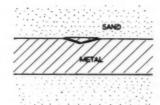


Fig. 1-Formation of a Scab

This illustrates the surface of the mold for a plain block casting. As soon as the bottom of the mold is covered by molten metal, the top is heated by radiation and a thin surface layer dries out. The dried sand then expands rapidly as its temperature rises. Since the green sand behind the expanding layer remains plastic, the latter has some degree of freedom of movement, and the strain can be concentrated in relatively few positions. The dried layer cracks and the strain is taken up by buckling, as shown. Metal enters through the crack at the apex of the buckle and flows behind the layer of sand and appears as a shallow, flat-topped projection on the casting, which may be of considerable length and has a rough surface. This metal can be removed fairly easily, leaving a depression, triangular in section, the sides of which are as smooth as that part of the surface of the casting which has been in contact with the undamaged mold. The actual shape of the defect can be modified by adding to the sand materials which

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become liquid or plastic when heated, e.g., pitch, boric acid or coal dust.

If the mold is only partly filled and allowed to stand, the cracks and buckles will joint up and the whole of the surface layer of sand will fall away. Figure 2 shows a die-block casting with a flat top which was incompletely filled and allowed to stand.



Fig. 2—Die-Block with Sand Falling Away from Mold Surface

The cope was removed and the casting, together with the broken layer of sand from the top of the mold,

photographed.

The mold surface remaining was almost as smooth as the original mold surface, and on standing a second layer of sand cracked and began to fall. There is one interesting feature of this surface layer of dried sand: whatever the size of the casting or the type of sand used, its thickness is about 1/8 in. to 3/16 in. Whether this is simply because the molten iron used varied between fairly narrow limits in temperature and the sands used had similar conductivities, or whether this has any other significance is not clear.

The Expansion of Sand

Molding sand consists essentially of crystals of quartz bonded together with clay. When the sand is heated, the quartz expands and the clay contracts, but as the amount of clay which can be used is limited by other factors, it is not possible so to regulate the proportions of the two constitutents as to reduce the overall expansion to any great extent. The factor of greatest importance is the expansion of quartz.

The expansion of quartz, as given by Sosman, is

shown in Fig. 3.

Up to 573 C (1065 F), quartz has a high, but regular, expansion. At 573 C there is a change from α quartz to β quartz with a further immediate expansion after which there is a slight contraction. From this point of view there are few materials less suitable than quartz as a base for a molding mixture.

It seems that if the mold surface has not buckled by the time it reaches a temperature of 573 C (1065 F), it will remain undamaged at higher temperatures.

Minimizing Effects of Expansion of Quartz

If a mold is composed of quartz grains of uniform size, the surface layer in a mold such as that shown in Fig. 1 will heat up at the same rate. The maximum expansion, and also the sudden expansion at 5/3 C (1065 F), will occur in all these grains at the same time, and the tendency of the mold surface to crack will be greatest. If grain size is spread out over a wider range, the small grains will heat up more rapidly, and will be contracting slightly when the larger grains are still expanding, and the tendency to crack will be less. The more uniform the sand grains, the greater the risk of scabbing.

In this connection it is interesting to note the gradual change which has taken place in the recommendations in American literature as to the best grain distribution. Moldenke,³ more than 30 years ago, thought that the "ideal" molding sand was composed of round grains of uniform size bonded with the smallest amount of the fattest clay available. As time went by, this was modified and it was recommended that the sand grains should be spread over three adjacent sieves in the A.F.S. series, but this optimum "spread" still tends to increase and one of up to seven sieves has been recommended.⁴

The breaking of bonded sand as a result of expansion is not confined to clay-bonded sand. J. J. Sheehan⁵ noted that some cores bonded with linseed oil and dextrin failed when the casting was poured if the base sand was composed of uniform sand, but no difficulty was experienced when a mixture of two sands differing widely in grain size was used. In this case the strain was relieved by the opening up of a crack along one edge of the core, and the defect was a fin of metal rather than a scab, but the cause of the defect was the same.

Limitations of Widening Range of Grain Size

It is possible to obtain only limited help by increasing the "spread" of grain size, owing to the necessity of maintaining mold permeability.

If a mold is composed of uniform spherical grains packed in the closest possible manner, pore space is independent of grain size and is about 25 per cent

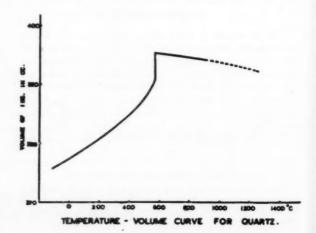


Fig. 3—Expansion of Quartz (after Sosman)

of the total volume. In the case of fine grains there will be a large number of small pores, and with a coarser sand a smaller number of larger pores. Resistance to the passage of gas through these pores will be greater when they are small, and the fine sand has a lower permeability than a coarse sand.

If a second sand differing in grain size from the first is added the mixture may or may not have a lower total pore space, depending on the difference in size between the two sands used. If the coarse sand has a diameter more than two and a half that of finer, there will be a decrease in the pore space, but if it is less than two and a half times, pore space

will be unchanged.

The grains of the sands used in foundries are never quite uniform in size or spherical in shape and when formed into a mold are not assembled in the closest packing possible. Indeed, a sand mold composed of fairly uniform grains and having a pore space below 40 per cent would be rammed exceedingly hard. It can, however, be shown that sand grains rammed to a hardness comparable with that of, say, the standard compression test piece do behave like close-packed spherical grains. (More detailed descriptions of the effect of grain distribution on permeability are given elsewhere.^{5,6})

Confirmation was obtained by experiments with four grades of sand:

No. 1-Between 10 mesh and 20 mesh

No. 2-Between 30 mesh and 40 mesh

No. 3-Between 50 mesh and 70 mesh

No. 4-Between 70 mesh and 100 mesh

None of the grains in No. 2 has a diameter more than two and a half times that of the grains in No. 3. All of the grains in No. 1 have a diameter more than two and a half times that of the grains in No. 4.

75 grams of No. 1 sand were placed in standard A.F.S. compression test piece holder and 75 grams of No. 4 added without mixing. The sand was then rammed by the standard rammer and its height measured after each blow. These heights are shown in the upper curve of Fig. 4. The sands were then removed and mixed, returned to the holder and again rammed. The new heights are shown in the lower curve of Fig. 4.

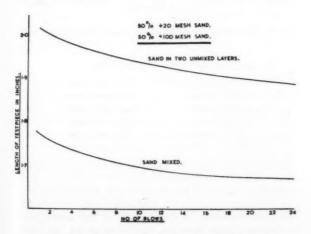


Fig. 4-Mixing Sand Grains and Pore Space

As a result of mixing, the height of the test piece has fallen from 2.0 in. to less than 1.8 in., or by rather more than 10 per cent. As the cross-section is constant, the volume has fallen by more than 10 per cent. To obtain a reduction of 10 per cent in the volume of the test piece, about one quarter of the pore space must have been filled in.

When the experiment was repeated with No. 2 and No. 3 grades, the curves obtained were very close to the upper curve in Fig. 4, and it would be necessary to use a more open scale than that shown if the

differences are to appear in the graph.

When a coarse uniform sand is added to a finer uniform sand

 (i) if the diameter of the coarse sand is less than two and a half times that of the finer, there will be no change in pore space;

(ii) if the diameter of the coarse sand is more than two and a half times that of the finer, there

will be a decrease in pore space.

As permeability is controlled by two factors, total pore space and pore size, in case (i) the addition of coarse sand to a finer sand, which leaves pore space unchanged, but reduces the number of individual pores and therefore increases their size, must increase permeability. In case (ii) the addition of the coarser sand reduces pore space at the same time as it reduces the number of pores and the final result may be a decrease in permeability.

The results of tests using the four fractions of

sand are shown in Figs. 5 and 6.

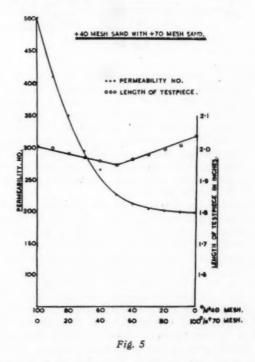


Figure 5 shows the permeability of mixtures of the +40 mesh and +70 mesh sands, in which the coarsest grains have a diameter less than two and a half times that of the finest. The lengths of the test pieces are also shown. The lengths are reasonably constant

(the slight decrease occurs because none of the grades is quite free from undersized grains), showing that pore space also remained constant, and the first addition of coarse sand produced an immediate, though slight, increase in permeability. It should be noted that permeability does not begin to rise rapidly until more than 50 per cent of the coarser grade is present.

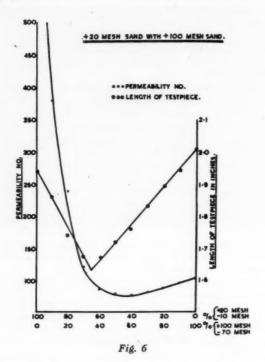


Figure 6 shows the permeability of mixtures of the +20 mesh and +100 mesh fractions, in which all grains of the coarser fraction have a diameter more than two and a half times that of any grain in the finer fraction. The addition of coarse sand to fine results in a considerable fall in pore space and a slight but measurable fall in permeability. Pore space reaches a minimum with between 50 and 60 per cent coarse sand, and permeability with between 40 and 50 per cent.

Increasing the difference in size between the two sands moves the minima in the direction of higher proportions of the coarser sand, and a mixture of +10 mesh and 200 mesh has a minimum permeability with about 70 per cent of the coarse sand present.

When the sand used is spread over three adjacent sieves in the A.F.S. series, there are few grains outside the 5: 2 ratio in diameter and pore space will be similar to that of a uniform sand. Permeability will therefore lie between that of the coarsest and that of the finest present. When, however, the grains are spread over as many as seven adjacent sieves, pore space will be lower than that of a uniform sand and permeability may be lower than that of the finest sand present. The proportions of the various grades present must therefore be controlled in such a way as to give the permeability required, and not so as to eliminate scabbing.

On this account there is a limit to the help which can be obtained by controlling grain distribution.

Other Methods of Preventing Scabbing

When the surface of a mold is heated, the grains of which it is composed expand, but the surface layer as a whole is rigidly held and the expansion is therefore taken up by buckling and cracking. There are two ways in which cracking can be reduced or eliminated:

- 1. If the sand grains are assembled in the mold so that they have room to expand, the mold will not crack, i.e., a soft mold is less likely to give a scab than a hard mold.
- 2. If the maximum deformation of the sand at any temperature reached by the time the mold is full is not less than the expansion up to that temperature, the mold will not crack.

Effect of Mold Hardness

The harder the mold is rammed, the greater the tendency for expansion scabs to be formed. At the same time, permeability also falls, particularly in the case of sands whose grains are spread over a wide range in size. The variation of permeability with mold hardness of a natural red sand is shown in Fig. 7.

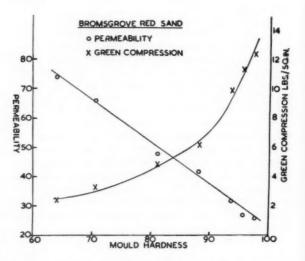


Fig. 7-Variation of Permeability with Mold Hardness

This sand has the following grading:

+ 60 mesh	3.6	per	cent
+100 mesh	31.0	per	cent
+150 mesh	36.9	per	cent
+200 mesh	10.4	per	cent
-200 mesh	6.7	per	cent
Clay	11.4	per	cent

Sand of this type occurs in England in very large quantities in the Midlands and in Lancashire and Yorkshire. It is still widely used, but in many foundries has been replaced by synthetic sands. It was at one time exported to many parts of the world, not excluding North America, and even as late as 1920 there was a small export to places as far away as

New Zealand. Quite large and difficult castings were, and to some extent still are, produced in green molds made from this sand and it is very instructive to go over one of these large molds. Hardness varies over a very wide range, and everywhere seems to be adjusted so that the mold is only just hard enough to prevent swelling. As soft ramming also raises permeability, it not only of itself reduces the risk of scabbing, but makes possible the use of a sand less likely to cause this defect. The objections of many skilled hand molders to synthetic sands are often due not only to the difficulties which result from their narrow moisture range, but to the increased risk of scabbing.

When molds are rammed by mechanical means, this selective control of mold hardness is not possible, at least to the same extent, but a mold hardness tester should be available for use in every foundry.

Reducing Scabbing by Increasing Deformation of Dried Sand

In the foundry, several additions are made to molding sand in order to prevent scabbing, the most important being sea coal and wood flour. It will be shown later that at certain stages during the heating of a mold, these additions increase the deformation of the sand.

It is, however, an extremely difficult matter to determine the effect of any organic material on the properties of molding sand at high temperatures if these tests are to be related to the incidence of scabbing. The expansion of quartz reaches a maximum at 573 C (1065 F), a temperature reached by the mold surface a short time after pouring commences, and this time and temperature are insufficient to cause complete decomposition of the organic matter present. On the other hand, when a test piece is to be heated to a uniformly high temperature throughout, any organic matter is completely decomposed, leaving only ash and perhaps inert carbon. It is doubtful whether tests made under these conditions can be applied to what happens in a mold when a casting is poured. Indeed, expansion tests made in this way appear to indicate that the addition of sea coal reduces the expansion of sand on heating, but it is difficult to believe that the addition of a material which expands with great rapidity when suddenly heated can reduce expansion during the critical early seconds when a scab is formed.

Use of Test Castings

In consequence of this difficulty, investigators^{7,8,0,10} into causes of scabbing have relied to a great extent upon test castings poured in molds faced with the sand to be tested, or on test pieces subjected to sudden heating. The results obtained are most informative, but the procedure is laborious, and their full value can only be realized if the results can be related to certain physical properties of the sand. Accordingly, most investigators working with test castings also measure as many as possible of the physical properties of the sand used as their equipment permits.

This method is perhaps the most satisfactory which

can be designed at the moment, but it has certain pitfalls. For example, most of the sands which gave castings free of defects in the experiments made by the A.F.S. Committee⁹ have a low "hot strength" and it might be argued that any sand with a low hot strength will be free from scabs. It is, however, difficult to see how a sand can be prevented from breaking by reducing its strength, and it is possible to reduce hot strength by means other than those used, without in any way reducing the amount of scabbing.

Test castings have been used in the laboratories of the British Cast Iron Research Association for an investigation into the causes of scabbing for a number of years. Some of the results of this investigation have already been given¹⁰ and some of the photographs shown now are taken from Taylor's paper. At about the same time the construction of apparatus to measure the deformation under load of dry sand was commenced, but a number of mechanical difficulties were experienced, and the results of many of the test castings were available before the apparatus was ready for use.

Design of Test Castings

The test casting is a simple block, 1 ft square and 2 in. in thickness. Two different sizes of runner are used, one of which fills the mold in about 25 sec, and the other in 50 sec. There is a riser 2 in. in diameter in the center of the top. This aperture in the top of the mold has been used as a peep-hole, with the object of noting the time which elapses between the commencement of pouring and the first breaking away of sand from the top surface, and obtaining the temperature of the mold surface at this time. The information so obtained was insufficiently exact to be of use, but the riser has been retained so as to avoid any change in the design of the casting.

Several other designs were considered, particularly one in which the metal flows over a step, but a few preliminary tests showed that a scab due to low permeability might be obtained. The simpler design was therefore retained.

This particular casting was originally designed by J. J. Sheehan and has been used by a number of observers for studying scabs, but only one reference¹¹ to it has been found.

As the object of the investigation was to compare sand mixtures, everything else was kept as constant as possible. Mold hardness was maintained at 70-75 and only two sizes of ingate used.

The casting was molded in the cope, and the sand under examination was used as a facing round the top and sides of the pattern only. The remainder of the cope and the whole of the drag were rammed up in moistened floor sand, originally the red sand described earlier, and containing about 4 per cent sea coal. Incidentally, not a single scab was found on the bottom of the casting, although more than a hundred have been made to date.

Preliminary Laboratory Tests

In an earlier investigation into the possibilities of controlling scabbing, an attempt had been made to show that coal dust may increase the deformation of heated sand.12 All the high volatile sea coal used in molding sands become temporarily plastic or even

liquid when heated.

Sand test pieces heated externally are useless for this purpose. Assuming for the moment that heated coal dust does increase deformation, the effect can be only temporary as when all the volatile matter has been removed a non-plastic coke is left. When a sand test piece is heated from the outside, the coal dust in the surface layers becomes plastic, but the remainder is unaffected. Heat penetrates only slowly owing to the low heat conductivity of the sand, and by the time the coal dust a little way beneath the surface becomes plastic, that in the surface layers loses its plasticity because all its volatile matter has been driven off. The plastic layer gradually moves towards the center of the test piece, but at no time is the coal dust in a complete cross-section of the test piece plasic. No tests on such a test piece can show the effect of the plasticity of heated coal dust.

A "sand" was therefore prepared by bonding steel grit with bentonite. Test pieces 2 in. in length and 11/8 in. in diameter were prepared from this sand, with 6 per cent and without coal dust. The steel grit was almost entirely on 22 and 30 mesh sieves.

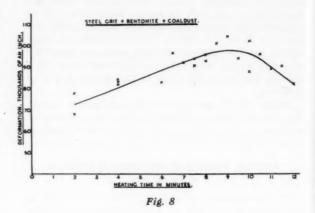
The test pieces were placed in an induction furnace and heated. They were then removed from the furnace and broken, the deformation being recorded by means of an Ames dial. The results are shown in Fig. 8, in which deformation is plotted against time of heating.

This method of testing is not very satisfactory as the rate of heating is too slow, the temperature after 12 min being only 930 F. If an attempt is made to

Synthetic Sand 100 Pouring Time 50 sec

Moisture	2.1	pct
Permeability No.	167	•
Green Strength	3.6	psi
Dry Strength	46	psi

increase the rate of heating, the test piece disintegrates, individual grains being shot off with considerable violence. Nevertheless, the test does show that even under these unfavorable conditions coal dust does increase the deformation of sand at high temperature.



· When the coal dust was replaced by 1 per cent of wood flour, deformation exceeded 0.150 in. throughout the test. Fibrous materials also increase deformation.

Materials Used in Casting Tests

Test castings were made in mixtures to show the effect of the composition and grading of the base sand and of adding the following materials:

- 1. Materials which become plastic when heatedsea coal, pitch and boric acid.
- 2. Wood flour.
- 3. Fibrous materials-asbestos, grass cuttings.



Bonded Olivine 100 Pouring Time 53 sec

Moisture 2.9 pct Permeability No. 180 Green Strength 7.3 psi Dry Strength 80 psi

Fig. 10



Natural Sand 100 Moisture 4.6 pct
Pouring Time 50 sec Permeability No. 27
Green Strength Dry Strength 85 psi

Fig. 11

The mechanical compositions of the two sands used were as follows:

	Erith Silica Sand	Red Sand
+ 44 mesh	2.6	0.3
+ 60 mesh	26.1	3.3
+100 mesh	63.3	31.0
+150 mesh	. 7.8	36.9
+200 mesh	0.5	10.4
_200 mesh	0.3	6.7
Clay	0.4	11.4

The silica sand was bonded with 5 per cent by weight of bentonite.

Test Castings Illustrated

Some of the test castings made are shown in Figs. 9 to 25. These particular castings have been selected to show the effects of:

- Thermal expansion of the sand grains—molds made in silica sand (Fig. 9) and olivine of the same grading (Fig. 10).
- Grain distribution—uniform grain size (Fig. 9) and wide range of grain size (Fig. 11).
- Materials which become plastic on heating coal dust, pitch and boric acid (Figs. 12-21).
- 4. Wood flour (Figs. 22 and 23).
- Fibrous materials—asbestos (Fig. 23) and grass cuttings (Fig. 24).

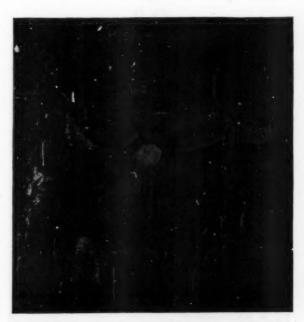
Discussion of Results with Test Castings

Composition of Basic Sand. The castings shown in Figs. 9 and 10 show that scabbing is due to some property of quartz. When quartz was used the surface of the casting was covered with a network of scabs, but when the quartz was replaced by olivine,



Synthetic Sand 94 Moisture 3.0 pct
Coal Dust 6 Permeability No. 115
Pouring Time 57 sec Green Strength 6.7 psi
Dry Strength 48 psi

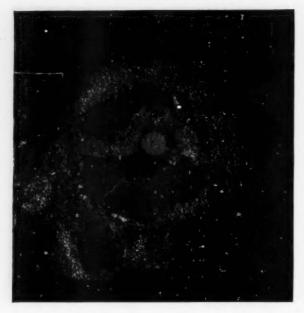
Fig. 12



Synthetic Sand	99	Moisture	3.7	pct
Pitch	1	Permeability No.	125	-
Pouring Time	48 sec	Green Strength	5.1	psi
		Dry Strength	110	nei

Fig. 13

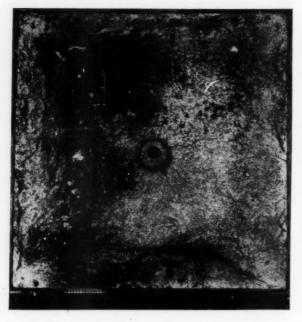
graded so that its mechanical composition was as far as possible identical with that of the quartz, no scabs were formed. Scabs were also absent when sillimanite or zircon were used in place of olivine, although in these cases the evidence is not so strong as there was a difference in grading.



Synthetic Sand 98
Pitch 2
Pouring Time 57 sec

Moisture 3.6 pct Permeability No. 136 Green Strength 5.7 psi Dry Strength 195 psi

Fig. 14



Same Sand as Fig. 14 Pouring Time 26 se

Fig. 15

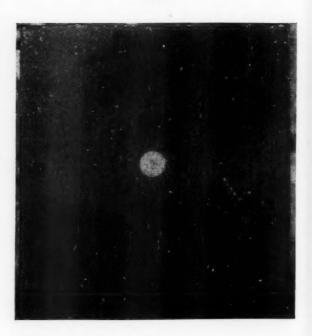
The coefficient of expansion of quartz is much greater than those of olivine, sillimanite or zircon and the test castings gave strong support to the generally accepted theory that the scab is due to expansion of sand grains.

Influence of Grading. Increasing the range of grain



Synthetic Sand 74 Moisture 4.0 pct
Natural Sand 25 Permeability No. 107
Pitch 1 Green Strength 6.1 psi
Pouring Time 45 sec Dry Strength 135 psi

Fig. 16



Same Sand as Fig. 16 Pouring Time 23 sec Fig. 17

size does not affect the expansion of either the individual grains or the sand as a whole, but it does reduce the shock at 1060 F by ensuring that all grains in the surface layer do not undergo the sudden expansion at the same time.

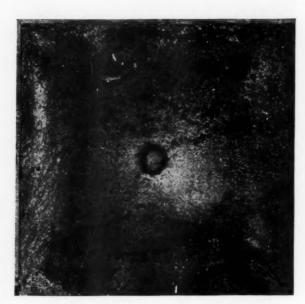
When the synthetic sand was replaced by increas-

120 psi



Synthetic Sand Natural Sand 69 Moisture 4.5 pct Permeability No. 25 13 **Green Strength Coal Dust** 6 8.1 psi **Pouring Time** 45 sec Dry Strength 160 psi

Fig. 18



Same Sand as Fig. 18 Pouring Time 23 sec

Fig. 19

ing proportions of the natural sand so as to widen the range of grain size, there was a continuous, though slight, decrease in the extent of scabbing, its general form remaining unchanged. When natural sand only was used, the casting shown in Fig. 11 was obtained.

The difference between Fig. 9 and Fig. 11 is an

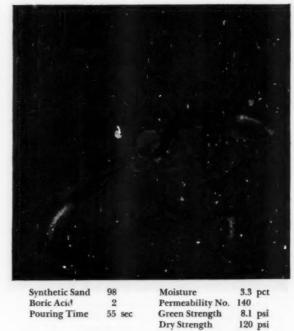
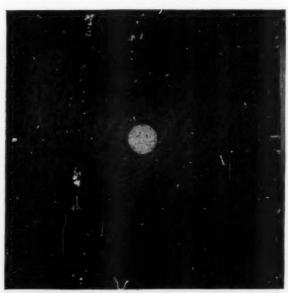


Fig. 20

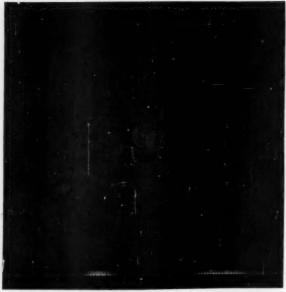


Synthetic Sand	95	Moisture	3.9	pci
Boric Acid	2	Permeability No.	110	•
Coal Dust	3	Green Strength	7.6	psi
Pouring Time	52 sec	Dry Strength	146	

Fig. 21

indication of the extent to which scabbing is influenced by the immediate expansion which accompanies the change from a quartz to β quartz.

Additions of Materials which become Liquid or Plastic when Heated. The addition of pitch or coal dust, which become plastic or liquid for a short time when



Synthetic Sand 98 Wood Flour 2 **Pouring Time** 52 sec

Moisture 3.8 pct Permeability No. 205 Green Strength 4.5 psi Dry Strength 35 psi

Fig. 22



Synthetic Sand Wood Flour 1 **Pouring Time** 53 sec

Permeability No. 150 Green Strength 5.4 psi Dry Strength 42 psi

Fig. 23



2.4 pct Moisture Permeability No. 180 **Green Strength** 6.0 psi Dry Strength

Synthetic Sand 99.5 Asbestos (1 in.) 0.5 **Pouring Time Pouring Time** 50 sec

Fig. 24

99.5 Synthetic Sand 0.5 Permeability No. 35 Grass Cuttings 5.1 psi 50 sec Green Strength 28 psi Dry Strength

Fig. 25

heated, or boric acid which remains liquid so long as the temperature is maintained, can reduce, and sometimes eliminate, scabbing under certain conditions. For example, 6 per cent coal dust or 1 per cent pitch completely suppressed scabbing when a mixture of 3 parts synthetic sand and 1 part natural

sand was used, provided that the pouring time was kept down to 25 sec (Figs. 17 and 19). No mixtures containing boric acid alone were completely satisfactory (Fig. 20), but there were no scabs on even the synthetic sand in the presence of 2 per cent boric acid and 3 per cent coal dust (Fig. 21).

There are, however, limits to the usefulness of pitch and coal dust. Both materials are liquid or plastic only in that range of temperature in which they decompose. When the time required to pour the casting is increased, because of size or some other factor, decomposition may result in removal of most of the liquid phase before pouring is completed, and a scab may form. If the amount of the addition is raised to offset the loss by decomposition, then at some stage during pouring the amount of liquid phase may become so great that the sand collapses under its own weight and the influence of a slight stress due to expansion.

In this case a scab will also occur, but generally its form is changed (Figs. 14 and 18). The number of fractures of the mold surface is reduced, but the in-

dividual scabs are much larger.

There is one unexpected feature about these larger scabs. In one case (Fig. 18) there were two large scabs, each covering a quarter of the mold surface. In a second (Fig. 14) the large scabs were arranged in a uniform pattern round the center of the mold. When these tests were repeated, the same mixture invariably gave the same kind of scab. If, when a mixture was tested, opposite quarters of the mold surface came down, then repeat tests always gave the same result, although it was not possible to forecast the form the defect would take when a different mixture was used. No reason for this difference has been found.

These results are entirely in line with the experience in a number of foundries. One claims that scabbing can be eliminated by increasing the proportion of coal dust, whereas another is quite satisfied that such a procedure makes matters worse. It will be found that the first foundry pours fast and the second, perhaps of necessity, pours slow.

Wood Flour. The test results illustrated in Fig. 8 show that coal dust temporarily increases the deformation of heated sand, and its influence on scabbing was attributed to this cause. At the same time it was pointed out that wood flour gives a much greater deformation both at room temperature and when heated.

In the presence of 1 per cent wood flour (Fig. 22), no scabs were obtained on any of the test castings.

Wood flour is a much safer material to use than coal dust or pitch. As the effect of maintaining it at a high temperature is confined to a more or less steady loss through decomposition, and as there is at no time a liquid phase present which can weaken the sand so that it fails under its own weight and a small additional stress, increasing the amount and the time of exposure does not result in the formation of a large scab. With 2 per cent and a pouring time of 50 sec, no scabs were formed (Fig. 23). The limiting factor so far as wood flour is concerned is its effects on molding properties.

Fibrous Materials. The addition of 0.5 per cent asbestos fiber cut into half-inch lengths, completely eliminated scabs (Fig. 24) even with the 50 sec pouring time. Grass cuttings were almost completely successful, only a very small scab near the riser being found (Fig. 25). For this particular mold, stiff grass stems were used. When these were replaced by leaf a clean casting was obtained.

All fibrous materials are useful, but in general pliable fibers are completely successful, but the tendency for scabs to form increases as stiffer fibers are

used.

Asbestos, and to a lesser degree the other fibers used, appear to have a dual effect. In the first place they reduce cracking of the heated mold surface at least as effectively as wood flour, and in the second place they hold the broken parts of the mold together even if cracks are formed. If, for example, a standard dried test piece is broken in compression in the usual way, it can be picked up, supporting it from the top, if 0.5 per cent asbestos is present, although a close examination shows that it consists of eight or more pieces loosely attached to each other by fibers. Cohesion between sand grains may have been completely destroyed along several planes, but the fibers still hold. In the presence of asbestos no scabbing would occur, and the worst that could happen would be a few sand grains in the surface of the casting and a few shallow fins.

A further advantage of using an inorganic fiber is that only a very small amount would be destroyed during the casting operation. To control the amount present with sufficient accuracy for practical purposes, it would only be necessary to add a definite proportion to the new sand going into the system.

The one drawback is the difficulty of mixing a fibrous material uniformly with molding sand. From this angle the most promising results have been with asbestos waste, which goes in quite readily when shaken up with water. As it contains a large proportion of powder it has to be used in comparatively large quantities.

Deformation of Dried Molding Sand

Tests made on steel grit bonded with bentonite and heated in an induction furnace showed that coal dust increases the deformation of sand at temperatures near 500 C (930 F) and that wood flour gives an even greater deformation at all temperatures up to 500 C (930 F).

The results shown in Fig. 8, however, are not sufficiently precise to do more than confirm that materials which tend to reduce scabbing increase deformation, at least over a considerable temperature range.

It was therefore decided to build apparatus to obtain with greater accuracy the deformation of molding sand, and preferably a complete stress/strain curve. To obtain all the information required it will be necessary to arrange for the test piece to be heated rapidly and uniformly, so that the whole of it is at a temperature above 573 C (1065 F) before any organic matter it contains is completely decomposed. There are no technical obstacles to this, and sand test pieces have already been heated to 1000 C (1830 F) inside 60 sec, but bringing together the heating and testing equipment is a lengthy operation.

The apparatus for obtaining stress/strain curves has been completed, and a large number of sands examined at room temperature. It was not expected that very much information would be obtained from such tests until the heating equipment was available, but, with the important exception that little comes out of the tests on sands containing coal dust or pitch, there is such a constant relation between the form of the stress/strain curve and the tendency of the sand to scab, that the results seem worth considering.

The ultimate object of the investigation, however, is to determine the behavior of sands at high temperatures, and the results given here can be regarded as

merely a progress report.

Design of the Apparatus. When an increasing load is applied to a test piece of dried sand, a small deformation, increasing to the point of fracture, takes place. When the test piece breaks, it usually shatters, and it is not easy to determine the deformation, since the only apparent change is a very rapid acceleration in the rate of deformation.

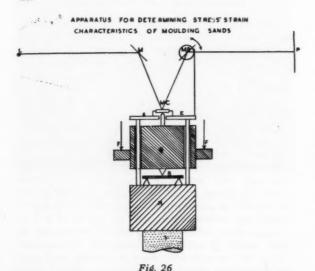
A preliminary investigation showed that not only was it difficult to determine deformation at rupture point accurately, but the results obtained with sands which behaved very differently in the foundry, differed by an amount not much greater than the possible error. The most likely method of overcoming this difficulty seemed to be to arrange for the stress to be relieved automatically as the test piece deformed. This has the effect of flattening out the curve after the maximum stress has been reached.

The preliminary specification for the apparatus therefore included the following two points:

1. The method of applying the stress must be such that as the test piece deforms there is an automatic relaxation of stress.

2. Stress and strain must be recorded automatically.

The final design of the machine is due to R. G. Godding¹³ and the essential parts are shown in Fig. 26.



The load is applied at FF by means of a motordriven screw jack and is transmitted to the plunger H by means of a knife edge and a calibrated beam B. The plunger rests on the specimen S which is supported by a freed anvil.

A tilting mirror carriage E carrying a mirror MC is attached to the plunger by means of rods passing through block G. When a load is applied at FF, beam B bends and causes mirror MC to tilt in proportion to the applied load.

Should the specimen deform under load, block H moves downwards and this downward movement causes a second mirror, to which block H is connected by a rack and pinion linkage, to rotate pro-

portionally.

The downward movement of H automatically relaxes the load on the specimen, the relaxation being 40 lb for each 0.001 in, deformation in the tests reported here.

A beam of light from lamp L is reflected from a fixed mirror M on to mirror MC. As the latter tilts, the light beam is deflected by an amount propor-

tional to the applied load.

The beam then falls on mirror MR and, as this mirror rotates, is again deflected, the deflection in this case being proportional to the deformation of the specimen.

The apparatus is so arranged that the two deflections are at right angles, and a complete stress/strain curve can be obtained on a photographic plate at P.

In order to obtain an idea of the rate of strain, the lamp is made to flash five times per second.

Test Results

The results of tests on a number of sands are shown in Fig. 27-30. The stress/strain curves can be related with the results of the corresponding test castings from the series shown in Figs. 9-25.

Sand of Uniform Grain Size. Stress/strain curves for the synthetic sand at two moisture contents are shown in Fig. 27. An increase in moisture content raises dry strength, but does not increase deformation.

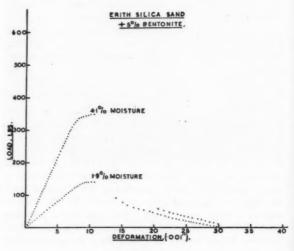


Fig. 27

As the load is increased, deformation increases until the maximum load is reached, after which there is a further slight deformation followed by a sudden collapse. With the higher moisture content one dot is shown at 0.012 in. deformation and the next at

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0.021 in., i.e., the test piece has collapsed 0.009 in. in a fifth of a second, although the stress has been relaxed by some 200 psi.

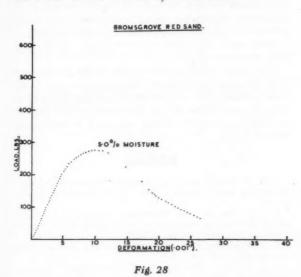
At the lower moisture content, the general effect is similar although maximum stress is lower and collapse is a little less rapid. At this stage it is not possible to attach too much to the difference between these two curves.

This sand was used to produce the test casting shown in Fig. 9.

A number of sands of uniform grain size, bonded with several different types of clay have been tested. There were wide differences in maximum stress, but deformation was fairly constant and all stress/strain curves showed the sharp break after maximum stress has been reached.

All these sands gave castings similar to that shown in Fig. 9.

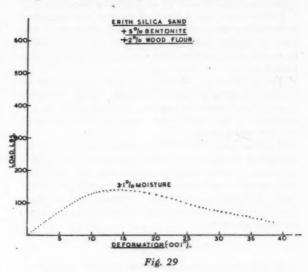
Sands of Non-Uniform Grain Size. The curve for the natural sand is shown in Fig. 28. This curve is of a slightly different shape and begins to bend over at a fairly low stress (this is probably a function of the clay bond), but maximum deformation is little greater than that of the uniform synthetic sand.



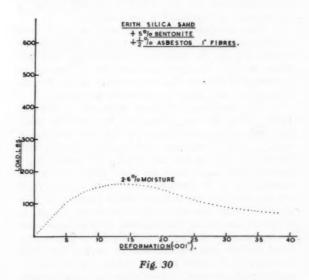
The casting obtained with this sand is shown in Fig. 11. There is a real reduction in the amount of scabbing and in view of the similarity of the stress/strain curves, this must be attributed to the effect of variation in grain size on the sudden expansion at 573 C (1065 F).

Effect of Wood Flour. The effect of adding 2 per cent wood flour to the synthetic sand is shown in Fig. 29. Maximum stress is halved and deformation increased, but the most important change is in the shape of the curve. After the maximum stress has been reached, the test piece continues to deform under a gradually decreasing stress, but there is no sudden fall in the stress required to continue this deformation. Maximum stress is around 150 psi and deformation at this point is 0.013 in., but when deformation is increased to 0.030 in. a load of 100 psi is still required to continue deforming the specimen.

The test casting made in this sand is shown in Fig. 22 and is entirely free from scabs.



Effect of Fibrous Materials. The effect of adding 0.5 per cent asbestos to the synthetic sand is shown in Fig. 30. In this case there is no appreciable change in maximum stress, but with this exception asbestos behaves very much like wood flour. Deformation at maximum stress is increased, and after the maximum stress has been reached deformation continues under a gradually decreasing stress.



The casting made in this sand is shown in Fig. 24. It is entirely free from scabs.

All pliable fibers give similar stress/strain curves and also eliminate scabbing. If more brittle fibers are used, the fall in the stress required to maintain deformation is somewhat more rapid, but never approaches that found in the plain sand. In such cases an occasional small scab is found on the test casting.

The Form of the Stress/Strain Curve and the Incidence of Scabbing. All sands with stress/strain curves simi-

lar to those shown in Fig. 27 and Fig. 28 gave scabbed test castings. When the curves were similar to those shown in Figs. 29 and 30 no scabbing occurred.

Furthermore, the severity of the test was increased by molding the casting in the drag. The mold was filled to within half an inch of the top and allowed to stand. Even under these conditions no sand was pulled down.

The addition of asbestos increases deformation at maximum stress by about one-half. This difference is hardly sufficient to explain the difference in the behavior of the sands when used to make the mold

for the test casting.

It seems plain that the important point is the deformation which occurs before the stress falls to that required to cause complete rupture of the mold surface, which is not more than a few pounds per square inch. The addition of either asbestos or wood flour increases deformation at this point to several times its original value.

Conclusions

1. The incidence of scabbing has been studied by means of test castings, and the results have been correlated with certain physical properties of the sand.

2. Under the conditions of the test, the defect was not produced when the quartz which normally forms the greater part of the molding sand is replaced by a granular material such as olivine or zircon, which has a lower thermal expansion.

3. Increasing the size range of the quartz grains reduces the tendency to scab, but the extent to which size distribution can be varied is limited owing to

the necessity of maintaining permeability.

4. The addition of materials such as pitch, coal or boric acid, which become liquid or plastic when heated, reduces the incidence of scabbing. The addition of these materials increases the maximum deformation of sand at the temperatures at which scabbing occurs and helps to eliminate the defect.

- 5. The usefulness of these materials is limited. Those which do not decompose, for example boric acid, reduce the refractoriness of the sand and would cause trouble in stripping and fettling. Those which decompose, such as coal and pitch, only persist for a limited time, and when the time taken to fill the casting has to be increased for some reason, their effectiveness may be reduced before pouring is completed. If the amount used is increased in an attempt to compensate for this loss, then at some time during pouring the amount of liquid or plastic material present may be sufficient to weaken the mold so that it fails under its own weight. As a result, a scab is again formed, usually differing in appearance from the scab obtained in the absence of a plastic material.
- 6. Wood flour and fibrous materials, both organic and inorganic, increase the deformation of sand throughout the whole of the temperature range in which expansion occurs and prevent scabbing. They do not themselves become liquid or plastic and are therefore useful under a wider range of conditions.
- 7. When the sand contains materials which become plastic or liquid when heated to room temperature,

tests do not indicate the tendency of a sand to scab. Tests made on sands heated from outside also fail if the materials decompose, because the whole of the liquid phase is destroyed before a test piece attains a uniform high temperature. The effect of such materials can, however, be shown qualitatively by including them in steel grit bonded with clay and heated in an induction furnace.

8. In the absence of materials which become liquid or plastic, the tendency of a particular sand to scab can be related with the stress/strain curve of the

dried sand at room temperature:

(i) If the deformation at maximum stress is low, and the test piece fails rapidly once maximum stress is reached, the sand will scab readily.

- (ii) If the deformation is high, and if, after maximum stress has been reached, a test piece continues to deform under a slowly decreasing load, the sand will not scab.
- 9. Sands containing sand and clay only have a low deformation and collapse rapidly once the maximum stress is reached. The addition of wood flour or of a fibrous material increases deformation at maximum stress. At the same time the shape of the stress/strain curve is radically changed, the test piece continuing to deform under a gradually decreasing stress after the maximum stress has been reached.

Sands with this type of stress/strain curve have

little tendency to scab.

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References

1. H. W. Dietert and E. E. Woodliff, "Measure Deformation of Molding Sand," Foundry, 1939, vol. 67, September, pp. 28-

2. E. L. Graham, "The Flowability of Moulding Sand with Special Reference to the Shatter Test," Foundry Trade Journal, 1948, vol. 84, April 15, pp. 365-367.

3. R. Moldenke, Principles of Ironfounding, 2nd ed., 1930,

McGraw-Hill, 654 pp.
4. J. Silk, "Molding Sand Standardized," American FOUNDRYMAN, 1949, vol. 16, No. 3, September, pp. 51-53.

J. J. Sheehan, "Core-shop control." Proceedings, Institute of British Foundrymen, 1938-9, vol. 32, pp. 43-58; discussion,

6. W. B. Parkes, "Synthetic sands," B.C.I.R.A. Research Re-

port No. 138, 1938, January, 34 pp.
7. H. W. Dietert and F. Valtier, "The Expansion and Contraction of Molding Sand at Elevated Temperatures," TRANSAC-TIONS, American Foundrymen's Association, 1935, vol. 43, pp. 107-122; discussion, pp. 122-124.

8. F. Hudson, "Composition and Its Effect upon the Properties of Mould and Core-sand Mixtures at Elevated Temperatures," Proceedings, Institute of British Foundrymen, 1935-6,

vol. 29, pp. 155-188; discussion, pp. 188-201.

9. A.F.S. Committee Progress Report on Physical Properties of Iron Molding Materials at Elevated Temperatures, "Scabbing Tendencies of Molding Sands," TRANSACTIONS, American Foundrymen's Society, 1949, vol. 57, pp. 400-407; discussion,

10. D. A. Taylor, "Influence of Sand on the Formation of Scabs," B.C.I.R.A. Journal of Research and Development, 1950, vol. 3, No. 9, December, pp. 693-698; discussion, pp. 699-701.

11. K. W. Slack and W. J. Rees, "The Effect of Variation in · ·

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the Drying Temperature on the Dry Strength of Moulding Compositions," Iron and Steel Institute, Special Report No. 23, Third Report of the Steel Castings Research Committee, 1938, pp. 183-200.

12. W. B. Parkes, "Effect of Coal Dust on Moulding Sand," B.C.I.R.A. Journal of Research and Development, 1945-7, vol.

1, No. 2, April, pp. 81-90.

13. R. G. Godding, "Load Deformation Curves on Dried Sand Specimens and Their Relation to Expansion Defects," B.C.I.R.A. Journal of Research and Development, 1951, vol. 4, No. 3, December.

DISCUSSION

Chairman: J. B. Caine, Consultant, Wyoming, Ohio.
Co-Chairman: H. W. Dietert, Harry W. Dietert Co., Detroit.
Recorder: R. G. Thorpe, Cornell University, Ithaca, N. Y.

W. Y. BUCHANAN (Written Discussion)¹: There is no doubt that silica sands expand on heating and may be a contributory factor in scabbing.

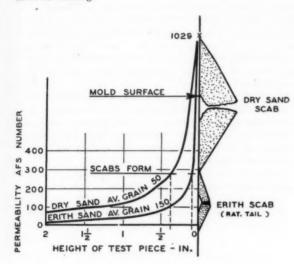


Fig. A

The total amount of expansion appears to be about 0.4 per cent while the air spaces in the rammed sand may reach 40 per cent. This suggests that much of this 0.4 per cent expansion can be accommodated on 40 per cent air space if a little resistance is offered.

After going carefully through expansion tests and various types of expansion test castings included the 12-in, x 12-in, x 2-in, the writer is of the opinion that the scab formation is predominantly a matter of gas pressure under the mold surface, permeability at that region, and strength to resist fracture under

pressure.

Isolated scahs can be better explained by gas pressure. Expansion must be uniform over the whole surface but gas pressure can concentrate on a weak spot like a blister on a defective car tire.

Wood flour and other combustible material shrink on drying or burn out and thus improve permeability and so reduce scabbing.

Fiber increases strength during the critical high pressure

period and so reduces fracture of scabbing.

If permeability is considered as a measure of the amount of resistance offered by rammed sand to the passage of gas, it will be seen from the study made of progressively shorter test pieces to the same degree of ramming without correction for reduction in length that the permeability number of a normal drysand specimen rises from 70 at 2-in. length to 290 at 3/8 in. and this is the normal thickness of a scab in this sand, which has a predominant grain size of about 50 mesh.

Plotting permeability of this sand against length of test piece and comparing with similar tests on Erith Loam of predominant grain size 150, it will be seen that if scabs form by back pressure at 3/k in. below the mold surface where the permeability is 290, a similar back pressure in Erith sand will take

place at slightly less than 1/8 in.

This suggests that the scab in the open-grained drysand is probably the same defect as the rat tail in the fine grained Erith sand, and that all scab defects are related whether they are called shell scabs, blind scabs, dumb scabs, buckles, draws down, or rat tails, with expansion a contributory factor (if it can be proved).

Good castings have been made in olivine but good castings have also been made in silica sand.

John Lang & Sons, Ltd., Johnstone, near Glasgow, Scotland.

PRACTICAL ASPECTS OF OLIVINE AS A MOLDING MATERIAL

By

John Sissener* and Bjorn Langum**

Olivine is an iron-magnesium silicate which has been used for years as molding material in the Norwegian foundry industry, especially in the steel foundries. The mineral was first described by J. G. Werner in 1790. The name is due to its green color.

The olivine rocks in Norway belong to the dunites, and contain minor quantities of other minerals like enstatite, magnesium emphibole, and traces of chromite, magnesite, phlogopite and corundum. Some deposits containing serpentine are less suited as molding material. Minor deposits exist as natural sands. The hardness of the mineral is 6.5 to 7 (Moh's scale) and the specific gravity is 3.25 to 3.4.

The deposits are enormous, and open air mining is employed. The rock is crushed and delivered as molding sand. As delivered, olivine has the following approximate composition: 50 per cent MgO. 41 per cent SiO₂, 6 per cent FeO, small quantities of Fe₂O₃, Cr₂O₃, Al₂O₃, NiO, MnO, and traces of CoO. Ignition loss is 0.5 per cent.

Sintering point is above 2900 F. Graph 1 shows the system MgO-SiO₂ after Bowen-Andersen¹. Graph 2 shows the system Mg₂SiO₄-Fe₂SiO₄ after Bowen-Schairer². Olivine may be regarded as a neutral or slightly basic material. The diagrams show that the pure magnesium orthosilicate—"forsterite"— Mg₂SiO₄—is very refractory, but that the refractoriness is considerably less with increasing SiO₂ and iron oxides.

The first attempts at using crushed olivine synthetically bonded as a molding sand started in a Norwegian steel foundry in 1927-1928³. As olivine was several times more expensive than silica, there was at that time no reason to proceed with olivine as molding material. But in 1938 industrial use of crushed olivine as a molding material for high chromium-nickel steel was successfully tried, and olivine has been used ever since because of the good casting surface produced.

Until 1938 only a few cases of silicosis were known in Norwegian foundries, but in that and the following year many foundry workers were x-rayed. Silicosis was found to such a degree that it called for remedies. Be-

cause olivine contains silicon dioxide combined with MgO and in smaller proportions, it was presumed that the risk of using olivine as molding sand was not as great as in using pure silica sand. Based on this presumption, olivine was again introduced as a molding material and has been used since then in Norwegian steel foundries.

In 1940, tests were carried out with rabbits⁴. The rabbits were exposed to silica dust and olivine dust under the same conditions. In England tests were carried out with rats⁵ in 1943-1945. In both cases examination of the lungs showed olivine dust to be less dangerous than silica.

After the German occupation the production of suitable crushed olivine as a molding material was in-

		SAND TYPE									
	1	2	3	4	5						
	Α. 1	F.S. GRA	IN FINEN	SS NUMB	ER						
	67	108	132	214	34						
SIEVE NUMBER		PERCE	NTAGE R	ETAINED							
6	-	-	-	-	-						
12	-	-	-		0.5						
20	9.5	0.5	1.5	-	24.0						
30	12.5	2.5	3.0	-	16.5						
40	16.0	3.0	2.5	-	17.0						
50	16.0	6.0	3.5	-	18.5						
70	17.0	15.5	10.5	6.0	15.0						
100	10.5	24.5	21.0	9.5	5.0						
140	7.0	19.0	14.5	10.0	2.0						
200	2.5	13.0	13.5	11.0	0.5						
270	1.5	5.5	14.0	12.0	0.5						
Pan	7.5	10.5	16.0	51.5	0.5						

Table 1 . . . Showing the grain composition of sand types referred to throughout this article.

President, Norwegian Foundrymen's Association and
 Foundry Manager, A/S Drammens Jern, Drammen, Norway.

Official Exchange Paper from the Norwegian Foundrymen's Association to the American Foundrymen's Society, presented May 2, 1952 at the International Foundry Congress in Atlantic City, N. J.

A	В	С	D	E	USES
1	90	2.3	7.8	5.4	Steel molding sand
2	60	2.8	9.2	-	Steel core sand
3	24	3.3	11.4	3.0	Molding sand for machinery castings
4	20	2.1	8.1	4.6	Molding sand for stove plates
5	12	2.4	8.0	3.3	Molding sand for stove plates
6	100	0.4	1.7	-	Oil sand cores
7	200	-	-	-	Oil sand cores

Table 2... Qualities of sand mixture after mixing. Letters at top of columns represent the following: A, sand mixture; B; green permeability; C, green shear strength, psi; D, green compression strength, psi; E, moisture.

creased, and the steel foundries tried to replace silica with olivine as molding material. It is of primary importance to use olivine in steel foundries because the danger of silicosis is greater than in iron foundries. In Norwegian iron foundries the trend to use olivine has not been so general because of certain technical difficulties which will be dealt with later on in this paper.

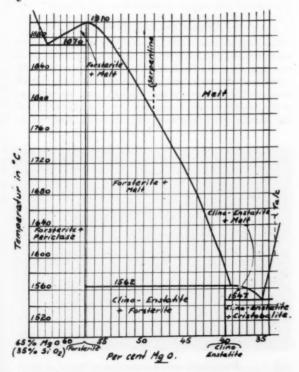
One very important aspect of olivine is its ability to withstand heating and cooling without disintegration. The thermal properties of a molding material based on olivine are different from those based on silica. Tests are going on in Norway to establish comparable figures. Of great practical significance is the heat flow in the molding sand. When finely crushed,

olivine rock is also used as a parting and dusting powder and in washes. The annual consumption of crushed olivine as molding material in Norway is 8,000 tons. The price of crushed olivine, Table 1, sand type 1, is \$9.00 per ton; silica sand costs \$5.60 per ton.

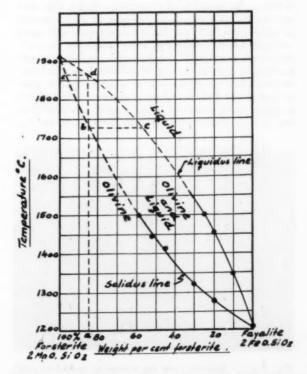
Olivine for Steel Castings

Olivine has given good results as a molding material for steel castings, but it has not been proved that results are any better than with ordinary silica sand except for manganese and high chromium steels to which olivine gives an excellent surface.

The sintering point is considerably lowered when the MgO content is lowered. Ferrous oxide (FeO) and



Graph 1 . . . Magnesium silicates in the system MgO-SiO₂ on a weight basis. (Bowen and Andersen)



Graph 2... Equilibrium diagram of the system forsteritetayalite on weight basis. (Bowen and Schairer)

aluminum oxide (Al_2O_3) have the same effect. Thus contamination with silica sand must be avoided in steel foundries. In iron foundries the sintering point is still high enough for ordinary castings, even when the olivine contains much silica sand.

All kinds of binders have been tried, such as bentonites, clays, cement, magnesia, dextrin, etc. Western bentonite has given the best results. Sulphite lye ("Totanin") and dextrin have been added to the bentonite, partly to prevent the sand from drying out before molding, and partly to obtain a hard mold surface after drying. Castings weighing up to 15 tons have been poured in olivine-bentonite molds. Cement can also be used as a binder for olivine sand. Castings have been made in this mixture with the same proportion of cement to olivine as for cement to pure silica sand. At present, castings weighing up to 20 tons are made in cement-bonded olivine sand with satisfactory results.

Olivine sand is treated and mixed in the same way as silica sand. Depending on the grain distribution, additions of bentonite from 2 to 4 per cent and of water from 4 to 6 per cent (by weight) produce the most suitable mixtures for green and dry sand molding. Table 2, mixture 1, gives the properties of sand type 1, Table 1, wth 3 per cent western bentonite mixed and aerated in a mulling mixer.

When sulphite lye and dextrin are added to the facing sand, the bentonite addition is corrspondingly decreased. The Dietert sintering point of these mix-

tures is usually 2640 to 2730 F.

Olivine sand bonded with bentonite is very sensitive to variations in water content. A slight evaporation of water or loss of moisture caused by absorption in bentonite added in dry condition makes the sand feel too dry and reduces its strength. But care should be taken not to add too much water, which will make the sand feel too pasty and make it difficult to work. After some time of practice the molder has no difficulty working in olivine. Olivine is about 25 per cent heavier than silica. The difference in apparent specific gravity due to the different shape of grains of the prepared molding sands, however, is less than 25 per cent. The finer the sand and the better the preparation, the less

the difference. Synthetic olivine sand has not been tried in a slinger.

For cores, approximately the same mixtures are being used as for molds, but with small additions of organic binders. Sand type 1 is in most cases too fine for oil sand cores, though very good results have been obtained with cereal binder and herring oil (used in Norway as a substitute for linseed oil). If very high strength is required, additions of 5 per cent western bentonite, 1½ per cent herring oil and 5 per cent water to sand type 1 give good results (mixture 2, Table 2).

The molds are usually dried. They are then coated with a wash of finely crushed olivine. The best results are obtained by using new olivine sand, adding bentonite, and drying the mold. In producing larger and more complicated castings drying is necessary to ob-

tain a good result.

At present all steel foundries in Norway use olivine as molding material. Plain carbon steel, manganese steel, chromium-nickel steel (18 per cent Cr, 8 per cent Ni), high chromium-low nickel steel (27 per cent Cr, 4 per cent Ni) and low chromium steel (14 per cent Cr) are poured in olivine molds. As previously stated, results are more convincing with alloyed steel than with plain carbon steel. When plain carbon steel is poured in olivine molds, penetration is often observed, especially adjacent to the ingates and where the pattern has sharp corners, where the cores are small as compared to the casting section, and where the casting has a large surface; the larger the casting, the greater the penetration.

For manganese and high chromium steels, the results with olivine molds are decidedly better in these respects. It is beyond doubt that olivine here represents a technical advantage compared to silica sand.

Olivine for Iron Castings

Crushed olivine is used as molding material in iron foundries in producing machinery castings and heavier castings weighing from less than 1 lb up to 8000 lb.

When producing heavier castings, the same considerations for grain sizes and binders must be observed as for olivine in steel foundries. For dried molds which are blacked with graphite for gray iron, addition of

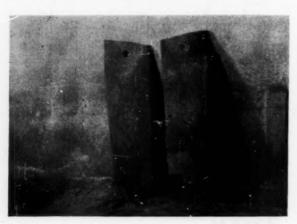


Fig. 1 . . . "Burnt-on" at the ingate of a carbon steel casting 1½-in. thick. Casting at right as stripped, casting at left after 3 minutes blast cleaning.



Fig. 2... Manganese steel crusher jaw. Casting at left as stripped, casting at right after 3 minutes blast cleaning. There is no "burnt-on," no adhering sand.

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coal dust is not necessary. In green molding, coal dust or another substance of similar effect should be used to prevent adherence of sand to the casting. A finer sand Table 1, type 2, for instance) should be used to obtain a smooth surface. The properties of this sand mixed with 3 per cent western bentonite, 4 per cent coal dust (both by weight) and water, are specified in Table 2, sand mixture 3.

It is not necessary to use pure olivine in the production of molds and cores. Olivine can be used together with silica in all proportions, if the refractoriness required is not too high. This makes the introduction of olivine molding sand easier for gray iron machinery castings.

Olivine sand with the necessary addition of bentonite and coal dust gives very good molding material when combined with old sand mainly consisting of silica. After some time, through the addition of pure olivine sand to cover the losses, the molding sand will practically contain olivine only.

At present only a few foundries have adopted the use of olivine in the production of machinery castings and heavier castings. However, crushed olivine has been used experimentally to a great extent in gray iron foundries in the production of hand and machine molded small castings, especially stove plate castings. Experiments have been carried out with many different types, but with pure crushed olivine only. Some of the types utilized are types 3 and 4, Table l. Various bentonites, clays, and organic substances have been used as binders. Experiments have been made with and without coal dust, and with addition of wood flour and similar agents.

These binders and other substances have been used alone or in combination with each other. The experiments have shown that technically the best molding material is obtained when they are added to olivine sand in such proportions and combinations as are usual in preparing synthetic sand based on silica. However, the necessary quantity of water is less. The following examples may be mentioned: Sand type 3 with an addition of 4 per cent clay, 3 per cent coal dust (both by weight), and water (see Table 2, sand mixture 4); and sand type 4 with addition of 1.5

Fig. 3... High-chromium, low nickel steel wheels. This picture was taken before sandblast cleaning treatment. Compare with Fig. 4 at right.

per cent southern bentonite, 3 per cent coal dust (both by weight), and water (see Table 2, sand mixture 5).

In spite of all experiments, no mixture has given a really usable result. The surfaces of the castings are not faultless. Small grooves partly filled with sand start at the ingates and spread inwards on the surface. This spoils the casting, and the specifications for the surface of competitive small castings are not satisfied.

Olivine from green molding has been more difficult to prepare for re-use than molding sand based on silica. This applies to all mixtures tested. The olivine close to the casting is especially hard, having a tendency to bake together. Ordinarily, the greater the necessary addition of binders to make the olivine efficient as a molding material, the easier it is to prepare the olivine again. Larger or smaller additions of fine silica sand are possibly necessary to obtain usable results. This is not a satisfactory alternative. Olivine is meant to prevent silicosis in the foundry industry, and therefore olivine ought to be used in pure condition at all times.

Various parting powders based on olivine have been tested. They have all proved to be less suitable than those which are obtainable in the market. Olivine parting powders are heavier, too.

At present no foundry has adopted the use of olivine in the production of flat, thin castings.

Crushed olivine is used as core sand in iron foundries either alone or together with sand mainly consisting of silica. Depending on the type of cores, sand types of different grain sizes have been tested. Sand types 1 and 5, Table 1, may be mentioned as examples. Various bentonites, clays, linseed oil, sulphite lye, dextrin, and cereal binders have been used. The nature and proportions of binders added to obtain a usable core sand for various purposes partly depend on the grain size. Best results are obtained with approximately the same addition of binders as when the core is made of silica sand. However, the addition of water is less.

Sand type 1, with addition of 1.5 per cent cereal binder, 1 per cent herring oil, and 3 per cent water (all by weight) may be mentioned as an example. The properties of this core sand are specified in Table 2, sand mixture 6. The properties of type 5, with the

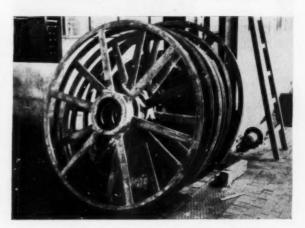


Fig. 4... Same wheels after blast cleaning. Castings are almost free from adhering sand. Cleaning costs are insignificant. Wheels were cast in sand mix 1.

same additions, are specified in Table 2, sand mixture 7.

Compared with silica sand as raw material for cores, olivine sand has several drawbacks worth mention:

1. Olivine sand is heavier. On account of this, the weight of the core will increase considerably.

2. Oil sand cores made from olivine sand have less green strength. The core flows. Stronger reinforcements are necessary if the core can be made at all.

3. The drying of olivine sand cores takes more heat, and a thorough drying takes more time.

4. Olivine sand cores retain the heat much longer.

5. The organic binders in olivine sand cores do not burn to the same extent. When the section thickness of the core is large in proportion to the section thickness of the casting, then the casting is apt to tear over the core when cooling, even if the content of binders has been reduced so much that green strength has reached its lowest limit.

6. It is more difficult to remove olivine sand cores from the castings, especially where the section thickness of the castings is small.

Olivine sand cores have no special advantages as compared to silica sand cores, but in some cases they are equal. In production of castings in which the core is comparatively small, the results are very good. The core is supplied with so much heat that the organic binder will burn, so that there is no difficulty in removing the core. All this explains why olivine sand cores are used only to a small extent in iron foundries.

Where refractory properties are of primary importance (such as for split cores and so-called "Cameron feeder cores"), olivine may be used to great advantage.

Olivine in Non-ferrous Foundries

In non-ferrous foundries, crushed olivine has been tested experimentally on a small scale. Olivine, sand type 3, Table 1, with 2 to 3 per cent bentonite and approximately 4 per cent water gives a satisfactory surface on heavier castings. As regards castings where

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Fig. 5... Stove plate casting $(20 \times 16 \times 1/4)$ in., the ingates on the short sides) poured in olivine sand. The right side is as stripped, left side after cleaning.

a high surface finish is required, no usable results have been obtained. Altogether, so few tests have been made that nothing can be concluded on the use of olivine in non-ferrous foundries.

Summary

Olivine is a highly refractory iron-magnesium silicate. Crushed to a suitable grain size, it is utilized as a molding material in Norwegian foundries, particularly steel foundries. Its thermal properties are very different from those of silica. The main reason for using olivine is the presumption that the risk of developing silicosis is diminished. It is a technical advantage to use olivine for the manufacture of manganese steel and chromium-nickel steel. The results in iron foundries are not convincing.

References

1. N. L. Bowen and Olaf Andersen, American Journal Sci., 37,487 (1924).

 N. L. Bowen and Schairer, American Journal Sci., 29 (2) 151-217 (1935).

3. J. Ringstad, "Synthetic Olivine Molding Sand for Steel and Iron Castings at A/S Drammens Jern," *Teknisk Ukeblad*, 1939, (38), p.493.

4. Dagfinn Elstad and Kristoffer Stenvik, "Prophylaxis of Silicosis in Foundries by means of Olivin Rock as Artificial Molding Sand, Oslo 1946." Publication No. 27 from Statens Rastofflaboratorium in 1946.

5. E. J. King, Nancy Rogers, Margaret Gilchrist, W. M. Goldschmidt and G. Nagel Schmidt, "The Effect of Olivin on the Lungs of Rats." *Journal of Pathology and Bacteriology*, Vol. LVII, No. 4 (1945), pp. 488-491.

Bibliography

Australian Patent No. 111508. British Patent No. 570719.

W. M. Goldschmidt, Teknisk Ukeblad, 1940, pp. 1-7 and Teknisk Ukeblad, 1946, pp. 685-687.

K. J. Stenvik, Journal of The Iron and Steel Institute, 1949, Vol. 162, Part 1, pp. 44-48.

Gilbert S, Schaller and W. A. Snyder, Foundry, 1951, Vol. 79, No. 5, pp. 104-105.

Gilbert S. Schaller and A. M. Nilson, Foundry, 1951, Vol. 79, No. 10, pp. 110-111.

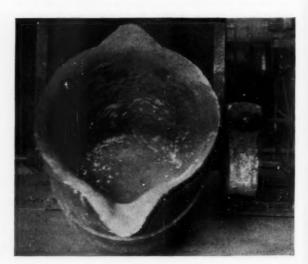


Fig. 6... Olivine mixed with 4 per cent bentonite is used for ladle lining. This ladle has been used five times for cast iron with no repairing.

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DISCUSSION

Chairman: C. C. Sigerfoos, Michigan State College, E. Lansing. Co-Chairman: V. M. Rowell, Archer-Daniels-Midland Co., Ceveland.

Recorder: F. P. GOETTMAN, Standard Sand Co., Grand Haven, Mich.

[RVING WISE: 1 Have you found the grain size you indicate to be the most desirable?

MR. SISSENER: No. More work must be done on sizing. The Norwegian government operates the mine and that is the only size available.

G. S. SCHALLER: 2 You could readily build a separator to obtain various sizes.

H. F. TAYLOR: 8 One should not expect olivine to greatly influence solidification and shrinkage. An advantage for olivine sand is that it is crushed and sized and theoretically, one can obtain any desired specific sizing which, when properly bonded,

can be used for special uses.

V. M. Rowell: 'If silica is introduced at the sand metal interface will it react with the iron oxide on the surface of the metal to form more fayalite and lower the refractoriness?

MR. SCHALLER: I doubt that such a reaction would occur.

Olivine Products Corp., New York.
 University of Washington, Seattle.
 Massachusetts Institute of Technology, Cambridge.
 Archer-Daniels-Midland Co., Cleveland,

METALLURGICAL BLAST CUPOLA

By

R. Doat* and M. A. DeBock**

Reamur in 1722 first mentioned the existence in France of itinerant iron melters who used, for melting purposes, a small furnace, about 20 in. high. It consisted of a pot operating as a crucible and having a shaft on top. They melted scrap iron using charcoal as fuel. Walking from one place to another, these artisans cast pots and pans, and when the requirements of their customers were fulfilled they dismantled the plant and went elsewhere.

At the end of the 18th century cupolas were stationary units 18 to 24 in. high, equipped with a tap hole. Wilkinson, at the beginning of the 19th century, started to use coke and this new melting process spread quickly.

Since then many inventors and foundrymen have tried to improve the cupola, the main object having been, in most cases, increase of the overall thermal efficiency.

We should like to mention that a hot blast cupola was already used in a gun foundry in Liége in 1838; another cupola of the same type was in operation in Germany in 1850² and one can still see this unit in Luisenhütte at Wocklum, near Balve, Germany.

We must admit that even with the up-to-date hot blast cupolas, foundrymen still have difficulty with oxidation of metallic elements and melting of refractories which result in lack of slag control and, consequently of the metallurgical reactions.

The M.B.C. has not been developed from the old Reaumur or the Wilkinson cupola but from the slagging gas producer.

Philipon, a French engineer, found that in slagging gas producers a certain amount of iron (approximately 8 pct) charged with the limestone, favorably promoted tapping and that use of a coke with a high ash content (25 to 30 pct) was quite possible. Progressively raising the amount of metal charged, Philipon obtained at the same time a gas of good calorific value and a cast iron having interesting qualities. Some literature concerning Philipon's process can be found in the *Revue de l'Industrie Minérale*³ and in Dr. Gumz⁴ recent publication on gas producers and blast furnaces.

In 1947, during a severe pig iron shortage in Belgium, the Compagnie Générale des Conduites d'Eau at Liége, which operates a large pipe spinning foundry in Belgium, decided to elaborate on Philipon's principle, the aim being production of good quality gray iron for pipe spinning, from charges consisting of 90 pct and even 100 pct mild steel scrap.

After a development period of several months the plant, which has a capacity of 4 to 5 tons per hour, worked satisfactorily and as shown on the production chart (Fig. 1) the monthly production continuously increased for two years on a full industrial scale.

During the experimental and development periods, all our efforts were constantly directed toward metal-lurgical improvements, having left for the time being, the problem of thermal efficiency, which has been solved satisfactorily as it can be deducted from the heat balance as discussed further on.

This means that, instead of attempting to increase the CO₂ content of the gases, which is the case in the conventional cold or hot blast cupolas, we have always tried to have the lowest possible CO₂ content.

In conventional cupolas the gases have a high CO₂ content, which originates oxidation of the charge. This disadvantage is still more important in the melting zone, the surface of contact of the molten iron being larger.

The iron runs down in droplets, meeting the rising gases at such a high temperature, that all chemical reactions take place very quickly.

Jungbluth⁵ has proven that the melting losses, calculated from FeO content of the slag, increase proportionately with combustion efficiency, which is given by the formula:

$$\% \text{ Efficiency} = \frac{\text{CO}_2}{\text{CO} + \text{CO}_2}$$

Compagnie Générale des Conduites d'Eau, Liége, Belgium, and ** Consulting Engineer, Brussels, Belgium.

This is an Official Exchange Paper from the Association Technique de Fonderie de Belgique to the American Foundrymen's Society presented at the International A.F.S. Foundry Congress in Atlantic City, May 5, 1952. The paper presents some results obtained with a 4-5-ton Metallurgical Blast Cupola (abbreviated M.B.C. in the paper) plant operating at Liége, Belgium.

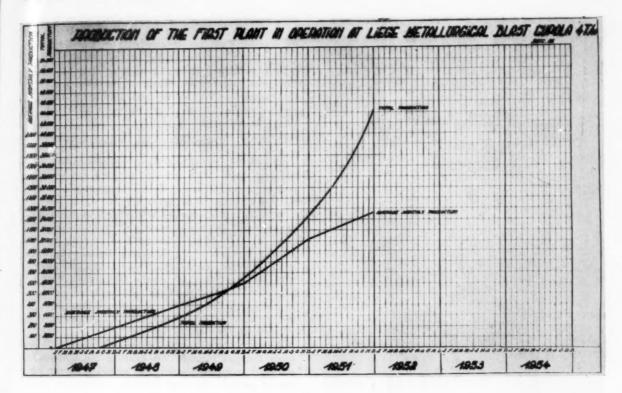


Fig. 1

Some cupolas, as mentioned by Longden, are run under such conditions that the CO in the combustion zone is high, the CO being burned to CO₂ by introduction of air above the combustion zone. Gumz suggested introducing flue gases from the combustion of blast furnace gas (waste gases having a high CO₂ content) instead of air. But Prof. Jungbluth's statement remains valid for cold and hot blast cupolas.

The M.B.C. operates with a reducing atmosphere in the combustion zone as well as in the melting zone. We should like to point out that when mentioning a reducing atmosphere, we mean an atmosphere which reduces SiO_2 under prevailing conditions. A constant stream of iron droplets falls through a constant depth of slag at high temperature, of constant composition.

At high temperature (approximately 2780 F) oxygen in presence of C and Si reacts with the carbon and not with the silicon. From experience we know that with the M.B.C. there is no melting loss from oxidation of iron, silicon or manganese. The tuyeres have a special design and are projected into the body (Fig. 2). The blast is preheated at a temperature approximately 930 F and kept constant within close limits.

When examining the problem of the slag, it is well known that there exists no difficulty in obtaining a slag having good fluidity by adding to the metal charges appropriate quantities of flux, e.g., limestone. The amounts of limestone required can be estimated from the figure given in the HANDBOOK OF CUPOLA OPERATION? for the following purposes:

1. Oxidation Reactions. In average cupola prac-

tice, with a soft iron mixture (about 2.75 Si, approximately 20 lb of limestone are required to form calcium silicate.

- 2. Coke Ash. An equal weight of limestone to the ash content is required to combine with the ash.
- 3. Cupola Lining. Normally 10 to 15 lb of limestone per ton of metal charge to combine with the lining coming down during the melt, but depending upon the type of metal charge and temperature.
- 4. Sand and Dirt. In average practice of ordinary gray iron melting approxmately 20 lb of limestone for making fluid slag.

With the Metallurgical Blast Cupola, there is practically no limestone required for oxidation reactions.

For coke ash, more limestone is necessary, the consumption of coke being higher.

For cupola lining materials, the amount of limestone is extremely low; the water-cooled shell, and the thin special lining preventing the wear of the refractory material.

For sand and dirt, obviously, the quantities remain identical when charging the same materials.

Attempts have already been made to prevent the wear of the lining by using basic refractory or by water-cooling the shell, but in both methods the oxidation of Si, Mn and Fe is not prevented when the cupola operates in an oxidizing atmosphere.

We may add that in basic cold blast cupolas the reactions between metal and slag are confined within certain limits, the basic slag having a higher melting point and a lower fluidity. The high temperature of slag and iron, reached in the Metallurgical Blast

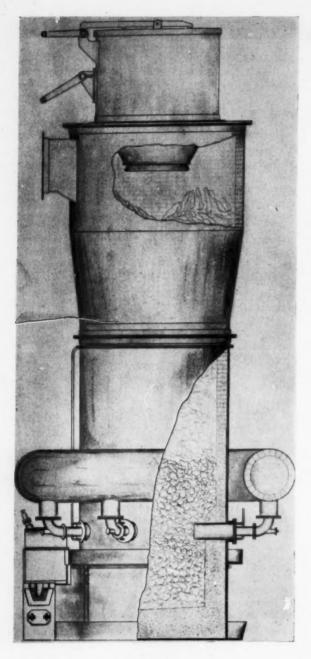


Fig. 2

Cupola promotes desulphurization as well as FeO reduction.

The slag is not contaminated either by melted refractory material or by oxidized metals, two factors on which no control exists in conventional cupolas, thus ensuring a slag having a constant composition during the whole melting period.

The constant stream of iron droplets falling through a constant depth of slag of constant composition results in conditions which, for all practical purposes, are consistent with chemical equilibrium.

One of the important advantages of the M.B.C. is

that it is possible, within a very short time, to alter the composition of the metal running out the spou, by operating on the composition of the slag onl, going over from basic to acid, or vice versa.

Even from iron charges consisting of 100 pct mil-i steel scrap, the M.B.C. gives the possibility to vary the carbon content of the iron within the range of approximately 2.8 to 4.2 pct always with a low sulphur content as shown later.

Description of the Plant

The M.B.C. at the first glance appears to be a combination small blast furnace, slagging gas producer and a cupola. Figure 3 shows the general layout of an M.B.C. plant. The top of the M.B.C. is closed by a bell-shaped lid which is removed automatically, preparatory to introduction of the charge. The gases are collected at the top and pass through insulated pipes first to the dust collector (2) where the heavy dust particles are eliminated and after to a gas cleaner (3) in which the cleaning of the gases is performed, this in order to prevent the pollution of the atmosphere and the dust accumulation in the air preheater. The exhaust fan (4) compensates the draught loss in the gas cleaner.

A portion of the clean gas (approximately 40 pct) feeds the gas burners (5) heating the combustion chamber; auxiliary burners are provided for the starting up periods. In some cases the burners are of the combined type and can be fed by fuel oil for starting up the plant while normally fed by gas. The fan (6) provides the necessary air for the combustion of the gas. The exhaust fan (7) takes the flue gases from the air heater and blows them into the atmosphere. The blower (8) blows the blast into the air heater in which the temperature of the air is preheated to 930 F. From the outlet of the air heater the blast is led to the tuyeres.

The water-cooling of the shell of the cupola and that of the tuyeres is performed by a water pump (9). A continuous slagging spout (10) is provided and ensures a constant height of slag in the well. The separated slag and the cooling water are drained together, the slag being granulated and becoming a recuperable by-product.

Several instruments such as air-flow meters, indicating control pyrometers, pressure and suction indicators, CO/CO₂ analyzers as well as apparatus ensuring the automatic regulation are provided as shown on the instrument panel (11), Fig. 3.

Heat Balance

Experiments carried out at Liége on the 4-ton per hr cupola have shown the following results (Fig. 4):

Per metric ton of metal charged:

Coke consumption-198 kg (436 lb).

Analysis of coke-Volatile matter, 1 pct; Moisture, 2 pct; Ash, 7 pct; Net calorific value, 7,200 cal.

Temperature of molten iron at spout-1540 C (2800 F).

Slag-55 kg (121 lb).

Temperature of gases, top of cupola—300 C (570 F). Analysis of gases (dry)—CO₂, 4 pct; CO, 26.6 pct; H₂, 1 pct.

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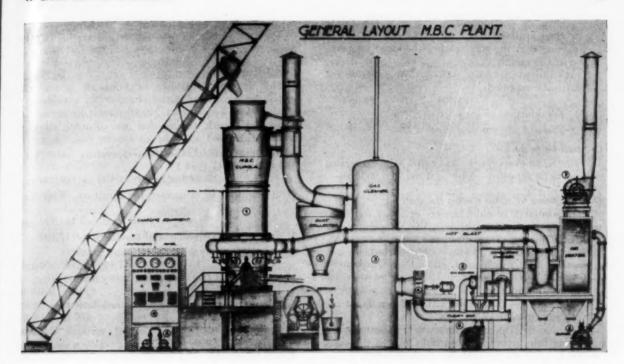


Fig. 3

Volume of air blast-760 cu meters/hr (SCM at O C, 760 mm).

Moisture in air-11 gr/cu meter (SCM).

Atmospheric pressure-760 mm Hg.

Non-metallic charge-Limestone, 26 kg (57 lb); Thomas slag (8 pct P), 25 kg (55 lb).

Metal charge—Mild steel scrap, 964 kg (2121 lb); Analysis—0.05 C, 0.20 Si, 0.60 Mn, 0.10 P, 0.04 S; FeSi (50 pct), 36 kg (79 lb).

Analysis of cast iron-3.3 C, 2.0 Si, 0.6 Mn, 0.30 P, 0.05 S.

For every 1000 kg (2200 lb) of metal charged, at the spout we get:

1000 kg (2200 lb) of metal

28 kg (62 lb) of C of recarburization reincorporated

2 kg (4.4 lb) of P reduced from Thomas slag which means a total of 1,030 kg (2266 lb) of cast iron.

Heat Input

Coke—Taking into account the carbon reincorporated into the metal remains for heating purposes:

198 — 32 = 166 kg (365 lb) of coke giving 143.5 kg (316 lb) C.

- 1. Calorific value of coke $166 \times 7200 = 1,195,000$ cal.
- 2. Heat from hot blast air at 500 C (930 F) 686 \times 160 = 109,800 cal.
- Heat developed by combustion of coke:
 12.7 pct of the carbon in coke is burned to CO₂
 87.3 pct of the carbon in coke is burned to CO

Combustion reaction

$$C + O_2 = CO_2 + 8074 \text{ cal/kg}$$

 $C + \frac{1}{2}O_2 = CO + 2414 \text{ cal/kg}$

Heat developed

$$CO_2$$
—143.5 \times 0.127 \times 8074 = 147,090 CO —143.5 \times 0.873 \times 2414 = 302,210

4. Since the M.B.C. operates with a reducing atmosphere, there is no heat developed by oxidation of silicon or manganese.

Heat Output

A. Metal

Temperature of metal charges—20 C (68 F) Melting temperature—1510 C (2750 F) Final (pouring) temperature—1540 C (2800 F)

Heat Required

(a) In solid state steel scrap-

$$1004 \times 0.207 \times 1490 = 309,700$$
 cal

- C and P- $30 \times 0.435 \times 1520 = 19,800$ cal
- (b) Heat of fusion Steel scrap—

$$1004 \times 65.6 = 65,900 \text{ cal}$$

(c) Final (pouring) temperature—

$$1004 \times 0.23 \times 30 = 6,900 \text{ cal}$$

B. Slag

- (1) Calcination of limestone —Wght. of $CaCO_3=26$ kg (57 lb) Heat of decomposition — $26\times438=11,390$ cal.
- (2) Formation and melting of slag
 Heat of reaction, per kg CaCO₃—516 cal/kg

Heat of decomposition " -438 " " Heat available $26 \times 78 = 2,028$ " "

Assuming that the slag reaches the same temperature as the metal:

Heat required $-55 \times 0.28 \times 1520 = 23,408$ cal Heat to be supplied

-23,408-2,028 = 21,380 "

(3) Total heat to be supplied -11,390+21,380 = 32,770 cal

C. Moisture in the air

Water to be evaporated— $11 \times 760 = 8,360 \,$ grams Heat to be supplied— $8360 \times 0.586 = 4,900 \,$ cal

D. Heat losses by water cooling the shell

Approximately 10 cubic meters of water has been used per ton of metal charged, the temperature of which having risen from 20 C (68 F) to 30 C (85 F). Heat losses- $10,000 \times (30 - 20) = 100,000$ cal

E. Gases at Top of Cupola

Composition of gases

- CO₂, 4 pct; CO, 26.6 pct; H₂, 1 pct

Net calorific value

- 795 cal/cu meter at O C, 760 mm Hg

Taking into account the sensible heat of the gases at 300 C (570 F) the calorific content is 889 cal/cu meter.

Volume of gases per ton of metal charge - 807 cu meters/hr (SCM) $895 \times 807 = 717,400 \text{ cal}$

F. Air Heater

The theoretical volume of blast air is 686 cu meters/hr per ton of metal charged but considering the losses we assume that 760 cu meters/hr are to be preheated, which means that we consider all the leakages into and after the air heater.

Heat required to raise the temperature from 20 C (68 F) to 510 C (950 F)

 $---760 \times 0.321 \times 490 = 120,000$ cal

Heat required to evaporate the moisture-4,900 cal

Total heat-124,900 cal

Heat losses by radiation and conduction—17,100 cal Heat to be supplied by the effluent gases—

124,900 + 17,100 = 142,000 cal

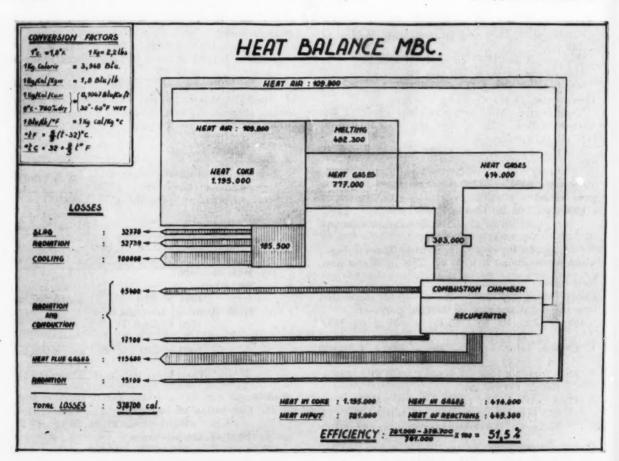
Temperature at the inlet of the air heater-

850 C (1560) F

Temperature at the outlet of the air heater-

400 C (750 F)

Volume of effluent gases passing through the air heater— $142,000/450 \times 0.371 = 850$ cu meters/hr Losses by sensible heat in effluent gases at 400 C (750 F)— $850 \times 136 = 115,600$ cal



Heat to be supplied at the inlet of the air heater-142,000 + 115,600 = 257,600 cal

Heat to be supplied at the inlet of the combustion chamber assuming a combustion efficiency of 85 pct-257,600/0.85 = 303,000 cal

Volume of gases at 889 cal/cu meter for heating the 303.000/889 = 341 cu meters which means $(341/807) \times 100 = 42.2$ pct of the total gases.

Heat available in gases-

717,000 - 303,000 = 414,000 cal

Overall efficiency =

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Heat absorbed by iron

Heat content of coke-heat content of top gases 402,300 = 51.5 pct

1.195,000 - 414,000

In the plant considered, the temperature of the gases at the inlet of the air preheater has been brought down to 850 C (1560 F) by mixing fresh air. In case the top gases remaining available are used, a more economical solution would be to recycle a portion of the flue gases from the outlet of the air heater into the mixing chamber before the air heater resulting in a substantial increase of above-mentioned efficiency. We may add that the temperature of 400 C (750 F) of the flue gases at the outlet of the air heater represents a maximum, this temperature being between 300 C and 400 C (570 F and 750 F).

When considering the coke consumption only, one would judge at first glance that the overall efficiency of the M.B.C. plant must be lower than those reached with conventional hot blast cupolas. Above results do show that this is not correct, the efficiencies of hot blast cupolas being between 40 and 50 pct, depending upon the CO/CO2 contents of the top gases and the temperature of the preheated blast air whereas in cold blast well operated cupolas, efficiencies of 30 to 35 pct are attained.

The high coke consumption is due to the fact that the charges consist of 100 pct steel scrap and it is well known that the specific heat and the heat of fusion are higher for steel scrap than for pig iron and cast iron scrap. On the other hand the temperature of the cast iron is higher in the M.B.C. approximately 1540 C (2800 F) whereas 1470 C (2680 F) in the con-

ventional cupolas.

Figure 4 shows the heat disposition and distribution of the losses. It may be objected that in conventional hot blast cupolas the top gases still have a latent and sensible heat content, but from a practical point of view it does not seem well to clean the gases in such a way that they can be led and utilized at a certain distance from the cupola. Therefore the top gases, not used for preheating the blast, are considered waste gases.

It substantially pays to install a gas cleaner for cleaning gases having a heat content of 889 cal/cu meter whereas it is debatable for gases having a heat content of roughly 400 cal/cu meter. The gas cleaner simultaneously solves the delicate problem of dust accumulation in the air preheater.

To summarize, it should be noted that the M.B.C. operates as a combination cupola and slagging gas producer. Some examples of application of the M. B.C. are as follows:

1. Cast iron for spinning pipe foundry. The M.B. C. plant installed at Liége has melted as of Jan. 1952 more than 45,000 metric tons of cast iron. The average of the charges is approximately as follows:

Metallic charges-Steel scrap - 65 pct Iron scrap - 41.5 pct

Ferrosilicon - 3.5 pct Non-metallic charges (in per cent of iron melted)-

> Coke - 17.5 pct Limestone - 2.6 pct

Thomas slag (8 pct P) - 2.3 pct

Producing a cast iron with the following average analysis-C-3.53, Si-2.35, Mn-0.45, P-0.55, S-0.55. while the average analysis of the slag has been-CaO₃ -44.0, SiO₂-36.0, Al₂O₃-15.0, CaS-3.5, MnO-0.70, FeO-0.80. The quantity of slag was approximately 150 lb per ton of metal melted. Consumption of refractory in the lining and front slagging spout has been approximately 2.2 lb of graphite-base compound per ton of cast iron.

2. Low sulphur iron. Sam Carter8 and E. Renshaw⁹ have published some experimental results on the subject of desulphurization of basic slags. The best results, with regard to desulphurization, reported to these authors, are shown in Table 1, to which we added two compositions obtained with the Metallurgical Blast Cupola slags at Liége.

We have calculated the ratio D. = Su - Sr for

each of the compositions reported. In this formula: S_M = Sulphur charged in metal (Column 8, Table 1) $S_r = Final sulphur in cast iron (Columns 9 and 10)$ The ratio D_e shows the desulphurizing effect (Columns 11 and 12). When comparing the results of

TABLE 1-DESULPHURIZATION

Test No.	SiO ₂	CaO	MgO	MgO	CaO+MgO	Al ₂ O ₃	Total	S_{M}	Sp		D.	%
			0	Equiv.	Equiv.				Aver.	Min.	Aver.	Min
	1	2	3	4	5	6	7	8	9	10	11	12
Carter 513	22.6	53	7.8	10.8	63.8	9.0	95.4	0.050	0.032	0.015	36.0	70.0
	23.7	-	_	-	66.9	9.4	100.0	_	_	_	_	_
Renshaw	32.0	44	10.0	14.0	58.0	9.0	99.0	0.075	_	0.030	-	60
	32.3	-	_	-	58.6	9.1	100.0	_	_	_	-	_
M.B.C. LIEGE	34.5	49	_	_	49.0	12.5	96.0	0.050	0.024	0.017	52.0	66
No. 4	36.0	51	-	-	51.0	13.0	100.0	_		_	_	_
M.B.C. LIEGE	33.0	48.5	-	_	48.5	15.5	97.0	0.070	0.050	0.036	28.6	48.5
No. 8	34.0	50.0	_	-	50.0	16.0	100.0	-		_	_	_

the desulphurization of the basic slags with those of the M.B.C. slags, we came to the following conclusions:

 The results of M.B.C. slags compare favorably with those of basic slags.

2. The quantity of sulphur taken off per pound of slag must be more important in the M.B.C., for the following reasons:

Per ton of iron melted:

(a) The amount of sulphur arising from coke is higher, the coke consumption being higher.

(b) The quantity of slag is lower, which means a higher concentration.

This favorable desulphurizing effect reached with the M.B.C. is the result of a lower FeO content in the slag, and the high temperature of same.

R. Rocca, W. J. Grant and J. Chipman¹⁰ and C. W. Pfannenschmidt¹¹ have shown the bad effect of the FeO in the slag on desulphurization. Furthermore it is easier to produce a basic slag with the Metallurgical Blast Cupola than with the conventional cupola even with basic lining. Renshaw and Sargood¹² have shown that, without lining wear, as in a basic cupola.

the ratio CaO/SiO₂ increases from 0.56 to 1.57. When making the calculations from the figures given by these authors it is easy to demonstrate that preventing oxidation of silicon, as in the M.B.C., this ratio increases from 1.57 to 2.6.

It may be of some interest to show on the ternary diagram of basic slags the points corresponding to above-considered M.B.C. slag compositions. Referring to Fig. 5 it appears from the position of points 4 and 8 that they are near the area in which basic slags operate. It should be noted that, on the ternary diagram, it has been necessary to consider the four components: SiO₂, Al₂O₃, CaO and MgO. It seems logical to group the strong bases CaO and MgO and to convert the MgO into CaO equivalent (40 gr MgO = 56 gr CaO). From above considerations it appears that it is possible to obtain with the M.B.C. cast irons having very low sulphur contents with slags being far less calcareous, thus having better fluidity than in basic cupolas.

The benefit of obtaining economically low-sulphur iron is most interesting for the production of nodular iron, and for steel making in a convertor where

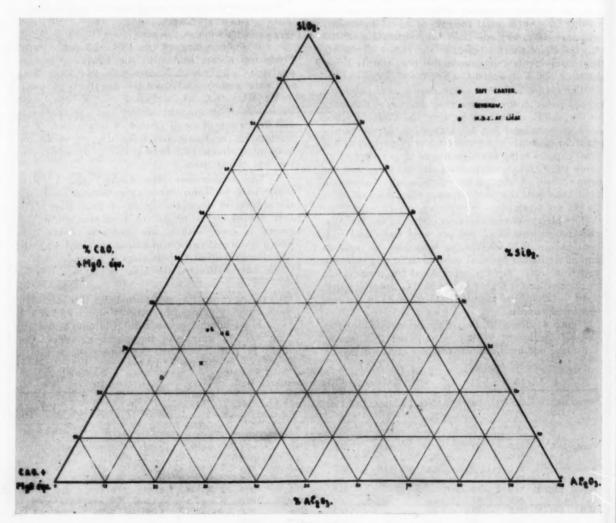


Fig. 5

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the low sulphur and phosphorus content of the iron, produced from all steel scrap charges, is a special advantage.

- 3. Ingot molds. Experimental castings have been made from charges consisting of approximately 100 pet steel scrap and a small quantity of ferrosilicon. The cast iron had a carbon content of 4 pct and the coke consumption was 20 pct. This metal is most suitable for ingot molds.
- 4. Alloy cast iron. The reducing temperature of the M.B.C. results in the possibility of melting alloys such as nickel and chromium with no melting loss.

Advantages

1. Metallurgical control of reactions. By adjusting the slag analysis, alterations and corrections on the charge are possible. The metal poured at the tap hole can definitely be different from the metal charged at the top, while the conventional cupolas produce at the best a cast iron having an equal quality to the average of the material charged. For instance cast iron with a high carbon content can be produced from all steel scrap charges, as well as low carbon cast iron from high carbon charges.

When taking into account the difference in prices, existing on some markets, between cast iron scrap, steel scrap and pig iron it is sometimes possible to amortize an M.B.C. within a few months, and in any case to be independent of pig iron in case of short-

age of this material.

2. The melt is continuous or intermittent over 5day periods, according to production requirements. The cupola operates 8 or 16 hr daily with no patching being necessary. The cupola can be kept banked for a period of 40 hr, and then a constant analysis metal can be obtained from beginning of remelting. Due to the above features, one M.B.C. is able to perform the duty of two conventional cupolas.

3. The temperature of the iron is very high, approximately 2800 F; it is completely deoxidized and of outstanding quality. Without special additions, the sulphur can be held under 0.06 pct, granting to melt iron with a low manganese content, less than 0.5 pct. Increased fluidity, decrease in scrap, an improvement in the melting operation and a substan-

tial increase in production result.

4. As it can be deducted from the calculations of the heat balance, the overall thermal efficiency is somewhat higher than in conventional hot blast cupolas, when taking into account the elimination of the melting losses. Even when no use can be made of the effluent gases the thermal efficiency becomes 33.7 pct which is still somewhat higher than that in cold blast cupolas.

5. There is no melting loss from oxidation of the iron, silicon or manganese in the cupola. The melting losses in foundry practice, consist of three differ-

ent losses:

(a) The material being paid for as iron may contain adhering sand (sand-cast pigs), iron oxides and other undefined dirt found in the scrap pile.

(b) The real melting loss resulting from oxidation of iron, silicon and manganese in the cupola.

(c) Iron lost and taken out with the old sand. The sum of these three items can be found by subtracting the weight of the good castings produced from the iron charges during a given period, taking into consideration the weight of the returns (gates, risers, scrap castings). From experience we can state that, for instance, with charges consisting of approximately 75 pct steel scrap, 5 pct FeSi, the balance being cast iron scrap, no iron loss will be found. The reason for this is that the iron oxide of the charge is dissolved in the slag and recovered by reduction to iron; the oxidation of iron, silicon and manganese is practically nil owing to the reducing atmosphere and the carbon pickup ranging from 0.10 up to approximately 3.6 on 75 pct of the charge, fully compensates for the dirt and the iron lost with sand.

6. The cupola can use coke of low quality with

high sulphur and ash content.

7. The cupola may be run with an acid or a basic slag with no difference in refractory consumption, which, in both cases, is extremely low (2 lb per ton of metal melted).

8. The slag, granulated by the cooling water running from the cupola, is turned into a saleable byproduct with low FeO content which can be used for cement making.

9. One hour melting production is held in a receiver at a temperature of 2800 F without heating

(ladle preheated before starting).

10. Due to the fact that the gases are cleaned before the air heater the delicate problem of atmospheric pollution is also solved.

The cost of this type of plant is obviously higher than that of a conventional cupola, but the above enumerated advantages fully pay for the extra cost.

References

1. M. N. Briavoinne, "Sur les inventions et perfectionnements dans l'industrie depuis la fin du XVIIIe. siécle jusqu'à nos jours. Bruxelles 1838, p. 142.

2. E. Kosack, "Die Luisenhutte in Wocklum, ein technisches

2. E. Kosack, "Die Luisenhutte in Wocklum, ein technische Kulturdenkmal," Westfalen, vol. 24, mo. 6, pp. 250-256 (1939).

3. Revue de l'Industrie Minérale (1934).

4. Wilhelm Gumz, Gas Producers and Blast Furnaces, John

Wiley & Sons, Inc., New York (1950).
5. H. Jungbluth, "Der Abbrand beim Schmelzen im Kupolofen und der Aufbau der Kupolofenschlachten," Die Neue Giesserei, technisch-wissenschaftliche Beihefte I, 1949-50, 11-16. 6. E. Longden, Institute of British Foundrymen, June, 1946,

T. P. 85.

7. HANDBOOK OF CUPOLA OPERATION, American Foundrymen's Society, p. 311 (1946)

8. S. Carter, "Basic-Lined Cupola for Iron Melting," Trans-ACTIONS, A.F.S.,, vol. 58, p. 376 (1950).

9. E. Renshaw, "Basic Cupola Melting and its Possibilities,"

Transactions, A.F.S., vol. 59, p. 20 (1951). 10. R. Rocca, N. J. Grant and J. Chipman, "Distribution of Sulphur between Liquid Iron and Slags of Low Iron Oxides Con-

centration," Journal of Metals, AIME, p. 319, April 1951. 11. C. W. Pfannenschmidt, "L'elimination du soufre au cubilot et au four electrique," Congres International de Founderie, Bruxelles, 1951, p. 468.

12. E. S. Renshaw and S. J. Sargood, Foundry Trade Journal, vol. 87, no. 1728, Oct. 13, 1949.

DISCUSSION

Chairman: J. D. SHELEY, The Black Clawson Co., Hamilton, Ohio.

Co-Chairman: R. A. CLARK, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

Recorder: W. T. BOURKE, American Brake Shoe Co., Mahwah, N. L.

S. F. Carter (Written Discussion): 1 The authors have reported some very progressive developments in cupola operation which are commendable.

The cupola described is a basic cupola since the slags obtained have the basicity sought in basic operation, and accomplish the chemical objectives. One of the greatest influences on the slag is the refractory lining consumed. The usual approach is to substitute basic refractories to obtain slag basicity. The authors, following another approach, have used water cooling and carbon refractories to the point of reducing refractory consumption to a negligible effect on slag chemistry. Reduced oxidation from the high temperature blast also promotes slag basicity. This seems to be a "defensive" basic cupola, from the refractory standpoint, where the production of acid consutuents is restrained, in contrast to a cupola with a basic lining in which an "offensive" effort is made to add bases faster than acids are produced.

I have a few questions. Is water circulation continued while the cupola is shut down at night and over week ends?

Has any effort been made to investigate the effect of higher blast volumes and higher internal pressures, as tried in some blast furnaces? The enclosed top should retard the high velocity and usual troubles of overblowing and possibly permit increased output if thermal efficiency is not sacrificed.

Has any effort been made to operate the MBC without any gas production? If so, can the coke be reduced and the blast temperature lowered to the point where all the exit gas is consumed in pre-heating the blast? Could better efficiency be obtained, purely from the melting standpoint, by a plant not desiring gas production?

MR. DOAT: Water cooling was shut off during long shutdown periods (e.g. overnight), but the water cooling in the

tuyeres was left on. Higher pressures have not been tried; windbox pressure is about 30 in. of water. No attempt has been made to reduce coke, as it was felt that this would defeat the purpose of striving for a reducing atmosphere.

MARIO OLIVO: ² The most important feature of the Metallurgical Blast Cupola was that it was possible within a very short time to alter the composition of the metal by altering the composition of the slag from acid to basic or vice versa. This is a revolutionary development about which I would like further explanation.

MR. Doat: Since oxidizing conditions were not present there was no contamination by SiO₂ from the lining. We operate with a very deep layer of slag (approximately 15 in.) in the cupola, thus the reactions have more time to take place.

Mr. OLIVO: How long does it take for the slag to change from acid to basic?

MR. DOAT: This can be done in the same melt and requires only the time necessary for the flux to be charged to the melt. GOSTA VENNERHOLM: 3 The authors are to be congratulated on the fine work they have done on the Metallurgical Blast Cupola. Work on water-cooled cupolas, both acid and basic, is necessary if progress is to be made in the foundry. The cupola is the most neglected equipment in the foundry, the average cupola operating or. every second day. From our experience we find that refractory consumption was reduced from 56.3 lb per ton to 10 lb per ton with water cooling. Cooling tubes placed conically caused a Venturi effect resulting in an excessively high melting zone. We found 70 per cent iron oxide in the slag half way up the cupola due to cold air passing next to the wall. Projecting water-cooled tuyeres as used by Levi and Doat is a step in the right direction to overcome this condition. We operate the water coolers without refractory covering because the refractory lasts only about an hour and is therefore of no value. It would be possible to operate the cupola 16 hr daily for 6 months using 3.2 lb of refractory per ton of iron.

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A STUDY OF THE A.F.S FINENESS TEST

By

H. A. Stephens*

ABSTRACT

This paper is a critical survey of grading analysis for the A.F.S. fineness test. An examination of the various sources of error in sieving throws emphasis on the effect of oversize apertures, and on the importance of the duration of sieving. The errors due to sampling and sieving are examined and found to be of the same order. The correct time for a number of sieving machines is determined. Grading analyses on different sieve sets show apparently large discrepancies in the results for comparable sieves, even though the fineness numbers do not differ seriously; these discrepancies are shown to be due to slight differences in the effective apertures of the sieves. A method is developed for calibrating unkown sieves using standard sands under standard conditions. The A.F.S. fineness number is discussed and a new distribution index is proposed.

1. Introduction

The American Foundrymen's Society grain fineness test to determine the grain characteristics (texture) of a molding sand is highly important, since texture influences most other properties of the sand as well as the surface of the casting.

Considered broadly the test may be divided into three main sections:

i) Removal and estimation of "A.F.S. Clay."

ii) Determination of the size distribution of the "sand" grade by a grading analysis.

iii) Classification of results (determination of the A.F.S. grain fineness number, etc.).

The A.F.S. clay is defined as "that earthy portion of a foundry sand which, when suspended in water, fails to settle 1 in./min and which consists of particles less than 20 microns (0.02 mm) in diameter." The term "sand" grade is used for the material remaining after the removal of A.F.S. clay.

The A.F.S. method involves sedimentation of a 50-gram sample of the defloculated molding sand and the removal of particles failing to settle at the specified rate, followed by a grading analysis by sieving on standard sieves. The A.F.S. grain fineness number is the weighted average of the mesh sizes of the par-

ticles, i.e., the approximate mesh of the theoretical sieve through which the sample would just pass if its grains were of uniform size. Table 1 sets out the nominal apertures of the sieves specified by the A.F.S., the factors used and a typical calculation of a fineness number.

TABLE 1—DATA FOR CALCULATING A.F.S. GRAIN FINENESS NUMBERS

			Fineness Ca	lculation
ASTM Sieve No.	Nominal Aperture, mm	Factor	Material Remaining on Sieve, %	Product
6	3.36	3	0.0	0
12	1.68	5	0.6	3
20	0.84	10	3.6	36
30	0.59	20	8.7	174
40	0.42	30	27.4	822
50	0.297	40	21.1	844
70	0.210	50	8.9	445
100	0.149	70	4.2	294
140	0.105	100	1.8	180
200	0.074	140	1.2	168
270	0.053	200	0.4	80
Pan		300	2.2	660
Total	Percentage of	Sand Grade	e 80.1	
	Product			3,706
A.F.S. Grain	Fineness No.	Total p	roduct =	3706 — = 4
		Total pe	rcentage	80.1

Determination of A.F.S. clay has been surveyed critically by a number of authors, notably Haseman,² who discussed the effect of different electrolytes on deflocculation, and Hills,³ who studied different methods of estimating clay substance. More recently hydrometer and pipette methods have been studied^{4,5} to examine the "silt" fraction and to study the effect of silt on the properties of molding sand.

However, the strength of a molding sand depends not only on the amount but also the type of clay, so that, as a further step, quick routine methods of identifying clay types are needed. Clay substance will not be discussed further except to state the author's opinion that although the determination of A.F.S. clay in its present form is not of fundamental significance, it serves two main purposes:

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Note: This is the Official Exchange Paper to the American Foundrymen's Society from The Institute of Australian Foundrymen, Victorian Division.

i) it provides a useful service check on a) deliveries of new sand, and b) the state of sand heaps, especially with synthetic sands

ii) it provides a clean material for the grading

analysis.

II. Grading Analysis

Accuracy of Sieving: Sieving tests by their very nature are necessarily empirical and procedure must therefore be rigorously standardized to obtain reproducible results. MacCallman⁶ has discussed a number of variables in sieving of which differences in sieve fabrics, differences in sieve mountings and fittings, time of the sieving test (and the size of the sample), and sampling are most important in foundry sand testing. Sampling will be discussed at a later stage.

Sieve Fabrics are specified by the aperture, the wire diameter and the mesh number (the number of apertures per linear inch): any two are sufficient to define the fabric. Difficulties inherent in weaving necessitate tolerances not only on nominal and maximum aperture but also on the proportion of oversize apertures less than the maximum. This practice is followed in both ASTM⁷ and British Standard^{8*} speci-

fications.

 $^{\rm o}$ British Standard sieves with nominal apertures conforming to the ASTM specification can be chosen to correspond to the sieves specified by the A.F.S. The mesh numbers are 5, 10, 18, 25, 36, 52, 72, 100, 150, 200, 300. Many British workers use sets in which the sieves below 100 mesh are substituted by 8, 10, 16, 22, 30, 44 and 60 mesh; the author considers this undesirable owing to the break at 60 mesh in the $\sqrt{2}$ relationship between the apertures.

The work of a number of authors throws emphasis on the harmful effect of the oversize apertures. Rammler⁹ concluded that the offects of faulty fabric were more pronounced with a steep slope of the cumulative distribution curve in the region of the particular sieve size and with increased sieving time; that the sieving result always tended toward that

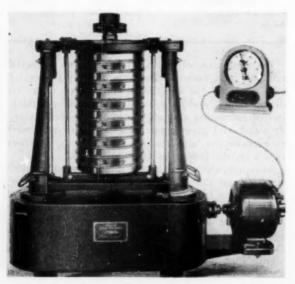


Fig. 1-Ro-tap sieve shaker.

value which would be obtained on a perfect sieve with an aperture equal to the greatest aperture in the sieve; and that oversize apertures approaching the maximum permissible variation from the nominal were the main cause of the error.

Weber and Moran¹⁰ showed that the effective aperture was increased when a sieve had a high dispersion of apertures (as measured by the standard deviation of the openings) and concluded that satisfactory results were obtained only if the standard deviation of the openings was not greater than 6 percent of the average.*

• This suggests that the tolerances of ASTM specification E11-39 are over-generous. With a standard deviation of 6 per cent, the proportion of apertures expected to exceed the nominal by 20 per cent would be less than 0.05 per cent as compared with 5 per cent as allowed by the ASTM tolerances for the 100-mesh sieve; with such a tolerance the standard deviation would be approximately 12 per cent. Measurements of apertures have shown that their frequency distribution follows the normal distribution.

Work¹¹ found that irregular apertures affect the grading results more when sieves are used in nests, and that comparatively small errors in successive sieves of a close series appreciably affect the values obtained.

To overcome these difficulties a number of writers^{6,12} consider that actual sieving tests in which the unknown sieve is related by a correction factor to an arbitrarily chosen master sieve are the best means of comparing and standardizing sieves. A method of calibration is developed in Section III.

Differences in Sieve Mountings and Fittings: These may have a definite (though generally slight) effect on sieving results. Overstretching in mounting may cause distortion of the cloth, and poor workmanship may result in retention of fine particles on a wide layer of the jointing solder or in losses from minute holes in the joints, or from bad fit of sieves in the nest. Minute tears are especially harmful; often they may appear narrower than the aperture but under the sieving action may open and pass a large part of the material normally retained.

Influence of Time and Size of Sample: Two stages are involved in the sieving process; first, the elimination of the fine particles which will pass the apertures of the sieve easily, and second, the elimination of particles which will only just pass the apertures when presented in the most favorable position, these being termed "near-mesh" particles. Heywood¹³ showed that whereas the fine material was eliminated quickly the rate of elimination of near-mesh particles decreased as sieving proceeded, the time required to sieve to a definite rate of passing being dependent on the aperture of the sieve and the proportion of near-mesh particles.

Protracted sieving time permits opportunities not only for the most favorable presentation of nearmesh particles to the apertures, but also for the passage of oversize particles through oversize apertures. Sufficient time must be allowed for the fraction retained on the sieve to approach constancy without allowing the effect of the oversize apertures to predominate.

If the amount passed by the sieve is plotted against

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time, the curve falls steeply and then flattens. The material passing the sieve on the "flat" portion of the curve is composed of both near-mesh particles ultimately presenting their least cross section to the sieve, and oversize particles passing oversize apertures.

The form of the curve will be modified by the intensity of the sieving machine and the size and size-distribution of the material; the narrower the size distribution and the finer the particles the longer the time to reach the flat portion of the curve. Thus the most suitable sieving time corresponds to a point past the "knee" of the sieving curve but not sufficiently far along the flat portion to allow the effect of oversize apertures to predominate.

The size of the sample is related to sieving time. For a given time samples of different size produce



Fig. 2-Inclyno sieve shaker.

different distributions. Since it may be shown that the number of particles per aperture increases proportionately as the size of particle decreases, the size of sample should be given sieving time decrease with the size of particle to allow the same relative opportunity for a particle to be presented to an aperature, or alternatively with the same size of sample the sieving time should be increased proportionately.

With a nest of sieves it should be borne in mind that although a constant weight of sample be adopted, the charge on any sieve will vary not only with the size-distribution of the material but also throughout the time of sieving. In practice, in the A.F.S. fineness test two factors lessen the disadvantage of a constant-weight sample with sands of different fineness:

i) In general (though not always) very fine sands contain a high proportion of clay substance and the sample for grading analysis is therefore often as low as 20-30 grams.

ii) For the coarser sieves the tolerances on oversize apertures are much more stringent and the dispersion of aperatures is generally less than for finer sieves.



Fig. 3—Selby jolter.

Hence, with coarser sands, a sieving time designed for finer sands will not cause serious trouble from the passage of oversize particles through the coarser sieves.

As small a sample as possible is desirable to keep the sieving time to a minimum, but too small a sample leads to loss in accuracy of both sampling and weighing. Samples of the order of 50-100 grams are convenient and will give consistent results. The author considers the A.F.S. sample of 50 grams very satisfactory except for occasional coarse sands and refractories, when 100 gram samples may be employed.

Experimental

Experiments were made to determine the variations due to sampling and sieving, to compare different sieve sets and shakers and to standardize the sieving time of different shakers againt a machine recommended by the A.F.S.

Experimental Procedure: Five groups of sands were used in the investigation:

Sand A	approx.	50
Sand Bfineness	approx.	80
Sand Cfineness	approx.	90
Sands D ₁ and D ₂ fineness	approx.	140-150
Sand Efineness	approx.	25

Sand D₁ was digested with hydrochloric acid for 4-5 hr to remove adhering iron oxide.

Three sets of sieves were used:

- i) a set of half height ASTM sieves;
- ii) a set of half height BSS sieves;
- iii) a set of full height BSS sieves belonging to foundry "A."

The first two sets conform to specifications^{7,11} with data as shown in Table 2.

The standard deviations of the measured apertures were less than 6 per cent with the exception of ASTM 270 mesh (10.4 per cent) and the BSS 60 (8.0 per cent), 150 (6.6 per cent), 200 (11.3 per cent) and 300 (10.3 per cent) mesh sieves.

For each test, samples of 50 grams were used. In

TABLE 2-DATA ON STANDARD SIEVES

ASTM Sieve	Nominal Aperture,	Measured Average	Variat Average A _l		Variation in Maximum Aperture, %	
No.	mm	Aperture, mm	Permitted	Measured	Permitted	Measured
			ASTM Sieves			
12	1.68	1.69	3	+1	10	5
20	0.84	0.83	5	-1	15ª	9
30	0.59	0.60	5	+2	15ª	6
40	0.42	0.43	5	+2	25ª	8
50	0.297	0.308	5	+4	25ª	10
70	0.210	0.215	6	+2	25ª	10
100	0.149	0.147	6	-1	40ª	4
140	0.105	0.106	7	+1	40a	13
200	0.074	0.075	7	+1	60a	10
270	0.053	0.055	7	+4	90a	19
			BSS Sieves	T.	50	***
10	1.68	1.72	3	+2	10	8
16	1.00	1.03	3	+3	10	8
22	0.70	0.70	5	0	20	8
30	0.500	0.514	5	+3	20	10
44	0.353	0.352	5	0	20	9
60	0.251	0.252	6	+1	35	24
100	0.152	0.155	6	+2	35	15
150	0.104	0.098	8	-6	50	7
200	0.076	0.072	8	_5	50	10
300	0.053	0.055	8	+4	80	24
		f the openings shall the permissible varia		opening		

preparing the samples a quantity of sand, mixed by rolling on a leather mat, was reduced by a riffle to between 50 and 100 grams. This was again mixed by rolling on the leather mat and 50 grams were sampled carefully from it with a spatula.

When samples were sieved for varying times the amount on each sieve at any one time was weighed and then carefully replaced on that sieve.

Fineness numbers for sands sieved on BSS sieves were calculated using the factors of Table 3, obtained by interpolation from the ASTM nominal apertures.

Sampling: To determine the variation due to sampling and sieving five samples of sand D₁ were each sieved five times through ASTM sieves for 30 min each on the Inclyno shaker. The results are set out in Table 4 and the fineness numbers in Table 5. The mean for the 25 tests is 136.4.

a) Fineness Numbers: The total variation shown by the 25 fineness numbers may be divided into two parts, one due to differences between samples and the other to sieving variations within samples. From the statistical analysis (Appendix I) it is seen that with the technique adopted in these tests the variations due to sieving are of the same order as the

TABLE 3-FINENESS FACTORS FOR BSS SIEVES

BSS Sie	eve No.		Factor
10		 	 5
16		 	 9
22		 *******	 16
30		 	 25
44		 	 36
60		 	 45
100		 	 60
150		 	 100
200		 	 140
300		 	 200
pan		 	 300

variations due to sampling. Also, 10 per cent of the observations may be expected to differ by at least 3 per cent from the mean, and 1 per cent of the observations may be expected to differ by at least 5 per cent.

This may be compared with the results of Krynitsky and Raring¹⁵ who found with a sand of fineness approximately 70 that the maximum sampling error under the conditions of their test was approximately 3.4 per cent.

b) Grading Results: Just as there are apparently large differences in the fineness number, so the amount retained on any one sieve varies, e.g., for Sample I the amount retained on the 200-mesh varies from 38.8 to 41.6 per cent. When the analysis is summarized by a fineness number the method of calculation tends to obscure the differences in the amounts retained on the various sieves. Hence, the variation in amounts retained on individual sieves would be expected to be somewhat greater than variations in the fineness numbers.

The amounts retained on the 200-mesh sieve are typical: with the technique adopted, 10 per cent of the results may be expected to differ by at least 5 per cent from the mean, and 1 per cent of the results may be expected to differ by at least 8 per cent (Appendix II). Similarly, in resieving the one sample, differences of at least 5 per cent may be expected from 10 per cent of the results. As a corollary, the practice of reporting the percentages on individual screens to more than one place of decimals is misleading.

Sieving Time (Inclyno Shaker): Sands B and D_2 were sieved for varying times on the Inclyno shaker to determine the effects of prolonged sieving. The results are set out in Tables 6 and 7, and the fineness numbers are graphed in Fig. 4.

For sands of fineness 80 or less the flat portion of

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the sieving-time curve is reached prior to the first 20 min, but prolongation of sieving to 30 min would cause little variation in the results. For sands of fineness 150 the comparable portion of the curve is not reached until 30-40 min.

It is interesting to compare the results for sand D_2 at 30 and 340 min. At 340 min the retained fractions had become substantially constant, but the fineness number had increased from 150 to 168 and the pan material from 9.4 per cent to 13 per cent; this confirms the effect of prolonged shaking on the part played by oversize apertures in determining size distribution.

Ro-tap Shaker: Three samples of sand D_1 were sieved on the Ro-tap shaker for periods of 5-15 min. The means of the results are set out in Table 8 and the fineness numbers are graphed in Fig. 5. Thus 15 min (the time specified¹⁵) is sufficient for a sand

Table 4-Sand D₁ Sieved Through ASTM Sieves 30 Min on Inclyno Shaker

ASTM Sieve	1		Sample	e I		Sample II					
No.	1	2	3	4	5	1	2	3	4	5	
30	0.6	0.6	0.6	0.6	0.6	0.6	0.6	0.6	0.6	0.6	
40	1.0	1.0	1.0	1.0	1.0	1.2	1.2	1.2	1.2	1.2	
50	1.2	1.0	1.2	1.0	1.0	1.2	1.0	1.0	1.0	1.0	
70	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	
100	7.8	7.2	7.4	7.2	7.2	7.4	8.0	7.4	7.2	7.6	
140	33.2	31.8	33.4	30.4	32.6	32.2	32.2	32.2	32.0	33.4	
200	40.0	41.6	38.8	41.2	39.2	42.2	39.6	39.6	41.4	39.6	
270	9.0	8.6	8.6	8.8	8.8	8.2	8.6	8.6	8.8	8.4	
pan	6.2	6.6	7.6	8.2	8.0	6.0	7.4	8.0	6.4	7.0	
Total	100.0	99.4	99.6	99.4	99.4	100.0	99.6	99.6	99.4	99.8	
Fine-											
ness	133	134	135	137	136	132	135	136	134	134	
No.											

AST N Sieve	I	Sa	mple	ш	Sample IV					
No.	1	2	3	4	5	1	2	- 3	4	5
30	0.6	0.6	0.6	0.4	0.6	0.6	0.6	0.6	0.6	0.6
40	1.4	1.2	1.2	1.2	1.2	1.0	1.0	1.0	1.0	1.0
50	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0	1.0
70	0.8	0.8	0.8	1.0	0.8	0.8	0.8	0.8	0.8	0.8
100	7.2	7.8	7.6	7.6	7.2	6.6	6.4	6.4	6.6	6.4
140	34.0	30.8	32.0	34.8	30.4	30.8	29.0	29.8	29.8	29.6
200	37.8	41.0	39.0	39.6	41.4	40.6	42.8	42.0	41.2	42.0
270	9.0	8.4	9.0	7.4	8.8	9.8	9.4	9.6	9.6	8.8
pan	8.0	8.2	8.2	6.4	8.2	8.8	8.8	8.6	9.0	9.2
Total	99.8	99.8	99.4	99.4	99.6	100.0	99.8	99.8	99.6	99.4
Fine- ness No.	136	137	137	132	137	139	140	140	140	140

ASTM Sieve		S	ample	· V		
No.	1	2	3	4	5	
30	0.6	0.6	0.6	0.6	0.6	
40	1.2	1.0	1.2	1.0	1.0	
50	1.0	1.0	1.0	1.0	1.0	
70	0.8	1.0	0.8	0.8	0.8	
100	7.2	7.2	6.8	6.8	6.8	
140	32.0	33.8	30.0	30.4	30.0	
200	40.0	38.6	42.0	41.6	42.0	
270	8.8	8.6	9.6	8.8	9.0	
pan	8.4	7.6	7.6	8.2	8.0	
Total	100.0	99.6	99.6	99.2	99.2	
Finene	SS					
No.	137	135	137	138	138	

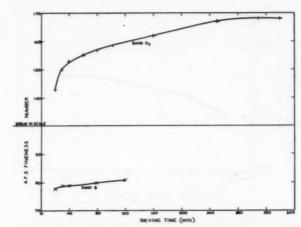


Fig. 4—Fineness number-time curves of sands B and D₂ sieved on Inclyno shaker.

TABLE 5—FINENESS NUMBERS FOR SAND D₁ SIEVED THROUGH ASTM SIEVES FOR 30 MIN ON INCLYNO SHAKER

		(Each sample	sieved five	times)	
Sieving			Sample		
No.	I	II	III	IV	V
1	133	132	136	139	137
2	134	135	137	140	135
3	135	136	137	140	137
4	137	134	132	140	138
5	136	134	137	140	138

TABLE 6-SAND B SIEVED ON INCLYNO SHAKER

135.8

134.2

Mean 135.0

139.8

137.0

BSS Sieve		Sie	eving Time,	min	
No.	20	30	40	80	120
8	0.3	0.3	0.3	0.3	0.3
10	0.2	0.2	0.2	0.2	0.2
16	0.8	0.8	0.8	0.8	0.8
22	0.6	0.6	0.6	0.5	0.4
30	1.8	1.7	1.7	1.5	1.5
44	2.7	2.7	2.7	2.8	2.8
60	5.5	5.1	5.0	4.9	4.7
100	58.8	57.8	57.2	55.7	55.0
150	26.2	27.0	27.6	29.2	30.0
200	1.4	1.8	2.1	2.2	2.4
300	0.5	0.6	0.6	0.6	0.6
pan	1.1	1.2	1.2	1.3	1.3
Total	99.9	99.8	100.0	100.0	100.0
Finene	85				
No.	78	79	79	80	81

TABLE 7-SAND D2 SIEVED ON INCLYNO SHAKER

BS				Sicving	Time,	min			
No	20	30	40	60	80	160	250	310	340
30	0.5	0.5	0.5	0.5	0.5	0.5	0.5	0.5	0.5
44	1.1	1.1	1.1	1.1	1.1	1.1	1.1	1.1	1.1
60	1.1	1.1	1.0	1.0	1.0	1.0	1.0	1.0	1.0
100	2.7	2.7	2.7	2.6	2.6	2.5	2.5	2.5	2.5
150	37.1	25.2	23.6	22.4	21.0	17.6	15.4	15.0	14.5
200	35.7	43.2	41.9	40.7	41.1	40.4	38.8	37.6	37.8
300	12.6	16.6	19.5	21.8	22.5	25.1	28.3	29.6	29.5
pan	9.2	9.4	9.7	9.9	10.0	11.8	12.5	12.8	13.1
Tota	1								
1	0.00	99.8	100.0	100.0	99.9	100.0	100.1	100.1	100.0
Fine	ness								
No.	143	150	153	155	157	162	167	168	168

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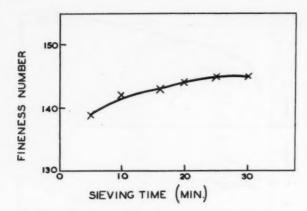


Fig. 5—Fineness number-time curve of sand D₁ sieved on Ro-tap shaker.

of fineness approx. 140 to reach the balance between satisfactory sieving and emphasis on oversize apertures.*

* In the 1944 edition of the FOUNDRY SAND TESTING HAND-BOOK no sieving machine is specified.

Defective Sieves: Over a period it was noticed that fineness numbers determined by foundry "A" were in general higher than those in the laboratory on similar sands.

Samples of sand D_2 were sieved in both laboratories for 30 min on a Selby Jolter (Table 9).

The fineness number from foundry "A" sieves was

TABLE 8—SAND D₁ SIEVED ON RO-TAP SHAKER

ASTM				1		
Sieve			Sieving	Time, min	1	
No.	5	10	15	20	25	30
20	0.0	0.0	0.0	0.0	0.0	0.0
30	0.6	0.6	0.6	0.6	0.6	0.6
40	1.1	1.1	1.1	1.1	1.1	1.1
50	0.9	1.0	0.9	0.9	0.9	0.9
70	0.8	0.8	0.7	0.7	0.7	0.7
100	6.5	5.8	5.7	5.5	5.3	5.2
140	28.3	27.1	26.5	26.0	25.7	25.4
200	43.2	43.8	44.0	44.2	44.2	44.4
270	10.7	11.3	11.6	11.8	12.0	12.1
pan	7.8	8.6	8.9	9.2	9.4	9.6
Total	99.9	100.1	100.0	100.0	99.9	100.0
Finenes	SS					
No.	139	142	143	144	145	145

Table 9—Sand D₂ Sieved Through Foundry "A" and BSS Sieves

BSS Sieve No.	Foundry "A" Sieves on Foundry "A" Selby Jolter	BSS Sieves on M.T.C. Selby Jolter		
30	0.1	0.2		
44	1.5	1.2		
60	0.3	1.0		
100	2.6	2.4		
150	8.2	19.5		
200	44.0	45.5		
300	12.0	20.1		
pan	31.2	10.1		
Total A.F.S.	99.9	100.0		
Fineness No.	190	156		

much higher than that from the C.S.I.R.O. sieves, mainly due to the larger amount of pan material. Differences in apertures were not sufficient to account for the discrepancy; in fact, the aperture of the foundry "A" 300-mesh sieve was approximately 14 per cent less than the nominal value, and therefore less pan material might be expected. The foundry "A" sieve contained two small tears (Fig. 6) which, however, under the microscope appeared unlikely to affect the results.

To eliminate the difference between the laboratory machine and that of foundry "A" and to investigate the effect of the tears, a sample was sieved on three different machines, using two sets of sieves (Table 10).

TABLE 10—SAND D SIEVED THROUGH FOUNDRY "A"

	Minus 300 Mesh							
Shaker	Foundry "A" Sieves, %	BSS Sieves, %						
Ro-tap	6.5	9.6						
Inclyno, 30 min	7.8	10.0						
M.T.C. Selby, 30 min	13.5	10.1						

The tears apparently had little effect with the Rotap and Inclyno shakers, but the action of the Selby Jolter caused the tears to open and pass an increased percentage of pan material. On inspection, the foundry "A" Selby Jolter was found to run in the reverse direction to that at the Melbourne Technical College, resulting in much more vigorous shaking which accounts for the much higher percentage of pan material obtained by foundry "A."

The two tears were repaired and sand D_2 was again sieved on the foundry "A" jolter with the result that the 300-mesh fraction (20 min sieving time) was reduced from 19.6 to 6.7 per cent.

The surprising effect of such a small defect illustrates the necessity for maintaining sieves in perfect condition. Each sieve should be examined weekly under a low power microscope or hand lens for defects, and grading analyses should be checked for abnormal results.

Comparison of Sieve Shakers

Ro-tap and Inclyno Shakers: The shakers were compared in two experiments using sands C and D₁. Ten samples of sand C and eight of sand D₁ were sieved through ASTM sieves. Half of each were sieved for 15 min on the Ro-tap and half for 20, 30, and 40 min on the Inclyno shaker. The fineness numbers are set out in Tables 11 and 12.

The statistical analyses (Appendices III and IV) showed that, with either sand (and presumably those of intermediate fineness), sieving for 40 min was necessary for results similar to those for 15 min on the Ro-tap. There was a significant difference for 30 min on the Inclyno, but for sands of fineness within the range 90 to 140 it was possible to convert the Inclyno results to those for 15 min on the Ro-tap by multiplying by 1.03 (Appendix IV).

The difference of 3-4 per cent between 15 min sieving on the Ro-tap and 30 min on the Inclyno would have little practical significance, since different

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TABLE 11—FINENESS NUMBERS FOR SAND C SIEVED ON THE RO-TAP AND INCLYNO SHAKERS

	Ro-Tap		Inclyno	
	15 min	20 min	30 min	40 min
	92	88	90	91
	93	90	91	91
	93	89	91	91
	93	90	91	91
	91	86	88	90
Mean	92.4	88.6	90.2	90.8

TABLE 12—FINENESS NUMBERS FOR SAND D₁ SIEVED ON THE RO-TAP AND INCLYNO SHAKERS

	Ro-Tap		Inclyno	
	15 min	20 min	30 min	40 min
	138	133	135	137
	142	132	135	137
	143	136	139	140
	143	131	133	134
Mean	141.5	133.0	135.5	137.0

grab samples from a heap or pit would vary by at least 4 per cent and further, other factors would have a much greater effect on surface finish than a 3-4 per cent difference in fineness. Thirty min was considered the most desirable sieving time. It was therefore decided to sieve all sands for 30 min on the Inclyno shaker and if necessary to convert to Ro-tap figures using the factor 1.03.

Inclyno and Coombs Type Shakers: Similar experiments were made (using sand D₁) to compare the Inclyno shaker with a Coombs type run for 10, 15 and 20 min. Fifteen min on the Coombs type did not differ significantly from 30 min on the Inclyno, but when the means of the Inclyno fineness numbers were multiplied by 1.03 to convert to Ro-tap figures (above) there was a significant difference, and 20-25 min sieving was necessary for comparable results.

Inclyno and Cheers Shakers: At a later date it was decided to change from the Inclyno to the Cheers shaker (a machine similar to the Coombs type with an action imparted by the out-of-balance motion of a rotating quadrant weight). Tests using sand D₁ showed that 25 min were required on the Cheers machine for results comparable to 30 min on the Inclyno shaker. The quadrant weight on the Cheers machine was altered to reduce the time to 15 min.

Comparison of Sieves

Twenty-four samples of sand D₁ were prepared. Eight were sieved through ASTM sieves, eight through BSS sieves and, eight through foundry "A" sieves, four of the samples in each group being shaken on the Ro-tap shaker for 15 min and four on the Inclyno shaker for 20, 30 and 40 min. The fineness numbers are set out in Table 13.

The fineness numbers for BSS and foundry "A" sieves have been calculated using the factors set out in Table 3. When these factors are used (Appendix V) there is no significant difference in the results for ASTM and BSS sieves, but the foundry "A" sieves differ significantly.

Fineness numbers calculated for ASTM and BSS sieves agree closely. However, an inspection of the

TABLE 13—FINENESS NUMBERS OF SAND D₁ SIEVED THROUGH ASTM, BSS, AND FOUNDRY "A" SIEVES ON THE RQ-TAP AND INCLYNO SHAKERS

Ro-Tap										
A	STM	BSS	Foundry "A"							
	138	149	146							
	142	141	154							
	143	139	151							
	143	138	153							
Mean	141.5	141.8	151.0							
		Incluno								

*,	As	STM			BSS		Foundry "A"			
	20 min	30 min	40 min	20 min	30 min	40 min	20 min	30 min	40 min	
	133	135	137	132	135	137	139	143	144	
	132	135	137	130	134	137	136	139	142	
	136	139	140	135	139	140	139	143	146	
	131	133	134	133	136	138	144	146	147	
· fean	133.0	135.5	137.0	132.5	136.0	138.0	139.5	142.8	144.8	

 Mean of all observations

 ASTM
 136.8

 BSS
 137.1

 Foundry "A"
 144.5

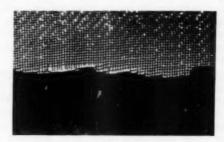


Fig. 6-Tear in 270-mesh sieve. ×8.

grading analyses on which these fineness numbers are based (Table 14) reveals what at first sight appear to be obvious discrepancies, e.g., there is a difference of 8.2 per cent between the average amounts remaining on the 150-mesh sieves, and a difference of 8.4 per cent between the average amounts remaining on the 200-mesh sieves.

Although each set conforms to its standard specification, 7.8 there are nevertheless slight differences in the measured apertures of comparable sieves. (It

Table 14—Sand D₁ Sieved Through ASTM, BSS and Foundry "A" Sieves for 30 min on the Inclyno Shaker

A	STM Sieve	es	В	SS Siev	Foundry "A" Sieves		
Sieve No.	Aperture,		Sieve	ture,	Average on Sieve, %	Sieve	Average on Sieve, %
70	0.215	3.5	60	0.252	2.9	60	3.0
100	0.147	6.5	100	0.155	4.4	100	5.4
140	0.106	32.4	150	0.098	40.6	150	24.4
200	0.074	41.1	200	0.072	32.7	200	42.1
270	0.055	9.5	300	0.055	12.2	300	20.5
Pan		6.9	Pan		6.9	Pan	4.6
Total Fineness		99.9			99.7		100.0
No.		136			136		143

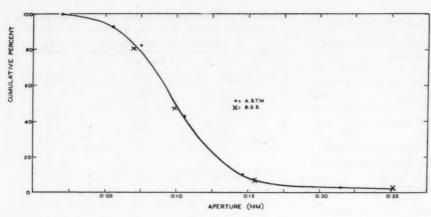


Fig. 7—Cumulative distribution curve for sand D₁ on ASTM and BSS sieves.

should be noted here also, that the nominal apertures of comparable sieves in the two series differ by similar small amounts.)

To examine the effects of these small differences in aperture, points for cumulative distribution curves were plotted for the BSS and ASTM sieves. A single curve could be fitted to the plotted points of both sets of sieves (Fig. 7), and thus there was no inherent difference between the results from the two sets, the apparent large discrepancies of Table 14 being due in reality to the small differences in both the aperture and the behavior of the sieves (differences in dispersion of apertures, etc.).

The magnitude of permissible variations in nominal aperture in any one sieve series is of the same order as the variation between the BSS and ASTM sieves, and therefore similar apparent discrepancies may occur when using different sets of the one series, e.g., the —40 +50 mesh fraction of standard sand ST2477B (see Section III) would vary between 23.0 and 29.5 per cent with 50-mesh sieves whose nominal apertures varied between the upper and lower tolerances of the ASTM specification.

The fineness number of foundry "A" sieves differed significantly and it was immediately obvious that this could be attributed to differences in average aperture coupled with differences in behavior of the sieves themselves (effect of oversize apertures, etc.). It was therefore decided to calibrate the sieves, taking the ASTM set as standard and using the method described below (Section III). The 300, 200 and 150-mesh foundry "A" sieves behaved as theoretical 320, 190 and 130-mesh ASTM sieves.

Although the fineness numbers for the BSS and ASTM sieves did not differ significantly, it was also decided to use the method of calibration to bring the apertures of the BSS sieves into line with those of the ASTM sieves; the 200- and 150-mesh BSS sieves behaved as theoretical 195- and 150-mesh ASTM sieves.

Fineness numbers were recalculated using these new factors (Table 15). The statistical analysis (Appendix VI) showed that the fineness numbers from any of the sets did not then differ significantly.

The tests were repeated using sands C (fineness 90), A (fineness 50) and E (fineness 25), and where necessary new factors were determined for the remaining sieves.

The sieving of these four sands has enabled the fineness factors of BSS and foundry "A" sieves to be studied, for all practical purposes, throughout the whole range. It has been shown that slight differences in average aperture coupled with the effect of oversize apertures has made some of the sieves behave as the equivalent of ASTM sieves of slightly different aperture. The effective apertures of such sieves have been determined and new fineness factors calculated. These new factors are compared with the A.F.S. factors in Table 16.

III. A Method for Calibrating Sieves

Figure 8 shows an ideal cumulative distribution curve in which the size is plotted against cumulative per cent oversize. The shape of the curve is a characteristic of the size distribution of the material; any one material will give the same curve whatever the method of determining size, provided the method is exact. In practice molding sands and similar materials are sieved on standard screens and the cumulative oversize is plotted against measured aperture. The results for two perfect sieves of aperture A and A₁ would be represented by the points X and X₁.

Conversely, if the curve has been determined and

TABLE 15-CORRECTED FINENESS NUMBERS OF SAND D.

			Ro-	Ttap				
A5	TM	*	В	SS		Found	iry "A'	1
	138		1	49		1	38	
	142		1	41		. 1	46	
	143		13	39		1	44	
	143		13	37		1	45	
			Inc	lyno				
	ASTM			BSS		Fo	undry '	'A'
20 min	30 min	40 min	20 min	30 min	40 min	20 min	30 min	40 min
133	135	137	132	135	137	132	136	137
132	135	137	129	134	137	129	132	135
136	139	140	134	138	140	132	136	139
131	133	134	132	136	139	138	139	140

Computed using complete set of new factors listed in Table 16 Mean of all readings

	747	16	-6	X.E	ĸ.	- 1	U,	В.	-	3,1	A	п	4	-	33	LE.	I.A	*8	
AST	M	E																	136.8
BSS																			136.8
Four	nd	lr	y		6	1	1	9.9				*							137.4

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the cumulative oversize on a sieve is C_1 , the aperture of the sieve would be A_1 . Thus, by sieving a sand on sieves of known aperture, plotting the cumulative distribution curve and then sieving the same sand on sieves of unknown aperture and determining the cumulative per cent oversize on each unknown sieve, the unknown sieves may be calibrated by reading off their apertures from the cumulative distribution curve.

This method has been used in this laboratory for standardizing sieves of Australian foundries. Three standard sands suitable for the range of A.F.S. sieves are sieved through the C.S.I.R.O. screens (whose apertures have been carefully measured). They are then sieved by the foundry concerned, the apertures of the screens are related to the measured C.S.I.R.O. (master) sieves by determination from the cumulative distribution curve, and new A.F.S. fineness factors are calculated.

As seen in Section II, the sieving value depends not only on the average aperture, but also on the dispersion of apertures and the method of shaking. The above calibration method therefore relates the unknown sieve set shaken under the particular foundry's conditions to the standard sieves and method of shaking adopted in this laboratory.

As an example, the author sieved sand "No. 72"

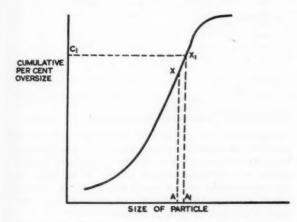


Fig. 8-Typical cumulative distribution curve.

six times on ASTM sieves and foundry "B" sieves with the results shown in Table 17.

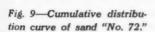
Both the grading and the fineness numbers differed considerably, and it was obvious that some sieves were behaving as oversize, e.g., 63.6 per cent only was retained by the 140-mesh foundry "B" sieve compared with 90.8 per cent on the 100-mesh ASTM sieve. The

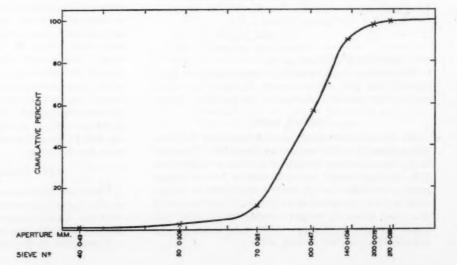
Table 16—Fineness Factors for ASTM, BSS, and Foundry "A" Sieves

ASTM	Sieves		BSS	Sieves				
	A.F.S.			New Factor				
Mesh No.	Fineness Factor	Mesh No.	Original Factor	C.S.I.R.O.	Foundry "A"			
6	3	8	4	4				
12	5	10	5	5	5			
20	10	16	9	9	9			
30	20	22	16	16	16			
40	30	30	25	25	25			
50	40	44	36	32	31			
70	50	60	45	44	49			
100	70	100	60	55	55			
140	100	150	100	95	97			
200	140	200	140	150	130			
270	200	300	200	195	190			
Pan	300	Pan	300	300	320			

TABLE 17-SIEVING OF SAND "No. 72" (Average of Six Sievings)

	AST	ASTM Sieves		y "B" Sieves
Mesh No.	Per Cent	Cumulative Per Cent	Per Cent	Cumulative Per Cent
12	0.2	0.2	0.2	0.2
20	0.3	0.5	0.2	0.4
30	0.2	0.7	0.3	0.7
40	0.4	1.1	0.5	1.2
50	1.4	2.5	0.6	1.8
70	8.8	11.3	2.9	4.7
100	45.2	56.5	20.4	25.1
140	34.3	90.8	38.5	63.6
200	6.9	97.7	33.6	97.2
270	1.1	98.8	2.1	99.3
Pan	1.3	100.1	0.8	100.1
Total	100.1		100.1	
	AST	M	87	ctors)





Ap	parent	Calibration Theoretical	Average Measured.	Aperture	Maximum D	eviation from	Standard Deviation
		Theoretical	Managed				To CATHETOIL
Sieve Ape			Measureu,	Deviation	Nominal	Aperture	of
Sieve Aperture, No. mm	ASTM mm Mesh No.	from Nominal, %	Measured, mm	Permitted,	Measured Apertures, %		
70 0	0.253*	60*	0.218	+4	+45	+25	15.3
100 0	0.190	75	0.154	+3	+60	+40	18.8
140 0	0.138	110	0.109	+4	+70	+40	23.4

apparent apertures of the foundry "B" sieves, read off from the cumulative graph of the ASTM sieve results (Fig. 9) together with data obtained by optical measurement are shown in Table 18.

The average apertures conform to the standard.⁷ However, the oversize apertures and the high degree of dispersion (as measured by the standard deviation) cause the sieves to behave with apertures as in-

dicated by the graphical method.

The theoretical ASTM mesh number is determined as follows, taking as an example the foundry "B" 100-mesh sieve whose nominal aperture should be 0.149 mm, but whose apparent aperture is 0.190 mm. The nominal apertures of the 70 and 80 mesh ASTM sieves are 0.210 and 0.177 mm. Hence the mesh number of the foundry "B" sieve will lie between 70 and 80. The difference between the apparent aperture of the foundry "B" sieve and the nominal aperture of the 80 mesh ASTM sieve is 0.013 mm. The difference between the apertures of the 70 and 80 mesh sieves is 0.033 mm, while the difference in meshes is 10. Hence on an approximate proportional basis the mesh number of the foundry "B" sieve will be less than 80 by

$$\frac{0.013}{0.033} \times 10 \text{ meshes} = 4 \text{ meshes},$$

i.e., foundry "B" sieve is a theoretical 76 mesh sieve. For practical purposes the factor 75 is used.

The fineness number was recalculated using 60, 75 and 110 as the new factors for the sand remaining on the 100, 140 and 200 mesh foundry "B" sieves. It agreed closely with that for the ASTM sieves:

Foundry "B" sieves: Fineness No.

Original factors = 108 Corrected factors = 86

ASTM sieves: Fineness No. = 87 The method was also used for determining the apertures of the BSS sieves and the foundry "A" sieves for the experiments described in the previous section.

Standard Sands

Six sets of three standard sands covering the sieve range from 20 to 270 mesh were prepared. These are being circulated to foundries for sieve calibration. The original weight of each sample was 51 grams approx. to allow for the small gradual loss in weight with use by a number of foundries. (The sample is discarded when its weight reaches 48.5 grams; this small variation in the size of the sample has no eignificant effect on the sieving value.)

The samples were sieved five times on ASTM sieves and the averages determined. A sieve analysis of standard set "B" is shown in Table 19, and the cumulative curves in Fig. 10.

TABLE 19—STANDARD SANDS FOR CALIBRATING SIEVES

ASTM	Average Aperture,*	Gra	ding Analys	sis, %
Sieve No.	mm	ST2477B	ST2479B	ST2483B
6	3.310	0.0	0.0	0.0
12	1.726	2.6	0.0	0.0
20	0.855	15.9	0.2	0.2
30	0.605	18.0	0.2	0.4
40	0.419	21.9	1.9	1.4
50	0.292	26.8	19.5	1.2
70	0.214	11.0	30.9	0.4
100	0.146	2.5	26.1	1.8
140	0.105	0.4	17.2	18.2
200	0.077	0.0	3.6	49.2
270	0.056	0.1	0.2	20.2
Pan		0.0	0.1	7.1
Totals		100.1	100.0	100.1
.F.S. Fine	ness No.	31	65	151

*ASTM sieve set No. 4 used for the test has slightly different apertures from set No. 1 (Table 2).

Experience has shown that sand ST2483 is not as suitable for calibration purposes as a less uniform sand. It will be replaced later by a sand with greater spread of particles over four sieves. In determining the new fineness factors, the A.F.S. factor is changed only if the calculated aperture falls outside the tolerance of the ASTM sieve concerned.

To ensure uniformity of sieving results in fineness tests of molding sand, there is need for a central authority to prepare standard samples for distribution to foundries. Three (or possibly four) sands should be sufficient to cover the range of ASTM sieves; each sand should have the majority of grains fairly uniformly distributed over four sieves. The samples should be sieved on carefully chosen ASTM sieves with a very small dispersion of apertures (S.D. less than 2 per cent if possible). The sieving time should also be standardized; 15 min on the Ro-tap shaker is recommended. It is suggested that the A.F.S. Grading and Fineness Committee is the appropriate authority for the work.

IV. Grading Classifications

There are three main methods for reporting the results of grading analysis: 1) tabulation; 2) graphical methods; 3) determination of an index (or indices) summarizing the results.

Tabulation of the screen analysis provides general

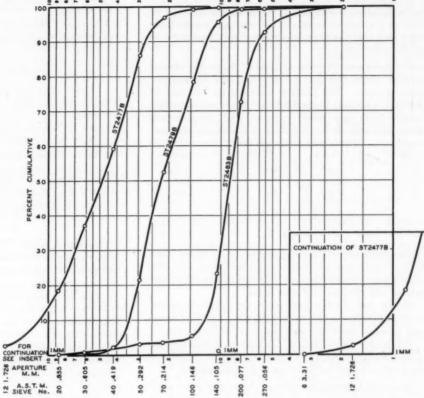


Fig. 10—Cumulative curves of standard sands.

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information as to the range of sizes and the proportions of coarse and fine material. Graphs may be either the normal type in which the individual sieve fractions are plotted against the sieve size or the cumulative type in which the cumulative percentage of oversize (or undersize) is plotted against the sieve size. The first contains one or more peaks corresponding to sieves retaining large amounts of material; a single high peak denotes a uniform sand and a number of peaks a poorly distributed sand. The second type rises gradually, then more steeply and then flattens; a steep curve denotes a uniform sand and vice versa.

After examining a number of methods of determining indices representing the fineness of molding sands, including indices of average particle diameter, the A.F.S. decided in 1925 to introduce the Grain Fineness Number. This designates the sand in terms of average mesh rather than particle diameter.

The A.F.S. Grain Fineness Number has proved very popular and successful for comparing sands. Its advantages are that the calculations involved are simple and the units (mesh numbers) are readily visualized by foundrymen and research workers alike. However, the limitations of sieving are such that significance should not be placed on small variations in fineness number, especially with finer sands. (This will apply to any index calculated from a sieving analysis.) In practice it suffers from the disadvantage that the calculation tends to weight the finer grains.

The fineness number has been severely criticized by a number of authors, mainly/on the grounds that

different sands with the same fineness number may differ in properties, e.g., Morey and Taylor⁴ describe two sands A and B with the grading analysis:*

Sand	A	. В
Clay Substance, %	26.5	0
A.S.T.M. Sieve No.	% Retained	% Retained
20	7.0	0.5
30	6.0	0.5
40	17.0	1.5
50	18.0	7.5
70	9.5	41.5
100	4.5	37.0
140	1.5	9.0
200	1.5	2.5
270	1.5	0.0
Pan	7.0	0.0
Total sand grade	73.5	100.0
A.F.S. Grain Fineness	No. 68	64

The green permeabilities at suitable moisture contents and after 5 per cent of bentonite had been added to foundry sand B were: Sand A, 27; Sand B, 141. While some of the differences can be attributed to the difference in clay substance, the major part must lie in differences of the grading sands.

The author agrees that the fineness number alone is not a guide to other properties but considers that the criticism arises largely from failure to realize its limitations. It is a measure of the average size but cannot convey information as to uniformity. Some measure of grain distribution is necessary to complement the fineness number.

Grading analysis calculated from cumulative curves (Fig. 4 of ref. 4).

A grain distribution number¹⁶ proposed by the A.F.S. Grading & Fineness Committee was adopted in 1936 as a tentative standard. It was apparently little used and has been omitted from the latest edition of the standards.¹

Trask Index: Morey and Taylor⁴ have described a method developed by Trask using the cumulative grading curve. The particle sizes for 25, 50, and 75 per cent retained are called the first quartile (Q_1) , median (M) and third quartile (Q_3) , respectively. The coefficient of sorting S_0 and coefficient of skewness S_k are determined from the formulae:

$$S_o = \sqrt{\frac{Q_1}{Q_3}}$$

$$S_k = \frac{Q_1 \times Q_3}{M^2}$$

A low coefficient of sorting denotes a uniform sand. If the coefficient of skewness is less than one, the maximum concentration of particles lies to the coarse side of the median and vice versa.

Morey and Taylor have used the method successfully to determine the texture of steel molding sands and to study sand reclamation plant performance.

While the Trask indices are easy to determine the quantities are difficult to visualize, and therefore the method may not readily appeal to foundrymen.

Proposed Distribution Index: The author considers the determination of the largest percentage of the "sand" grade remaining on three adjacent sieves in the screen test (together with an estimate of "fines"), to be a simple yet highly satisfactory method of designating distribution. Sands with more than 75 per cent on three adjacent sieves are considered uniform. With this method a major difference between Morey and Taylor's sands A and B becomes apparent immediately; sand A has 61 per cent of "sand" grade on three adjacent sieves (40, 50 and 70 mesh) whereas sand B has 88 per cent on three adjacent sieves (70, 100 and 140 mesh).

"Fines": Some workers regard "fines" as minus 100-mesh material. The author considers this unsound. With such a classification the major part of a fine sand would be "fines." Foundrymen rightly associate the term "fines" with grains which serve to block pore spaces and reduce the permeability, though they may improve the surface of the casting slightly. To designate the major part of any sand "fines" is clearly illogical. The term "fines" must therefore be considered an elastic one, the size varying with the nature of the sand. Thus for a coarse sand, minus 70-mesh material might be considered "fines"; for a medium sand, minus 100 or minus 150 mesh; for a fine sand, minus 270 mesh.

The author suggests that if the percentage in the screen test on three adjacent sieves is adopted as a measure of distribution, "fines" might be defined as the percentage of material passing the next sieve after the three adjacent sieves, e.g., if the majority of the sand remains on the 40, 50 and 70-mesh sieves the material passing 100 mesh would be fines.

Sand A may now be classified as follows: Clay Substance =26.5%

A.F.S. Grain Fineness No. =68

Per cent on 3 adjacent sieves=61% of "sand" grade on 40, 50 & 70 mesh

Fines =11.5%—100 mesh or briefly,

C26.5/N68/P61 on $40 \rightarrow 70/F$ (—100)11.5)

This is the requisite information on the texture of the sand.

Since its introduction in 1925 the fineness number has been most useful. Whatever its faults it should not be discarded lightly. As an adjunct to the fineness number the author recommends that the Foundry Sand Grading Classifications be modified to include a distribution index similar to that proposed above and to include a definition of "fines."

V. Conclusions

Grading Analysis: Sieving, though probably the best means of determining the size distribution of medium and coarse particles, is not an absolute method but depends on a number of factors which must be carefully standardized. The sieving value is a function of the characteristics of the sieve cloth, the size of sample, and the time of sieving. In particular, oversize apertures and defective sieves cause variations which are aggravated by a prolonged sieving time. Sieves with a small dispersion of apertures (standard deviation less than 6 per cent) provide consistent results.

In fineness tests on material sampled with a riffle, the variations due to sampling and sieving are of the same order. Ten per cent of the fineness numbers may be expected to differ by at least 3 per cent from the mean. A somewhat greater variation is to be expected in the sieving results of the individual screens; 10 per cent of the results may be expected to differ by at least 5 per cent from the mean.

The most satisfactory sieving time varies with different shakers. Sufficient time should be allowed for the fraction retained on the sieve to approach constancy without allowing the effect of oversize apertures passing oversize particles to predominate. Fifteen min on the Ro-tap or 30 min on the Inclyno shaker is sufficient to reach the "flat" portion of the sieving curve.

Sieving times corresponding to 15 min on the Rotap shaker are: Inclyno shaker, 40 min; Coombs type shaker, 20-25 min; Cheers type (with suitably modified quadrant weight), 15 min. For sands fineness 90-140, 30 min on the Inclyno may be converted to 15 min on the Rotap by multiplying by 1.03.

Sieve Calibration: Cumulative distribution curves prepared from the results of sieving under standard conditions on standard sieves may be used to calibrate unknown sieves under different sieving conditions. Theoretical mesh numbers calculated from the effective apertures of the unknown sieves may be used to determine new factors for the calculation of A.F.S. Fineness Numbers.

The author suggests that the A.F.S. Sand Division is the appropriate central authority for calibrating sieves for the fineness tests of molding sands.

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Grading Classifications: Results may be reported by tabulation, by graphical methods or by a summarizing index. The A.F.S. Fineness Number has proved very satisfactory as an index of average grain size. Criticism that it is not a reliable guide to other properties dependent on texture is mainly due to failure to recognize that the fineness number cannot provide information as to size distribution.

A number of distribution indices have been proposed (including the Trask Index), but the author suggests that the calculation of the maximum amount remaining on three adjacent sieves in the fineness test provides an index which is both simple and contains all the necessary information.

The term "fines" should designate material varying in size with the average size of the sand: the material passing the next sieve after the three adjacent sieves concerned in the calculation of the distribution index provides a simple index of fines.

Clay substance and fineness number together with the indices of distribution and fines provide the essential information on the texture of a molding sand.

Acknowledgments

The author's thanks are due to Mr. F. G. Nicholls, Head Office, C.S.I.R.O., for advice and criticism and for suggesting the statistical treatment; to Messrs. Cornish and Williams, Section of Mathematical Statistics, C.S.I.R.O., who planned the statistical tests and performed the analyses; to the officers of the Ore Dressing Laboratory at the Melbourne University for their advice and criticism and for permission to use the Ro-tap machine, and especially to Mr. A. E. Scott, the reader of the paper. Finally, it is a pleasure to acknowledge the ready cooperation of a number of foundries in this work.

Bibliography

1. FOUNDRY SAND TESTING HANDBOOK, "Standards and Tentative Standards," American Foundrymen's Society, Chicago, 5th Ed., 1944. (See also FOUNDRY SAND HANDBOOK, 6th Ed., 1952.)

J. F. Haseman, "The Effect of Different Electrolytes on the A.F.A. Clay Determination," A.F.A. TRANSACTIONS, vol. 46, p.

851 (1938)

3. R. C. Hills, "Comparison of Some Wet Methods Used for the Fineness Test of Sands and Clay," A.F.A. TRANSACTIONS, vol. 42, p. 101 (1934). 4. R. E. Morey and H. F. Taylor, "Use of the Cumulative

Curve for Foundry Sand Control," A.F.A. TRANSACTIONS, vol.

53, pp. 66-76 (1945).

M. Price and A. I. Krynitsky, "Molding Sand Fineness
 Test," A.F.A. TRANSACTIONS, vol. 54, pp. 402-411 (1946).
 D. MacCalman, "The Accuracy of Sieving Tests, Part 3,"

Indust. Chem. 14: 231-6 (1938).

7. Standard Specifications for Sieves for Testing Purposes, ASTM Specification No. E11-39.

8. Test Sieves, British Standard Specification No. 410: 1943

(Lendon, 1943). 9. E. Rammler, "Untersuchungen uber die Messung der Kohlenstaubsenheit durch Handsiebung," Arch. f. Warmenwirtschaft

10. M. Weber and R. F. Moran, "A Precise Method for Sieving Analyses," Ind. & Eng. Chem. (Anal. Ed.) 10: 180-4 (1938).

11. L. T. Work, "Methods of Particle Size Determination,"

ASTM Proc. 28 (Part 2): 771-812 (1928).
12. R. J. Wig and J. C. Pearson, "Standardization of No. 200 Cement Sieves," U.S. Bur. of Standards, Tech. Paper No. 42

13. H. Heywood, "Measurement of the Fineness of Powdered Materials," Inst. Mech. Eng. Proc. 140: 257-347 (1938).

14. TESTING AND GRADING FOUNDRY SANDS AND CLAYS-STAND-ARDS AND TENTATIVE STANDARDS, American Foundrymen's Society, Chicago, 4th Ed., 1938. (See also FOUNDRY SAND HANDBOOK, 6th Ed., 1952.)

15. A. I. Krynitsky and F. W. Raring, "Silica Sands—Sieve Analyses," American Foundryman, 18 (2): 40 (Aug., 1950).
16. W. G. Reichert, P. Bechtner, C. Mathiesen, and C. M.

Saeger, Jr., "A.F.A. Grain Distribution Numbers for Foundry Sand," A.F.A. Transactions, vol. 44, p. 545 (1936).

Appendix I-Statistical Analysis of Results of Table 5

TABLE 1A-ANALYSIS OF VARIANCE

Variation Due to	Degrees of Freedom	Sums of Squares	Mean Squares
Sampling	4	95.36	23.84
Sieving	20	44.40	2.22
Total	24	139.76	5.82

The total variation, i.e., the sum of the squares of the deviations of the 25 fineness numbers from their mean, may be divided into two parts, one due to differences between sample means and one to sieving variations within samples. Designating the variances of these parts by M and N, respectively, we have

5M + N = 23.84

and since the estimate of N = 2.22

we obtain 4.32 as the estimate of M.

The standard deviation of the difference between M and N is 3.37. Since the distribution of the difference is not normal, an exact test of significance cannot be made; we may, however, use the ratio, difference: standard deviation, as a rough guide in assessing significance. In the present case the difference is only about 3/5 of its standard deviation, so that we are not likely to be seriously in error by concluding that the variations due to sieving are of the same order of magnitude as the variations due to sampling with this type of sand and with the technique adopted in these tests.

Taking the two sources of variation together, the variance (s2) of a single observation is 5.82 and the standard deviation (s) is therefore 2.4. When n = 24and P = 0.1, t = 1.7 and ts = 4.1, which is 3 per cent of the general mean, and when n = 24 and P =0.01, t = 2.8 and ts = 6.7, which is 5 per cent of the general mean. Thus approximately 10 per cent of the observed values may be expected to differ from the mean by at least 3 per cent, and approximately 1 per cent of values to differ by at least 5 per cent.

Appendix II-Statistical Analysis of Amounts Remaining on 200-Mesh Sieve, Table 4

TABLE 2A—ANALYSIS OF VARIANCE

Variation Due to	Degrees of Freedom	Sums of Squares	Mean Squares
Sampling	4	11.13	2.78
Sieving	20	32.59	1.63
Total	24	43.72	1.82

Taking the two sources of variation together, we have $s^2 = 1.82$, whence s = 1.3. Then for n = 24and P = 0.1, we have t = 1.7 so that ts = 2.2 which is 5 per cent of the general mean, and for n = 24 and P = 0.01 we have t = 2.8 so that ts = 3.6, which is 9 per cent of the general mean. Hence it may be expected that in a sieving analysis of this type, with samples obtained by the riffle method, 10 per cent of the results would differ by at least 5 per cent from the mean, while 1 per cent of the results may be expected to differ by at least 9 per cent.

Taking the variation due to sieving separately, the variance is 1.63 and the standard deviation 1.3. Then for n=20 and P=0.1 we have t=1.7 and ts=2.2 which is 5 per cent of the general mean, and for n=20 and P=0.01 we have t=2.8 and ts=3.6, which is 9 per cent of the general mean. Hence in resieving the one sample, 10 per cent of the results may be expected to differ by at least 5 per cent from the mean, and 1 per cent of the results by at least 9 per cent.

Appendix III—Statistical Analysis of Results of Table 2
TABLE 3A—ANALYSIS OF VARIANCE

			* *************************************	
Variation Due to	Degrees of Freedom	Sums of Squares	Mean Squares	Ratio
Time (Inclyno)	2	12.93	6.47	
Samples (Inclyno)	4	15.06	3.77	
Time x samples	8	3.73	0.47	
	14	31.73		
Samples (Ro-Tap)	4	3.20	0.80	
Shakers	1	24.07	24.07	17:44××
Total	19	59.00		
All samples				
+ time x sample	es 16	22.00	1.38	

The variance of a single observation is 1.38 and consequently the variance of the difference betwen any two shaker means is

$$1.38 \times \frac{2}{5} = 0.55$$

and the standard deviation 0.74.

TABLE 3B—COMPARISON OF SHAKERS

Shaker	Time in min	Mean Fineness	Diff. from Mean of Ro- tap	τ
Ro-tap	15	92.4		_
Inclyno	20	88.6	3.8	5.1
Inclyno	30	90.2	2.2	3.0
Inclyno	40	90.8	1.6	2.2

For P=0.01 and n=16, t=2.92. Hence for Sand C, with fineness number approx. 90, there is no significant difference between the values for sieving on the Ro-tap shaker for 15 min and on the Inclyno shaker for 40 min.

The results for sieving on the Ro-tap shaker for 15 min differ significantly from those obtained after sieving for 30 min on the Inclyno shaker. The ratio of the mean fineness numbers is 1.024. The variance of this ratio may be obtained as follows:

Let R = mean fineness for 15 min on Ro-tap, I = mean fineness for 30 min on Inclyno,

$$F = \frac{I}{I}$$
 and let V(F) stand for the variance of F.

Then
$$V(F) = \left(\frac{\gamma F}{\gamma R}\right)^2 V(R) + \left(\frac{\gamma F}{\gamma I}\right)^2 V(I)$$

$$= \left(\frac{1}{I}\right)^2 V(R) + \left(\frac{R}{I^2}\right)^2 V(I)$$

and since, from the analysis of variance, V(R) = V(I) = v, say,

$$V\left(F\right) = v \left(\frac{1}{I^{2}} + \frac{R^{2}}{I^{4}} \right).$$

When the variance of the ratio considered here is

$$\frac{1.38}{5} \left(\frac{1}{90.92} + \frac{92.4^2}{90.94} \right) = 0.000070$$

Appendix IV-Statistical Analysis of Results of Table 12

TABLE 4A-ANALYSIS OF VARIANCE

Variation Due to	Degrees of Freedom	Sums of Squares	Mean Squares	Ratio
Time (Inclyno)	2	32.67	16.34	
Samples (Inclyno)	3	49.67	16.56	
Time x Samples (Inclyno)	6	1.33	0.22	
	11	83.67		
Samples (Ro-tap)	3	17.00	5.67	
Shakers	1	120.33	120.33	21.22xx
Total	15	221.00		
All samples + tir	ne			
x samples	12	68.00	5.67	

The variance of the difference between any two shaker means is

$$5.67 \times \frac{2}{4} = 2.84$$

and the standard deviation 1.7.

TABLE 4A—COMPARISON OF SHAKERS

Shaker	Time in min	Mean Fineness	Diff. from Mean of Ro- tap	t
Ro-tap	15	141.5		_
Inclyno	20	133.0	8.5	5.0
Inclyno	30	135.5	6.0	3.5
Inclyno	40	137.0	4.5	2.6

For P = 0.01 and n = 12, t = 3.06. Hence for Sand D_1 , with fineness number approx. 140, there is no significant difference between the values for sieving on the Ro-tap for 15 min and on the Inclyno for 40 min.

The results for sieving on the Ro-tap for 15 min differ significantly from those for 30 min on the Inclyno. The ratio of the mean fineness numbers is 1.044. The variance of this ratio, determined by the method given above, is

$$\frac{5.67}{4} \left(\frac{1}{135.5^2} + \frac{141.5^2}{135.5^4} \right) = .000161$$

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We may now determine the significance of the difference between the ratios obtained above for sands C and D_1 . The variance of the difference is:

0.000070 + 0.000161 = 0.000231 and the standard deviation is, therefore, 0.0152. Since the difference between the ratios is 0.020

$$t = \frac{0.020}{0.152} = 1.32$$

The variances 0.000070 and 0.000161 are based on 16 and 12 degrees of freedom, respectively. The best value of n with which to enter the t table is the harmonic mean of these two numbers, and this is

$$2\left(\frac{12\times16}{12+16}\right) = 13.7.$$

For P=0.01 and n=14, t=2.98. Hence we must conclude that the ratios do not differ significantly and to correct fineness numbers for sieving for 30 min on the Inclyno to the values to be expected for sieving for 15 min on the Ro-tap shaker we may multiply by the average of the two, i.e., by a factor of 1.03 (at least for sands with fineness numbers 90-140).

Appendix V-Statistical Analysis of Results of Table 13

TABLE 5A-ANALYSIS OF VARIANCE

Variation Due to	Degrees of Freedom	Sums of Squares	Mean Squares	Ratio
Inclyno				
Times	2	148.17	74.08	
Sieves	2	392.67	196.34	33.85××
Times x sieves	4	2.66	0.67	
Samples + Tin x Samples with				
Sieves	27	156.50	5.80	
	35	700.00		
Ro-tap				
Sieves	2	234.50	117.25	8.13**
Samples	9	129.75	14.42	
	11	364.25		
Shakers	1	451.56	451.56	
Total	47	1515.81		
Samples + times				
x samples	36	286.25	7.95	X

The variance of the difference between any two sieve means is $7.95 \times \frac{2}{16} = 0.9931$ and the standard deviation, 1.0.

TABLE 5B-COMPARISON OF SIEVES

Sieve	Mean Fineness	Difference from mean of ASTM	t
ASTM	136.8	_	_
BSS	137.1	0.3	0.3
Foundry "A"	144.5	7.7	7.7

For P=0.01 and n=36, t=2.7. The results for Foundry "A" sieves differ significantly from those for both ASTM and BSS sieves, but the latter types do not differ significantly.

Appendix VI-Statistical Analysis of Results of Table 15

TABLE 6A-ANALYSIS OF VARIANCE

Variation Due to	Degrees of Freedom	Sum of Squares	Mean Squares
Times	2	162.89	81.45
Sieves	2	0.39	0.20
Times x sieves	4	6.44	1.61
Samples + times			
x samples			
within sieves	27	161.50	5.98
	35	331.22	
Ro-tap		007144	
Sieves	2	8.17	4.09
Samples	9	138.75	15.42
	46	146.92	
Shakers	1	416.84	416.84
Total	47	894.98	
Samples + times			
x samples	36	300.25	8.34

The variance of the difference between any two sieve means is $8.34 \times \frac{2}{16} = 1.04$ and the standard deviation 1.0.

TABLE 6B—COMPARISON OF SIEVES

Sieve	Mean Fineness No.	Difference from Mean ASTM	t
ASTM	136.8	_	_
BSS	136.8	0.0	0.0
Foundry "A"	137.4	0.6	0.6

For P=0.01 and n=36, t=2.70. Hence there is no significant difference between results with ASTM and either BSS or Foundry "A" sieves, using the new factors in the calculation of fineness numbers.

DISCUSSION

Chairman: H. J. WILLIAMS, New Jersey Silica Sand Co., Millville, N. J.

Co-Chairman: STANTON WALKER, National Industrial Sand Ass'n, Washington, D. C.

Recorder: D. C. WILLIAMS, Ohio State University, Columbus. Co-Chairman Walker: The mounting for the Ro-tap sieve shaking machine is very important. The machine set on a wooden base will produce results quite different from those obtained when a concrete base is used. The paper "Reproducibility of Tests of Foundry Sand" by myself, published in A.F.A. Transactions, vol. 49, pp. 789–835 (1941) presents many of the difficultie in trying to obtain reproducible results among laboratories for the grain size distribution.

Chairman Williams: The A.F.S. Grain Fineness Number

CHAIRMAN WILLIAMS: The A.F.S. Grain Fineness Number does not mean a thing for it gives no indication of grain distribution. Furthermore the Grain Fineness Number or grain distribution as furnished by the supplier is of little value when dry sand is dumped into a silo and subsequently drawn off as required.

M. F. Drumm: ¹ The A.F.S. Grain Fineness Number provides no indication of the gravity packing ability of sands for the production of shell molds. I would like to know if there is less variability in the cumulative curves than there is in the Grain Fineness Numbers for small lots.

Co-Chairman Walker: There is better reproducibility for the cumulative curves.

¹ Research Chemist, Monsanto Chemical Co., Plastics Div., Springfield, Mass.

MANUFACTURE OF CAST EDGE TOOLS IN ROTARY FURNACE STEEL

By

Gordon Keech,* A. T. Batty** and W. M. Dummett†

This subject will bring many different comments from members of the Institute of Australian Foundrymen. From the big steel manufacturer, making thousands of tons each day for re-rolling, we may find indifference; from the small jobbing steel foundry, using expensive electric arc or induction melting, we may find criticism and doubt; from others temporary interest before final disinterest. To the few, having the gift of adding imagination to the process of analysis, this achievement could suggest enormous possibilities.

While no novelty is claimed for any individual feature of the process, the entire cycle being a simple application of well known technical facts, the achievements of this rather small foundry organization reflect the spirit of progressive metallurgical and foundry enterprise that brought our Institute into being.

To introduce the subject we will say that first interest would lie in the fact of being able to replace forgings with sound castings of high grade tool steel in not just ordinary rough blanks, but very intricate and difficult shapes, more economically and more accurately than the corresponding forging. Second claim to interest lies in the method of production of both mold and metal; and third, the application of the product to usages that are systematically destructive, and for which previously only tailor-made steels, laboriously produced through many expensive and complicated processes, were considered.

The manufacture of edge tools started in 1943, when the production of bren gun carrier sprockets was undertaken in an arrested cycle malleable iron. This proved so successful, giving three times the life of those made from the specified steel, that tool merchants and users, starved for supplies, asked the possibility of making many types which they required. Prominent among them were axes, which are un-

doubtedly in this country the most abused of all tools, and expected to stand such abuse in cutting ours, the hardest and toughest timbers in the world. This tool presented the most problems, but having in mind the building of a post-war tool industry, competitive against all products in price and quality, local and imported, it provided a challenge. After much research and experiment, an order for hatchets was accepted from U. S. Navy Stores, and delivered with entire satisfaction, passing severe tests with flying colors.

Black sand molding was quickly abandoned because of its failure to produce sound and accurate castings, which are now made in pitch-bound core sand. A silicon-manganese steel was adopted for the process of tool manufacture, and heat treatment furnaces were installed.

With the cessation of hostilities in 1945, this firm found itself the proud possessor of a complete tool manufacturing plant, useless in their everyday prewar activities when it supplied gray and malleable iron castings to the engineering trade. They had to decide whether to return to their well known steady business, or whether to attempt to break into the tool trade, knowing that buyers in peacetime would exhibit preference for the long established drop forged tools.

Having faith that their products were top-grade, they made the decision to specialize in tools, and were encouraged by two successful procedures they had developed and perfected, namely:—

(1) Inoculation of the molten metal with special alloys to give toughness and fine-grained structure so essential in tools, and

(2) The core sand mold not only produced most accurate castings but had increased production per molder up to eight times of that in black sand.

To-day the firm is Australia's largest manufacturer and supplier of hand tools and edge tools, and simple multiplication could make it the largest in the world. To a country already famous for producing the world's cheapest mild steel rolled sections, they have become the first producers of cast alloy steel hand tools. The cost of a casting is about the same as that

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Fig. 1-Photo of two half molds of hand tools.

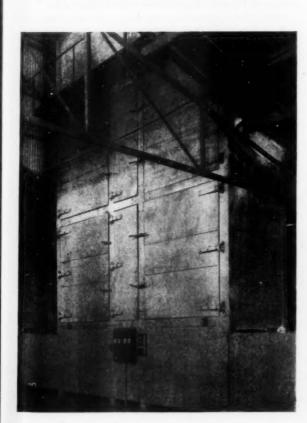




Fig. 2—Producing coresand molds—continuous core oven in rear.

of the piece of rolled steel required to make a forging, leaving a greater margin in which to finish the tool and sell competitively.

Having briefly described the origin of this development, we shall devote the rest of the paper to the detailed description of the various tools and their actual manufacturing procedure.

The products include axes, hatchets, carpenters' hammers, tinsmen's snips, tailors' and trimmers' shears, etc.

In making the mold, two grades of sand are used, and these are dried in a rotary oil-fired drier and conveyed to overhead storage bins ready for mixing. An edge runner mill of 1080-lb capacity carries out the mixing cycle in 15 min. The whole cycle of drying and preparing the core oil sand is done by one man, handling an average of 10 tons per day.

The material used in this process is: Sand a—Average bulk screen, 120-150 GFN. Sand b—Average bulk screen, 70 GFN.

Fig. 3 (Left)—Continuous vertical core oven and instrument panel.

Fig. 4 (Below)—Mouth of conveyor oven.

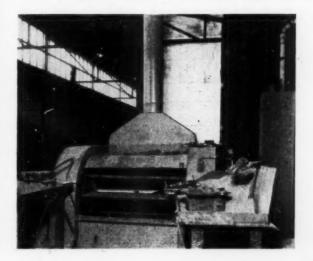




Fig. 5—Pouring molten metal from buggy ladles in casting area.

Sands a and b are mixed in the proportion of 1a:4b; to which is added $2\frac{1}{4}$ per cent of soft pitch binder and is used as a facing sand. Sand b is mixed with a pitch binder at 1.5 per cent and is used as backing or bulk sand.

Half mold weight varies from 5 to 10 lb. These are at present made by hand at an average rate of 1000 halves per man, per 8-hr shift. See Fig. 1.

Core box equipment is made so that each single plate makes both halves. Male and female dowels on each plate provide perfect indexing and no molding boxes are used. The plate and shroud are in one piece and molds are turned over onto a cast aluminum tray.

Molders work in pairs and can fill the conveyor oven with four halves to each shelf in a 11/4-hr cycle.

The coremaking process has been made a quite simple one, perfect pattern equipment has eliminated any need to trowel the mold and patterns need only be cleaned every 10 or 12 molds. The process clearly illustrated in Fig. 2 shows a molder in the process of adding the facing sand, this is quickly spread over the pattern face, followed by a double-handed scoop action which fills the remainder of the box with backing sand. The box is then scraped level with the bar shown opposite the molder, the core tray placed on, followed by the cross jig on the molder's left. The mold is next turned over, vibrated for several seconds, then stripped. This leaves the core on the plate ready to be placed on the automatic spray turntable by the molder on the right hand side of the illustration. The mold is then given a light spray coating of linseed oil and kerosene, and placed onto the oven shelf, the complete molding cycle requiring about 20 sec.

Mold baking is done in a conveyor type oven (Fig. 3) with two horizontal and two vertical sections providing 4 zones, (1) the pre-heating, out through

which passes the exhaust gases; (2) a superheating zone where the temperature is 100 degrees above the baking temperature to allow quick raising of the mold temperature without the addition of new oxygen; (3) the baking zone, where oxidizing conditions exist, and (4) the cooling zone, through which the new air is admitted, cooling the mold and taking on much of the heat to the oxidizing or baking zone. No molding boxes are used, indexing being done with mounds or recesses molded onto and into the faces. The whole sequence is automatically controlled once the cycle in each particular area is decided and set. Figure 4 shows an empty shelf arriving at the mouth of the oven.

Figure 3 shows the conveyor oven, specifications of which are as follows:

The belt consists of 138 two-storied trays. The average cycle is 75 min on a hot oven and the average operating temperature in the baking zone is 500 F.

Cores for hollowing out cavities for handles and forming the claws are made on a small core blower and pass through the oven with the core sand molds. They are made from the same sand that is used in the molds and are bonded with the same binders.

On leaving the oven the molds are cooled enough to handle, they are immediately assembled and clamped together between plates in groups of 14 and conveyed on special trolleys to the casting area.

This area is adjacent to the rotary melting furnaces and is open on three sides, which makes a quite comfortable process of casting by quick dissipation of oil fumes and heat. (Fig. 5.)

Metal is poured from two-wheel pneumatic-tired buggy ladles, holding 100-lb of steel. Casting time for a bank of 14 molds is $1\frac{1}{2}$ min after which the castings are allowed to cool until the mold is greatly burnt away. Castings are picked out of the sand and



Fig. 6.—Tapping rotary furnace into buggy ladle.

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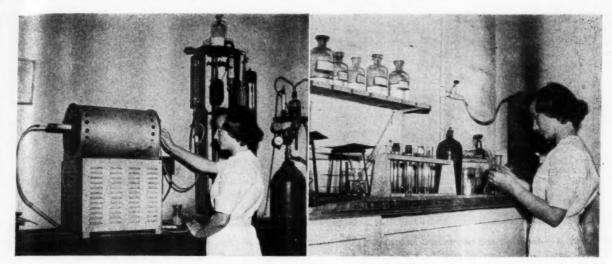


Fig. 7-Sections of chemical laboratory. Routine carbon determinations are made before tapping any heat of steel.

hauled away for cleaning. The casting area is then cleaned up with a small bulldozer and loader and the used sand dumped.

The method of melting is of some interest, and a description of this phase of the production now follows. The melting of steel for foundry work usually requires electric furnaces of one type or another, and it is freely admitted that they are preferable from the viewpoint of control of temperature, chemical analysis, etc. However, as Keech Castings Ltd. pioneered, developed and perfected the process on the basis of tar-fired rotary furnaces, no change is warranted on account of quality or melting cost, and shortage of electric power in this Country is a further deterrent to a change to electric furnaces.

The metal is melted in a battery of five acid-lined rotary furnaces, each of 1-ton capacity (Fig. 6). A monolithic lining of rammed siliceous ganister has been found more reliable than a brick lining, and the average life of this rammed lining is about 25 melts. The average time per melt is 11/2 to 2 hr, and the charge mainly comprises steel scrap drawn from sources where there is no likelihood of contamination from unwanted alloy steels, sulphur, freecutting steels, etc. Use of good quality spring steel minimizes the problem of building up the carbon content, and all runners and risers are returned for melting. Carbon is added inside steel tubes, and pig iron is not favored as a means of increasing the carbon, because pig iron available in Australia does not generally contain a suitably low phosphorus content.

The alloy content of the metal is composed within the furnace by means of ferro-alloys, the following being the basic analysis of all tools:—

Small amounts of chromium, molybdenum, tungsten and vanadium are incorporated to give improved qualities for specific applications.

The molten charge is carefully check-analyzed before tapping, and the metal is inoculated in the ladle to produce an inherently fine austenitic grain size. (Fig. 7 shows chemical laboratory.)

In handling the metal between furnace and molds, good equipment and perfect teamwork ensure the minimum delay, so that a pouring temperature of 3200 F or better can be maintained. Needless to say the whole process depends for its success on clean, sound castings, free from molding or metallurgical defects, and every casting has been the subject of individual research to develop the most perfect methods of running, risering, etc.

This job of producing good castings is the most important part of the process, but yet the shortest. There still remain dressing and finishing processes, as well as carefully evolved heat treatment procedures.

A good hand tool is required to resist a combination of abrasion and shock. Ordinary mechanical tests do not give an accurate indication of the quality of hand tools and it has been found necessary to perform arduous, practical tests on a vast number of tools to appraise their fundamental requirements. For this reason heat treatment cycles and microstructures (Figs. 8-13) do not necessarily conform to theoretically perfect ideals, but are to a large degree determined by trial. By these methods it has been established beyond doubt that the as-cast structure is more resistant to pure abrasion than is any other produced from the same composition. As can readily be appreciated, the quenched and tempered structure is more resistant to shock, but complete removal of the as-cast ferritic dendrites is unnecessary, and possibly undesirable. Some evidence of incomplete annealing can be seen in Figs. 11 and 12.

Getting down to details, the following heat treatments of axes and claw hammers are typical of all tools produced,

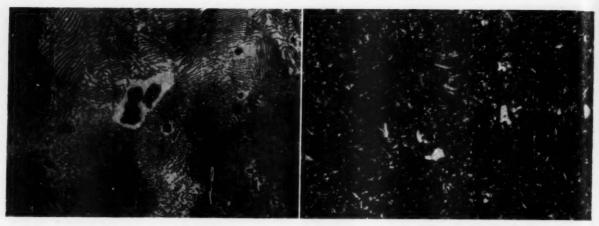


Fig. 8—Annealed and normalized structure of axe, Rockwell C 24. Lamellar pearlite. x 600, Nital etch.

Fig. 9—Hardened structure of racing axe close to cutting edge. Rockwell C 60. Fine grained tempered martensite. x 600, Nital etch.

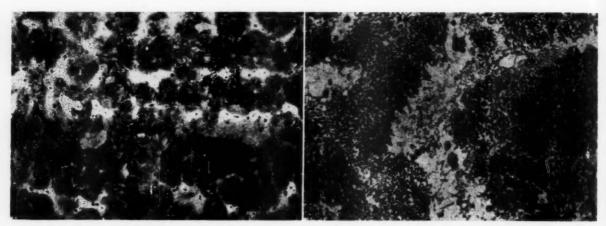


Fig. 10—As-cast structure of claw hammer Rockwell C 26. Fine lamellar pearlite with interdendritic ferrite dotted with manganese sulphide inclusions. x 100, Nital.

Fig. 11—Structure of annealed hammer body, Rockwell C 24. Partially spheroidized pearlite. x 600, Nital etch.

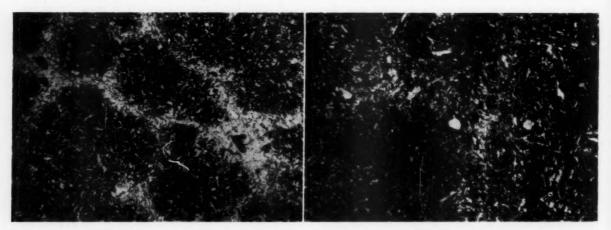


Fig. 12—Structure of quenched and tempered body of hammer (Rockwell C 28)—partially spheroidised lamellar pearlite with interdendritic ferrite still apparent. x 600, Nital etch.

Fig. 13—Structure of flame-hardened face of hammer, (Rockwell C 58), tempered martensite. x 600, Nital etch.



Fig. 14-View of heat treat department.

Axes and hatchets are first of all annealed at 1650 F and then normalised at 1560 F to break up the ascast dendritic structure to the required degree. The actual annealing time and temperature depends on the mass and the ruling of the tool in question. To harden the cutting edge, 2 in. of the blade is immersed in a lead bath at 1560 F, the axe being suspended by a bar through the eye. After heating for a sufficient time to complete austenitization, the whole axe-head is oil quenched, and before being allowed to cool down to room temperature, it is transferred to a lead tempering bath at 650 F.

This treatment gives the tool no chance to form quench cracks, and a uniform structure of fine tempered martensite found at the cutting edge has a hardness of Rockwell C 52-60. (Racing axes are maintained at Rockwell C 58-60, and the structure is shown in Fig. 9.) The back of the axe is lamellar pearlite, with a Rockwell hardness of C 24, as shown in Fig. 8. Manganese sulphide crystallites are readily visible, but do not seem to act as stress raisers or cause the axe to fail in any way in service. This is probably because they are equiaxed and not elongated by rolling or forging. Claw hammers are annealed in a carburizing atmosphere at 1650 F, then oil quenched from 1560 F and tempered at 1100 F to produce a partially spheroidised fine grained pearlite.

Figure 10 depicts the as-cast structure, with pronounced ferrite dendrites which become partially removed by annealing (see Fig. 11). The quench and temper treatment produces very fine partially spheroidized pearlite as shown in Fig. 12. This structure is strong but tough, and has a hardness of Rockwell C 28. There are two parts of a claw hammer which do hard work, viz the striking face and the claw, both of which are now flame hardened to produce a local hardness of Rockwell C 58. The martensitic structure partly tempered extends to a depth of more than 1/4 .a., and is capable of withstanding the rigors and abuses to which the hammer is subjected. In all heat treatment, particular care is taken to avoid decarburization by putting carbonaceous matter on the hearth of the muffle furnaces where any air is likely to leak in. This is particularly important in the case of the claw hammers, as there is no practical way of grinding the inside of a claw, and a soft decarburized

claw will be burned and damaged by allowing nails to slip through.

The heat treatment equipment is all batch type at the present moment, electric muffle furnaces as shown in Fig. 14 and 15 being used for annealing and austenitizing with oil-fired lead baths for some hardening and tempering. The tools are charged in buckets and long soaking times are necessary to ensure that the center of the charge reaches temperature for adequate time. To overcome this disadvantage, a continuous process is being designed and built, embodying rotary-hearth gas-fired furnaces for hardening, and apron-conveyor furnaces for tempering.

At this stage, most of the dressing operations have been already performed as may be expected on the raw castings, elastic abrasive wheels being mostly used for removing runners and risers. Because of fine core-sand molds the skin is smooth, and does not need shot blasting. Better finish is obtained, in fact, by rumbling, and this is generally performed before any heat treatment is commenced.

In the case of axes, hatchets, etc., requiring grinding over an extensive area, or pliers which must be machined, these operations are carried out after annealing.

All of the grinding is carried out at the Works itself, and grinding jigs ensure perfect control of shape and profile so essential to the balance of tools such as axes, and so important in the smooth cutting properties of tailors' shears, tinsmen's snips, and the like. A certain amount of machining must be carried out at this stage in such tools as pliers and wrenches, and it is generally found more convenient to let this work out to other firms.

After the hardening and tempering operations



Fig. 15—A load of mattock and plier castings entering electric annealing furnace.

described previously, the tools are polished and given one of several finishing treatments. Axes are protected by a rust preventive grease, and hammer heads may otherwise be given a coat of crinkle paint or they may be chrome-plated. Handles of tailors' shears or tinsnips are given a smooth decorative finish of synthetic enamel.

Wooden handles as used for so many tools are made entirely at their own woodworking shop. The billets are bandsawn to a rough form and then shaped in copying lathes. Sand belt smoothing down and barrel-rumbling ensure that the handles are free from splinters and rough spots before fitting and packaging.

North American hickory is the best timber procurable for handles, being strong, tough and resilient. The only other timber put into handles by this firm is Australian spotted gum, which is quite good enough for utility tools, and allows the tool to be sold at a considerably lower price.

Like every manufacturer, this firm receives some complaints from its clients. Hand tools are the most commonly abused articles either in the home or the workshop, and many failures are due to the use of the tool in a manner or for a purpose for which it was not intended. The few failures due to founding or metallurgical shortcomings, however, are carefully noted, and recurrence prevented.

The successful application of the process to hand tools and edge tools has made this foundryman realize that many other small steel articles or parts could likewise be made cheaper or better than they could by forging, pressing, machining or welding.

Let us realize, therefore, that founding is not an old-fashioned, obsolescent, crude or inferior method of shaping metal, but one that can, by thought and care, maintain and improve its position of importance amongst modern manufacturing processes.

INFLUENCE OF ALUMINUM ON PROPERTIES OF CAST GUN METAL AND REMOVAL OF ALUMINUM BY SLAG

By

Ake V. Larsson*

It is well known that it may be difficult at times to attain the desired pressure tightness in copper alloys, including gun metal, cast in sand. The reason may be, for example, the presence of pores caused by freezing shrinkage, of blisters produced in the melt by dissolved gas that has been released during solidification of the metal, or of some harmful impurity such as aluminum.

Opinions are rather divided on the aluminum content that may be permitted in gun metal. Thus according to ASTM the aluminum allowance should be none,† while the tentative Swedish Standards allow maximum 0.005 and D.I.N. (German) up to 0.01 per cent.

In the present investigations the influence of aluminum on the pressure tightness and mechanical properties of gun metal have been examined, and experiments have been undertaken to remove aluminum from brass and gun metal by means of slag-forming refining agents. The pressure tightness and mechanical properties have again been examined after removal of the aluminum.

The investigations have been carried out under the direction of a Research Committee within the Foundry Section of The Federation of Swedish Mechanical Engineering Industries. The following have taken part in the work of the committee: S. Brennert, chairman; P. O. Björkman; O. Carlsson; M. Claesson, A. Larsson; E. O. Lissell; S. Lundqvist; and B. Lunn. B. Löfgren and B. Wadell, AB Svenska Metallverken, cooperated in the performance of the practical experiments.

Influence of Aluminum on Mechanical Properties and Pressure Tightness

Experimental Procedure: The investigations into the influence of aluminum on pressure tightness and mechanical properties have been exclusively made on

a leaded gun metal (83-6-5-6 copper-tin-lead-zinc).

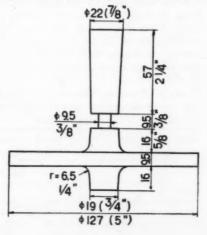
For examination of the pressure tightness, test castings were selected as used, among others, by Baker, Child and Glaisher, i.e., a poorly-fed disc shown in Fig. 1. For examination of the mechanical properties a well-fed test bar was cast, the so-called D.T.D. bar, Fig. 2. Test discs and bars were separately cast in the experiments. The investigations were made on 50-125 kilo (110-275 lb) melts. Melting was done either in an oil-fired or a coke-fired crucible furnace. The charge consisted of alloyed ingots.

The following procedure was generally adopted. After melting under charcoal or an oxidizing cover and heating to pouring temperature, the melt was deoxidized to 0.05 per cent phosphorus with phosphorus

Immediately the pouring temperature was attained discs and bars were cast in sand molds dried at 180° C for 10 hr. The aluminum was then introduced, new test specimens were cast, the aluminum-removing slag was introduced into the melt, and aluminum was removed. Further bars and discs were then cast.

An account is given later in the paper of the method adopted for removing aluminum from the melt by means of slag. In the present experiments, the removal of aluminum was accomplished with a slag consisting of sodium fluoride (30 per cent), calcium

Fig. 1 — Test disc for investigation of pressure tightness.



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[†]The term "none" as applied to aluminum allowance is defined as 0.005 per cent when determined on a 10-gram sample.

Official Exchange Paper from the Foundry Section, Federation of Swedish Mechanical Engineering Industries, Stockholm, Sweden.

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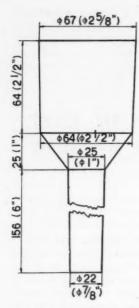


Fig. 2—Test bar, vertically cast (D.T.D. bar).

fluoride (20 per cent), sodium aluminum fluoride (20 per cent), sodium sulphate (20 per cent), and sodium carbonate (10 per cent).

The tightness of the discs as cast was tested with compressed air under water. Those discs which proved to be tight were retested after machining.

The test bars were examined for density, tensile strength and elongation. The contents of aluminum and other impurities were controlled by means of a spectrograph. Strength tests were carried out on a test bar of normal Swedish type.

Causes of Pores in Castings: As stated above, among the factors influencing the tightness and other properties of a casting are the following:

 Gases absorbed by the metal during melting and casting.

b. Shrinkage occurring on solidification of the metal, and which usually is especially concentrated when the metal is "gas free."

c. Impurities such as aluminum which can only be removed from the melt by special methods.

If the disc used in the experiments proves to be leaky, it does not give any certain clue as to the cause. This may be gas, which has been released on solidification of the metal and given rise to pores, or freezing shrinkage due to poor feeding or the presence of some impurity such as aluminum.

TABLE 1—PRESSURE TIGHTNESS IN DISCS CAST BEFORE AND AFTER ADDITIONS OF 0.01, 0.02 AND 0.2 PER CENT ALUMINUM

Time of	Alum-	No. of	No. of	Pressure Tightness							
Casting	inum, %	Melts	Discs	No. of Tight Discs	No. of Leaky Discs	Tight Discs, %					
Before Al- addition After Al-	<0.005	19	52	22	30	42					
addition	0.01	8	24	1	23	4					
	0.02	2	- 6	0	6	0					
	0.20	5	20	0	20	0					

In order to determine the cause of the leakage, test bars were cast as well as discs from the same melt. The test bar used has a very good feed, and freezing shrinkage in the bar itself may therefore be considered as completely eliminated. Since it is known, according, to, for example, Baker, Child and Glaisher, that ultimate strength and density fall comparatively quickly with an increase in the amount of gas, the test bars can give indications as to whether the pores in the disc are caused by freezing shrinkage or by gas.

With high density and good strength it may be considered probable that the leakage is caused by freezing shrinkage, and with low density and poor strength by gas in the metal, provided that no impurities are present that would impair pressure tightness.

Experience teaches that melts of certain types of

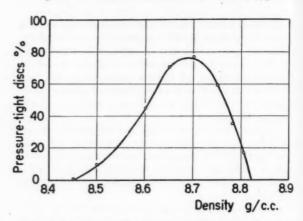


Fig. 3—Curve according to Baker, Child and Glaisher¹ showing percentage of pressure-tight discs secured for different densities in an 88-8-4 copper-tin-zinc alloy.

Density determined on D.T.D. bars.

alloy should normally contain a small amount of gas in order that a poorly-fed casting shall have a proper chance of being pressure tight. However, investigations by Baker, Child and Glaisher¹ and Mantle² show that the circumstances are somewhat different in casting leaded or lead-free gun metal.

These authors came to the conclusion that an 88-8-4 copper-tin-zinc alloy should contain a small quantity of gas, while an 85-5-5-5 copper-tin-lead-zinc alloy should contain as little gas as possible if the castings are to be pressure tight. The findings of Baker, Child and Glaisher are shown graphically in Figs. 3 and 4. The pressure tightness was investigated after machining. Casting was done at 1180 C in green sand containing 5 per cent moisture.

Figures 3 and 4 show that the majority of 88-8-4 alloy discs become pressure tight at a density of 8.67—8.68, and that practically all discs become leaky when the density exceeds about 8.8 and falls below about 8.45. For an 85-5-5-5 alloy the optimum pressure tightness is obtained at a density of about 8.8 or above.

In order to be able to study the influence of impurities on pressure tightness, the influence of gas and freezing shrinkage must be eliminated. This means that the melt must possess such properties AL

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that the discs become pressure tight when no impurity that would impair pressure tightness, such as aluminum, is present.

Normally, there is no difficulty in obtaining a "gassy" melt. On the other hand, it is much more difficult to reproduce a given gas content, which, on solidification of the metal, prevents the creation of

too large and concentrated cavities.

Influence of Pouring Temperature on Strength and Density: Pouring temperatures exercise a predominating influence on strength. To illustrate this, and partly to attempt to explain the occasionally low tensile and elongation values, Fig. 5 gives results of an experiment in which test bars were cast from the same melt at 'different temperatures. The first were cast at 1230 C, whereupon the melt was left to cool, further rods being cast at 1130 and 1050 C.

The diagram shows that ultimate strength, elongation and density fall with a rise in pouring temperature. The reason must be the slower solidification which gives rise to a coarser solidification structure,

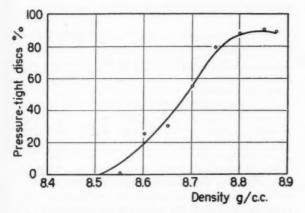


Fig. 4—Curve according to Baker, Child and Glaisher¹ showing percentage of pressure-tight discs secured for different densities in 85-5-5-5 copper-tin-lead-zinc alloy.

Density determined on D.T.D. bars.

and the greater gas solubility with rising temperature.

Influence of Aluminum: An addition of 0.2 per cent aluminum was made to certain melts. The content was selected as being representative of that which might occur in refining. It is considered, however, that considerably lower aluminum contents impair pressure tightness, and therefore the influence of contents down to 0.01 per cent has been investigated.

a. Pressure Tightness: The results of investigations of pressure tightness have varied somewhat, and in the cases where leaky discs were obtained before the addition of aluminum no opinion on the influence of aluminum could be formed. Nevertheless, the comparatively large number of discs cast from several separate melts have made a reasonable judgment on this point possible. It should be pointed out, however, that melting and casting were performed under proper operating conditions, and that no special measures were taken to make the melt "gassy" or "gas free."

It will be seen from Table 1 that an aluminum

TABLE 2—TENSILE STRENGTH, ELONGATION AND DENSITY OF D.T.D. BARS CAST BEFORE AND AFTER ADDITION OF 0 01, 0.02 and 0.2 Per Cent Aluminum

Time of	Alum-	No. of	No. of	U.T.	S.	Elong-	Density,
casting	inum, %	melts	bars	kg/sq mm	psi	ation,	gm/cc
Before A	Al-						
addition	< 0.005	11	28	20.9	29700	13	8.80
After Al							
addition	0.01	4	11	19.3	27450	11	8.91
	0.02	2	3	18.4	26150	8	8.87
	0.20	4	12	12.9	18350	3.5	8.67

content of 0.01 per cent exercises an unfavorable influence on the pressure tightness of leaded gun metal. Of the discs cast, 42 per cent were pressure tight before the addition of aluminum, while only one disc out of 24 became tight after the addition of 0.01 per cent aluminum. On additions of 0.2 and 0.02 per cent aluminum, no disc became pressure tight.

In the presence of aluminum, the melt becomes more viscous. Fluidity normally increases with a rise in temperature, and it has therefore been questioned whether it would not be more possible to attain pressure-tight metal if the pouring temperatures were kept comparatively high in melts containing aluminum. The experiments proved, however, that the same result was obtained in spite of pouring temperatures varying between 1060 and 1200 C for 0.02 per cent aluminum, and 1050 and 1130 C for 0.01 per cent aluminum.

b. Physical Properties: It is not only the pressure tightness which deteriorates in the presence of aluminum, but also the physical properties. Even with a content as low as 0.01 per cent this influence is pronounced both in gassy and comparatively gas-free melts. The result is shown in Table 2.

The number of melts with aluminum is limited, but the result shows a successive lowering of the tensile strength and elongation with increased aluminum content up to 0.2 per cent.

Properties After Aluminum Removal: In experiments for the purpose of determining the properties

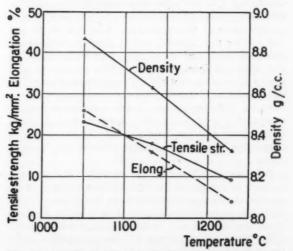


Fig. 5—Tensile strength, elongation and density as function of pouring temperature for 83-6-5-6 alloy.

TABLE 3-PRESSURE TIGHTNESS IN DISCS CAST AFTER REMOVAL OF 0.01, 0.02 AND 0.2 PER CENT ALUMINUM

Aluminum	Aluminum	No. of	No. of	Pres	Pressure Tightness							
Added,	After Removal, %	Melts	Discs		No. of Leaky Discs	Tight Discs,						
0.01	< 0.005	2	5	0	5	0						
0.02	< 0.005	2	6	1	5	16.5						
0.2	< 0.005	.7	19	17	2	89.5						

of the alloy after removal of aluminum by slag, aluminum in quantities of between 0.01 and 0.2 per cent was introduced as in the experiments for determining the influence of aluminum on mechanical properties and pressure tightness. Aluminum was subsequently removed, after which the discs and bars were cast. The pressure tightness data are recorded in Table 3, and the strength data in Table 4.

As shown in Table 3, it is possible to obtain pressure-tight discs after the removal of aluminum. In comparing Tables 1 and 3, it is seen that percentagewise a greater number of discs became pressure tight after the removal of 0.2 per cent aluminum than before the addition of aluminum. The number of pressure-tight discs before the aluminum addition was 42 per cent as against 89.5 per cent after the removal of 0.2 per cent aluminum.

On the other hand, a very small proportion of the discs became tight when 0.02 and 0.01 per cent additions were removed. But as only a very few melts were made with these low aluminum additions, it is unwise to draw any definite conclusions from them. It does, however, appear probable that pressure-tight metal would also be obtained after removal of aluminum of the order of some 0.01 per cent, as after removal of 0.2 per cent.

Aluminum Removal Improves Properties

In a comparison between Tables 2 and 4, it is found that a vital improvement in physical properties is achieved by the removal of aluminum. The most striking improvement was obtained in the metal containing 0.2 per cent aluminum before refining. The tensile strength has increased from 12.9 to 19.0 kg/sq mm, and the elongation from 3.5 to 10 per cent. After the removal of 0.01 per cent aluminum, tensile strength and elongation were higher than before the aluminum addition. The density is also higher, indicating that the melt had been degassed by the treatment with the aluminum remover, which contains oxidizing agents. Table 4 moreover demonstrates that a higher tensile strength was obtained with an increase in density, while in Table 2 this connection does not exist since aluminum is present

Influence of Silicon: In the investigations referred to later in the paper for developing a method of removing aluminum from melts, the metal has at times been contaminated by silicon. The removal of aluminum was carried out in silica-bonded graphite crucibles, from which silicon in small quantities sometimes dissolved and was found to be present in the discs and bars after casting.

The silicon has presumably influenced the properties of the gun metal in the subsequent tests. However, the experiments have proved that the metal has largely regained its normal properties after the removal of aluminum. The small quantity of silicon which has probably been present has therefore not exercised any pronounced influence on the properties of the metal.

In order to study the influence of silicon on the properties of gun metal more closely, small quantities of silicon were added to some melts and a number of bars and discs were cast. The additions were 0.01, 0.05 and 0.1 per cent silicon. In the case of the two higher silicon contents, only the physical properties were examined, while with the lower content the pressure tightness was also tested.

Judging from the results of the experiments, silicon does not possess the same deteriorating influence on the properties of gun metal as aluminum, either as to pressure tightness or physical properties. It is possible that 0.05–0.1 per cent silicon causes a slight deterioration in elongation. Pressure-tight discs have been cast in the presence of 0.01 per cent silicon.

TABLE 4-TENSILE STRENGTH, ELONGATION AND DENSITY OF D.T.D. BARS AFTER REMOVAL OF 0.01, 0.02

AND 0.2 PER CENT ALUMINUM

Aluminum	Aluminu	n		U.	T.S.	Elong-	
Added, %	After Re- moval, %				psi	ation,	Density, gm/cc
0.01	< 0.005	2	5	23.9	34000	16	8.89
0.02	< 0.005	2	6	19.9	28300	10	8.83
0.2	< 0.005	7	23	19.0	27000	10	8.69

ASTM Standards do not allow more than 0.003 per cent silicon in leaded gun metal. The D.I.N. Standards lay down no maximum silicon content. The tentative Swedish Standards allow maximum 0.003 per cent.

Removal of Aluminum from Brass and Gun Metal by Slag

The object of the experiments in removing aluminum as described below has been to examine the efficiency of different agents for aluminum removal and to prepare instructions for the best removal method.

The principle involved in the removal of aluminum by slag is the oxidation of aluminum in the melt to Al₂O₃ and the absorbtion of the oxide in the slag. The oxidizing process is performed by the oxidizing agents in the slags and the oxygen in the atmosphere. Fluorides of sodium, calcium, aluminum and of sodium aluminum (cryolite), possess the property, both individually and in combination, of dissolving aluminum oxide, so facilitating the removal of the oxide from the melt.

Methods Described in Literature: Attempts to remove aluminum from brass, nickel silver and gun metal have been made by Samson-Himmelstjerna,³ who used among other things slags of zinc oxide dissolved in borax.

Similar investigations have been carried out by Claus,⁴ who used a commercial aluminum removing agent. According to Claus the agent consisted of a mixture of sodium and aluminum fluorides and man-

ganese and iron oxides. His experiments were made on alloys of tin bronze, gun metal and brass.

In the investigations of both Samson-Himmelstjerna and Claus, a few hundredths of 1 per cent of aluminum generally remained in the metal after refining.

Since the aluminum contents in leaded gun metal, as allowed by American and German standards, are considerably lower, and moreover investigations by the Copper Committee of the Federation of Swedish Mechanical Engineering Industries have shown that even 0.01 per cent aluminum is sufficient to make it practically impossible to cast pressure-tight metal, it would appear that the means of aluminum removal found by Samson-Himmelstjerna and Claus were inadequate.

Experimental Procedure: Refining experiments were first carried out on a brass (copper 58-60, lead about 3 per cent, balance zinc) with a comparatively low melting point (solidification range). However, it was discovered at an early stage that an alloy which melted at so low a temperature could only with difficulty become free of aluminum, since the slags were insufficiently fluid at a temperature when moderate zinc fumes occurred.

Slags Tested for Removal Efficiency

Subsequent experiments were carried out on a brass with higher copper content (80-85 per cent), and tests were made to find efficient slags with this alloy. Determinations were made both of the minimum quantity of slag required at constant refining time, and of the shortest refining time, to reduce the aluminum to less than 0.01 per cent for a quantity of slag five times that of the aluminum in the melt. The most efficient agents were finally tested on 83-6-5-6 gun metal.

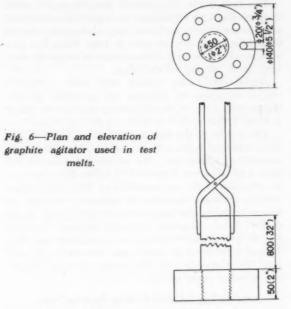
The experiments were made with 50-kilo melts. Melting was done in graphite crucibles in a cokefired furnace. The metal was protected against oxidation during melting by a covering of charcoal. After melting and heating to a suitable refining temperature, the charcoal was skimmed off and the slag added and mixed with the melt.

In the first experiments the slag was stirred in with an ordinary round graphite rod. It later proved advantageous to employ a special agitator (Fig. 6), with which more efficient mixing of the slag into the melt was obtained.

Samples were taken from the melt at given intervals with a spoon and cast in a small mold for spectrographic determination of the aluminum content,

TABLE 5-COMPOSITION OF ALUMINUM REMOVAL
AGENTS USED

		Co	mpositio	on of Sla	g, %		
NaF	CaF ₂	Na _s AlF ₆	Na ₂ SO ₄	Na ₂ CO ₃	Na ₃ B ₄ O ₇	ZnO	MnO
30	20	20	20	10	-	_	_
	50	20	20	10	_	_	-
	70		20	10	_	_	_
-	-	50	_	_	25	25	_
25	15	15	15	10	_	_	20
	30	30 20 - 50 - 70	NaF CaF ₂ Na ₃ AlF ₆ 30 20 20 - 50 20 - 70 - 50 - 50	NaF CaF ₂ Na ₃ Na ₅ SO ₄ AlF ₆ 30 20 20 20 - 50 20 20 - 70 - 20 - 50 - 50 -	NaF CaF ₂ Na ₃ Na ₂ SO ₄ Na ₃ CO ₈ 30 20 20 20 10 — 50 20 20 10 — 70 — 20 10 — 50 — — —	AIF _e B ₄ O ₇ 30 20 20 20 10 — 50 20 20 10 — 70 — 20 10 — 50 — 25	NaF CaF ₂ Na ₈ Na ₅ SO ₄ Na ₂ CO ₈ Na ₂ B ₄ O ₇ ZnO 30 20 20 20 10 — — — 50 20 20 10 — — — 70 — 20 10 — — — - 50 — — 25 25



and for study of the influence of aluminum on the surface appearance of the sample.

Slags Used: In the experiments for aluminum removal, tests have been made both with commercially available agents and with compounds made in the laboratory. The compositions of slags are shown in Table 5.

Influence of Temperature on Refining Speed: The influence of temperature on the speed of refining was studied in two experiments, using a brass containing 58-60 per cent copper. The temperature of the melt was in the one case maintained at about 900 C, and in the other at 1050 C. In both experiments 0.4 per cent aluminum and 1.2 per cent of slag (A) were added. The slag was mixed in with a graphite rod.

At the lower temperature the aluminum content decreased from 0.38 to 0.25 per cent in 30 min. At the higher temperature a marked improvement was obtained with an aluminum reduction from 0.35 to 0.035 per cent in the same time.

A melt at an even higher temperature (1180 C) was made with an 85 per cent brass, to which 0.4 per cent aluminum and 1.2 per cent slag had been added. As before, the slag was mixed in with a graphite rod. The aluminum content fell from 0.45 to below 0.01 per cent after 30-min agitation. These experiments confirm what could be expected—that a high temperature is beneficial to refining speed.

Efficiency of Different Types of Slags: Determination of the minimum quantity of slag necessary to remove 1 per cent of aluminum was made in an 85-15 copper-zinc alloy. The refining time generally was 1 hr. The slag quantity was reduced from test to test according as it was established that complete removal of aluminum had taken place. It proved to be most important that the slag was properly mixed into the melt.

The test results were somewhat widespread, which may be attributed to varying intensity in agitation and certain variations in melt temperatures.

The tests made it apparent that, for equal additions of slag and aluminum, it is possible under certain circumstances to lower the aluminum content from 1 to below 0.01 per cent in 1 hr. With this proportion of slag a long refining time is required, even with good mixing of the slag.

Appreciably better mixing than with a graphite rod is obtained by utilizing the graphite agitator illustrated in Fig. 6, which consists of a rod to which is fixed a disc with a number of drilled holes.

The results of the tests are shown in Table 6. For an addition of 2 per cent of slag (A), being twice the amount of aluminum, the aluminum content fell after 1-hr agitation from 0.94 to below 0.01 per cent. A reduction of the slag addition by half, while maintaining the same intensity of agitation, was not sufficient to bring the aluminum content down to the desired figure. However, with the agitator shown in Fig. 6, and equal additions of aluminum and slag, adequate removal of aluminum was achieved with slag (A). Slags (B) and (D) proved the most efficient of the other types tested.

Large Slag Addition Reduces Refining Time

A refining procedure as described above, i.e., with small slag additions in relation to the amount of aluminum in the melt, is nevertheless uneconomical due to the large metal losses and high fuel costs occasioned by the long refining time. A larger slag addition makes refining quicker. In a new series of tests on an 85 per cent brass with 0.3 per cent aluminum, a slag addition was made of five times the aluminum quantity. Only the best slags from the foregoing tests, namely, (A), (B) and (D), and a new slag (F), were included in the new series.

The results shown in Table 7 indicate that agitation for 10 min, using the best slags, sufficed to lower the aluminum content from 0.3 to less than 0.01 per

cent.

Control on Gun Metal: The test results were finally controlled in an 83-6-5-6 gun metal. Additions of 0.3 per cent aluminum and 1.5 per cent slag of types (A), (D) and (G) were made to the melts.

Generally speaking, the same results were obtained

as in the tests on brass, in that the aluminum content went down from 0.3 to below 0.01 per cent in 10 min. Samples taken after 5-min refining showed incomplete removal of aluminum, but the content did not amount to more than a few hundredths of 1 per cent. The results are given in Table 8.

Silicon Absorption in Refining—Removal of Silicon: It was noticed that the metal, after some of the refining tests, contained a small amount of silicon, in certain cases amounting to 0.1–0.2 per cent.

Experiments established the fact that the silicon originated from the silica-bonded crucibles used in the tests. The slag released silica, which was then reduced by the aluminum present in the melt and dissolved in the metal in the crucible.

It proved that aluminum must be present in the melt for silicon to be absorbed. No silicon has been observed in test melts treated with fluoride-containing slag without addition of aluminum. However, it may be assumed that silica is also released when aluminum is not present in the melt, but that in this case it is either directly absorbed in the slag or reacts with the slag to give off silicon fluoride.

Experiments have also shown that silicon can be removed from the metal by the slag used, if the metal is free from aluminum. The silicon content does not, however, fall with quite the same speed as aluminum. Thus it is possible to first remove the aluminum, and subsequently the silicon from the crucible material. Between the refining operations an exchange of

slag is recommended.

Manganese Absorption in Refining: In refining operations with slags containing fluorides with manganese dioxide as oxidizing agent, the melt has absorbed manganese. After additions of 0.3 per cent aluminum and 1.5 per cent slag (B), the melt contained about 0.4 per cent manganese after completion of refining. If the melt absorbed the entire manganese content of the slag, the slag must have contained about 40 per cent manganese dioxide.

ASTM makes no mention of the manganese content allowed in leaded gun metal, while D.I.N. Standards permit 0.2 per cent. The tentative Swedish

Standards allow 0.1-0.2 per cent Mn.

TABLE 6-ALUMINUM REDUCTION IN A 50-KILO (110-LB) MELT OF AN 85 PER CENT BRASS, REFINED WITH DIF-FERENT SLAGS. APPROX. 1 PER CENT ALUMINUM ADDED

Slag	Slag Composition	Slag, %	Temp., C		% Al C	content at	Different	Refining	Times, m	in
				0	10	20	30	40	50	60
A*)	30% NaF. 20% CaF ₂ , 20%	2	1170	0.94	0.44	0.28	0.18	0.12	0.05	< 0.01
A*	NagAlF, 20% NagSO4,	1	1170	1.00	0.84	0.37	0.29	0.23	0.18	0.12
A.	10% Na ₂ CO ₂	1	1130	0.99	0.71	0.62	0.54	0.45	0.36	0.28
A		1	1150	0.83	0.47	0.33	0.27	0.11	0.02	< 0.01
B	Fluorides and MnO ₂	1	1180	0.73	0.24	0.18	0.14	0.09		
C	50% CaF ₂ , 20% Na ₈ AlF ₄ , 20% Na ₂ SO ₄ , 10% Na ₂ CO ₈	3	1100-1150	0.89	0.52	0.37	0.37	0.21	0.08	0.03*
D	70% CaF ₂ , 20% Na ₈ SO ₄ , 10% Na ₈ CO ₅	1	1170	1.05	0.59	0.40	0.31	0.16	0.11	0.06
E	50% Na ₃ AlF _e , 25% Na ₂ B ₄ O ₇ , 25% ZnO	2	1100-1150	0.98	0.67	0.57	0.43	_	0.20	

^{*} Agitator—round graphite rod. In the other tests an agitator was used as shown in Fig. 6.

^{** 55} minutes.

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Sulphur Absorption in Refining: Some of the refining agents employed contain sodium sulphate as oxidizing agent. Since even comparatively small quantities of sulphur have a detrimental effect on the properties of the material, and the melt may absorb sulphur from the sodium sulphate, an investigation was made as to whether a deterioration in material properties might have occured through sulphur absorption.

In a 50-kg melt of 83-6-5-6 gun metal, 0.3 per cent aluminum and 1.5 per cent slag (A) were added. The sulphur content was determined before and after refining. Sulphur contents of <0.001 and 0.014 per cent, respectively, were found. The sulphur in the slag was estimated at 0.07 per cent of the weight of the melt, so that only a part of the sulphur had been absorbed.

ASTM and the tentative Swedish Standards allow 0.08 per cent sulphur, and D.I.N. 0.05 per cent;

TABLE 7—ALUMINUM REDUCTION IN A 50-KILO (110-LB) MELT OF AN 85 PER CENT BRASS, REFINED WITH DIFFERENT SLAGS. 0.3 PER CENT ALUMINUM AND 1.5 PER CENT SLAG ADDED. AGITATOR AS SHOWN IN FIG. 6

Slag	Slag T Composition	Cemp.,	%				at Diff	
	•		0	2	4	_ 5	10	15
A)	30% NaF. 20% CaF,	1170	0.28		_	_	0.015	< 0.01
A	20% NasAlF. 20%	1150_	_					
	Na ₂ SO ₄ , 10% Na ₂ CO ₃	1200	0.30	0.08	0.03	0.02		
A		1050-						
1		1100	0.30	0.14	0.06	0.04		
A		1150	0.27	0.11	0.05	0.03		
A		1050	0.34	_	_	0.06		
A		1090	0.27	_			< 0.01	
A		1180	0.21	-	<	0.01		
	Fluorides and MnO ₂	1100*	0.33		_	0.06	0.02	< 0.01
B		1100*						
D)	70% CaF2, 20% NasO4	1150*	0.27	_	_	_<	0.01	
	10% Na ₂ CO ₃							< 0.01
F	25% NaF, 15% CaF,	1240	0.24	0.03	0.01			
	15% Na AlF 15%							
	Na ₂ SO ₄ , 10% Na ₂ CO ₃							
	20% MnO,							
*	Approximate							

therefore, no problems caused by sulphur need be expected. Refining was carried out in a coke-fired crucible furnace, and it is therefore conceivable that part of the sulphur in the metal came from the furnace gases.

Slag Attack on Crucibles: Crucibles are attacked by refining agents that contain fluorides. This action, which is confirmed by the release of silica referred to earlier, has been determined approximately by weighing a crucible before and after a number of refining operations. After ten operations with slag (A), a silica-l-onded graphite crucible, weighing 23 kg (50 lb) had lost 5.5 kg (12 lb) in weight. The action by the slag has been found to be due not only to the slag quantity and the aluminum present, but also to the temperature and fluidity of the metal, and thus of the slag.

Hygroscopicity of Different Compounds: The hygroscopicity of slag (A) and of its chemical constituents was determined by normal methods. The following

Table 8-Aluminum Reduction in a 50-kilo (110lb) Melt of 83-6-5-6 Gun Metal, Refined with Different Slags. 0.3 Per Cent Aluminum and 1.5 Per Cent Slag Added. Agitator as Shown in Fig. 6

Slag Slag Composition	Temp.,		Content efining		
		0	2	5	10
A) 30% NaF, 20% CaF,	1120	0.3*	_	0.02*	< 0.01
A (20% Na ₃ AIF ₄ , 20% Na ₂ SO ₄ , 10% Na ₂ CO ₅	1100	0.3*	0.1*	0.04*	
D 70% CaF2, 20% NasSO4	, 1100	0.3*	_	0.03*	< 0.01
10% NagCO ₃	1100	0.3* (0.05 - 0.1	0.05	
F 25% NaF, 15% CaF ₂ 15% Na ₂ AlF ₄ , 15% Na ₂ S 10% Na ₂ CO ₂ , 20% Mn ⁴ Approximate		0.3*	0.1*	0.02*	

results were obtained:

Compoun												1	H	13	g	T	0	54	00	F	ì	C	i	y	(water, %)
NaF .	 						 . ,	 																	.0.5
CaF .				*			 	 		*															.0.01
Na _a AlF							 	 																	.5.0
Na,SO,								 			*														.4.9
Na ₂ CO ₂							 	 																	.4.5
Slag A								 																	.3.2

The fluorides and fluoride slags should be kept protected against water absorption.

Recommended Foundry Procedure for Aluminum Removal from Brass and Gun Metals

After melting the aluminum-containing alloy under a suitable cover, the melt is heated to refining temperature (about 100 C above the liquidus temperature). The cover is skimmed off and 1–1.5 per cent of the aluminum removal agent is added. The addition should be made in portions under careful agitation to avoid splashing of the metal in the event of the slag being moist. When the whole slag has been added, the melt is violently agitated, e.g., with an agitator such as shown in Fig. 6.

After about 5-min agitation a suitable sample is taken from the melt. With some experience it is possible to judge, with fairly high accuracy, from the appearance of the solidified specimen whether aluminum still is present. If even very small amounts of aluminum are present (0.01–0.02 per cent) the appearance of the specimen will differ slightly from that of aluminum-free metal. An aluminum-containing specimen has a comparatively bright and smooth surface, while the surface of an aluminum-free metal is duller and rougher (Figs. 7 and 8).

If the test casting shows that aluminum still is present, the melt is stirred for a few minutes longer with or without a new addition of slag, after which a new sample is taken. When the appearance of the specimen resembles that of the specimen in Fig. 8 the aluminum content is quite low. Since 0.01–0.02 per cent aluminum seems to be the lowest content that indicates the presence of aluminum in a test casting, refining must continue a minute or so longer, after coming below this aluminum content, in order to ensure that aluminum removal is complete. If there is any doubt about the amount of aluminum remaining, it is recommended that refining continue a minute or so longer than is really necessary.

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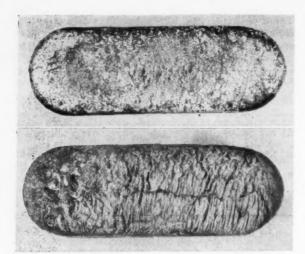


Fig. 7 (top)—Test specimen containing 0.2 per cent aluminum. Bright, smooth surface. Fig. 8 (bottom)-Test specimen containing < 0.01 per cent aluminum. Dull, rough surface.

Before pouring, dross and slag should be skimmed off the melt. There may be some difficulty in doing this if the slag is very fluid, but it can be thickened with sand and is then easier to skim. The melt should be left standing awhile before pouring in order to allow any oxides present to rise to the surface. The foundry treatment with slags must, of course, vary with the equipment and control measures available.

Summary

The influence of aluminum on the strength and pressure tightness of 83-6-5-6 gun metal has been examined. Different refining agents have been tested for removal of aluminum from brass and gun metal melts. The findings are as follows:

1. Strength and pressure tightness deteriorate with only 0.01 per cent aluminum present.

2. If a gun metal melt containing aluminum is treated with a slag-forming refining agent so that complete aluminum removal is achieved, normal strength and pressure-tight metal is again obtained.

3. The most efficient aluminum removing agents were slag (A) with 30 per cent NaF, 20 CaF₂, 20 Na₃ AlF₆, 20 Na₂SO₄, 10 Na₂CO₃; slag (D) with 70 per cent CaF2, 20 Na2SO4, 10 Na2CO3; and slag (F) with 25 per cent NaF, 15 CaF₂, 15 NA₃AlF₆, 15 Na₂SO₄, 10 Na₂CO₃, and 20 MnO₂.

4. A content of 0.3 per cent aluminum in gun metal or 80-85 per cent brass can be brought under 0.01 per cent in 10 min by adding 1.5 per cent slag.

5. Refining speed is dependent on the amount of slag added, the temperature of the melt and the intensity of agitation.

6. An opinion of the aluminum content in the melt can be obtained by casting small test specimens and studying the surface of the frozen metal.

7. If the slag contains manganese dioxide as an oxidizing agent, the melt absorbs manganese during refining.

8. Fluoride slags release silica from silica-bonded

graphite crucibles. The silica is reduced by aluminum and dissolved in the metal.

9. Silicon does not have the same harmful effect as aluminum on the material properties.

10. Silicon, like aluminum, can be removed by treating the melt with slag. Aluminum must be removed first, after which a new slag is prepared.

11. Sulphur absorption when refining with slag containing sodium sulphate as an oxidizing agent is not of a serious order.

12. Recommendations are made on methods of removing aluminum under practical conditions.

Bibliography

1. W. A. Baker, F. C. Child, W. H. Glafsher, "The Effect of Shrinkage and Gas Porosity on the Pressure Tightness and Mechanical Properties of Bronze Sand Castings," Journal, Inst. of Metals, 70 (1944). pp. 373-406.

2. E. C. Mantle, "Leaded versus Lead-Free Gunmetals," Met-

al Industry, Sept. 28, 1951, pp. 255-256.

3. H. O.v.Samson-Himmelstjerna: "Die Entfernung von Aluminum und Eisen aus Kupfer und Kupferlegierungen," Zeitschrift für Metallkunde, May, 1937, pp. 152-154.

4. W. Claus, "Zur Kenntnis der Entfernung von Aluminum aus Kupferlegierungen," Zeitschrift für Metallkunde, July,

1940, pp. 239-242. *
5. A. V. Larsson, "Avlägsnande av aluminum ur mässingoch rödgodssmältor med oxiderande och oxidlösande slagg, Gjuteriet 39 (1949), pp. 181-188.

6. Å. V. Larsson, and B. Löfgren, "Avlägsnande av aluminum ur mässing- och rödmetallsmältor med oxiderande och oxidlösande slagg-Del II," Gjuteriet 41 (1951), pp. 143-151.

DISCUSSION

Chairman: G. P. HALLIWELL, H. Kramer & Co., Chicago. Co-Chairman: H. J. Roast, London, Ont., Canada. Recorder: R. A. COLTON, American Smelting & Refining Co.,

J. G. Kura (Written Discussion): 1 Mr. Larsson has presented very interesting data on the influence of aluminum on the properties of cast gun metal. The method developed to remove aluminum from the melt appears to be quite efficient.

Because neither metallographic nor radiographic examination was made of the leak-test discs, there may be some question as to the actual cause of leakage. The turbulence encountered with top pouring and rapid solidification of the test disc could readily result in the formation of dross and entrapment of dross and gas bubbles.1 Shrink cavities may also be the cause of leakage, and, in this test piece, their occurrence would be dependent upon the amount cf gas in the melt and the pouring temperature.1,2 If the gas content of the melt is very low, if the pouring temperature is low, and if feeding of the casting is not adequate, shrink cavities are apt to form. Examination of the discs for the cause of leakage may result in a better concept of how the presence of aluminum promotes leakage.

The author has shown that the addition of aluminum has a marked deleterious effect on tensile properties and density. It is known that the D.T.D. test-bar design is relatively insensitive to melt quality and normally yields low tensile properties.3 Therefore, it must be assumed that the deleterious effect of aluminum is, indeed, very strong.

It is not unusual to find no clear correlation between the density of a tensile bar and the leak tightness of a separate casting. Comparing the data in Table 1 with those in Table 2, it appears that the correlation follows somewhat the accepted trend depicted in Fig. 4. In other words, too law a gas content can result in excessive shrinkage which, in turn, will not materially lower the density but could be the cause of leakage. Thus, with reference to the data in Table 1, the low gas content of the melt rather than the addition of 0.01 per cent aluminum may be the reason for the decreased leak tightness. This reasoning may also be applicable to the specific data in Tables 3 and 4, which deal with melts containing additions of 0.01 per cent aluminum.

Tables 3 and 4 show that leak tightness and tensile properties

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are improved by the removal of the aluminum. The disturbing observation is that the density of the tensile bars is of the same order after removal of aluminum as it was before the treatment. If the treatment had resulted in degasing the melts, all of the density values in Table 4 should have been of the same order of magnitude, say 8.88. Furthermore, it would appear from the data that a higher incidence of leak tightness was obtained at the lower density levels. The lower density values were obtained with the treated melts which originally had the highest aluminum content. Perhaps the author may have some explanation for these disturbing observations.

REFERENCES

1. J. G. Kura and L. W. Eastwood, "Effects of Gating Practice on the Leak Tightness of 85-5-5-5 and 81-3-7-9 Alloy Castings"; A.F.S. Preprint 52-67, A.F.S. Transactions, vol. 60 (1952).

2. J. G. Kura and L. W. Eastwood, "Effects of Mold Material on Leak Tightness and Mechanical Properties of 85-5-5-5 and 81-3-7-9 Alloy Castings"; A.F.S. Preprint 52-55, A.F.S. TRANSAC-TIONS, vol. 60 (1952).

3. "Te t Bars for 85-5-5-5 Alloy, Their Design and Some Factors Affecting Their Properties"; A.F.S. Transactions, vol. 54, pp. 1-36 (1946).

R. A. COLTON (Written Discussion): On numerous occasions I have had the opportunity to observe a corollary effect of aluminum in tin bronze and leaded tin bronze alleys. It should be noted that aluminum is a strong deoxidizer and undoubtedly will scavenge for oxygen in many of these common copper-base alloys. On several occasions we have encountered unusual shrinkage conditions in bronze castings which upon thorough investigation was found to be due to aluminum picked up as an impurity during the melting operation.

The most striking example of what a small amount of aluminum can do is the extraordinarily deep pipe found both in the risers and in the sprue of tin bronze and leaded tin bronze castings which have been unintentionally deoxidized with aluminum. Apparently, the effect is to degas the metal rather thoroughly by removing the oxygen available for the steam reaction. This action is so complete that metals which normally never exhibit deep shrinkage cavities or pipes exhibit a shrinkage more characteristic of narrow freezing range alloys.

In every case where we have found this extraordinary shrinkage, which many times has caused castings to be rejected, upon careful analysis we have found aluminum to be present in quantities up to 0.10 per cent. Usually this much aluminum can be detected on the surface of the casting as a silvery film. If the aluminum content runs 0.01 per cent or less, however, it may be difficult to detect it by appearance but its effect may be seen, nevertheless, in the unusual shrinkage pattern.

It is only comparatively recently that we have had analytical methods sensitive enough to detect aluminum contents below 0.10 per cent with any degree of accuracy. Since these techniques have been available, we have been able to attribute these cccasional unusual shrinkage cavities to the presence of aluminum as contamination in these alloys.

R. H. GILBERT: 2 Mr. Laisson's findings confirm our research data on the effect of aluminum on leak tightness of 81-3-7-9 alloy castings. A few hundredths of a per cent of aluminum seriously affects the density of the alloy and leakers become excessive. Castings that normally tested 5 per cent leakers, leaked 100 per cent when 0.05 per cent aluminum was added to the melt. Incidentally, the fractures were excessively discolored and while tensile properties were never determined it was obvious, from the extent of microporosity that was present, that they would be extremely low.

I am interested in the mechanism that is involved and am wondering if Mr. Larsson has any information that would shed

any light on this phase of the subject.

W. B. Scott: The role of the aluminum is to decrease the tin solubility resulting in a continuous tin-rich phase predominating in effecting the physical properties toward a brittle weak material. This comment is based on private unpublished obser-

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DEVELOPMENT OF STEEL FOUNDRIES IN INDIA

By

N. G. Chakrabarti*

India with all her past glory of civilization with philosophies, architects, arts and culture was not lacking in the art of forming metal to shape. Although an agricultural country, India also had her own industrial activities on metal industries particularly iron and steel. The late great Indian economist Justice V. R. Ranade in his "Essays on Indian Economics" (First edition pp. 159-160) wrote on Indian iron and

steel industry the following:

"The iron industry not only supplied all local wants, but it also enabled India to export its finished products to foreign countries. The quality of the materials turned out had also a world-wide fame. The famous iron pillar near Delhi, which is at least 15 hundred years old, indicates the amount of skill in the manufacture of wrought iron, which has been the marvel of all who have endeavored to account for it. Mr. Ball (late of Geological Survey of India) admits that it is not many years since the production of such a pillar would have been an impossibility in the largest factories of the world, and even now, there are comparatively very few factories where such a mass of metal could be turned out. Cannons were manufactured in Assam of the highest caliber. Indian steel furnished the materials out of which Damascus blades with a world-wide reputation were made; and it paid Persian merchants in those old times to travel all the way to India to obtain these materials and export them to Asia. The Indian steel, found once considerable demand for cutlery even in England. The manufacture of steel and wrought iron had reached a high perfection at least two thousand years ago."

Recent history of the Indian steel foundry industry has linked up with the development of her railways, some time in the later part of the 19th Century.

The first steel foundry in the country was established in 1890 in a railway workshop which is now known as the Central Railway. An acid Tropenas convertor of 2 tons capacity was installed mainly for the purpose of producing steel castings for the maintenance of locomotives, carriages, wagons, etc. At that time pig iron was not available in India and hematite pig iron and refractories used to be imported from the United Kingdom. It is interesting to note that this foundry is still in existence producing about 40/50 tons of steel castings per month mainly for internal consumption by the railway. Between the period 1890 and 1920, the major iron and steel producing units were installed and developed in the country.

The success obtained by the promoters of these plants enccuraged Indian business men to develop other subsidiary branches of iron and steel industry and steel foundry industry is one of them. Between the period of 1920 and 1940, seven commercial steel foundries have been set up in the country by private enterprise, three steel foundries by the Defense Department of the Govt. of India and one state sponsored. Leaving aside the foundries of the ordnance factories which are mainly engaged on defense requirements there are altogether nine foundries in the country with a productive capacity of 12,000 to 13,000 tons of black castings per annum. In addition to this the two major Indian iron and steel producers have their own steel foundries for the purpose of meeting their own maintenance demands,

I would like to mention here a brief picture of the working of the Indian steel foundries.

Melting Units

The melting processes employed by our steel foundries are as follows:

1. Acid Tropenas convertor in one foundry (2 tons capacity).

2. Acid open hearth in one foundry (Two 5-ton capacity).

3. Basic openhearth in one foundry (5-ton capacity).

4. Basic electric Herolt-type in 6 foundries (1-ton to 10-ton capacities).

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In these commercial steel foundries only 4 per cent of the total steel is produced by acid Tropenas process, 25 per cent by acid openhearth, 12.5 per cent by basic openhearth and 60 per cent by basic electric process.

Raw Materials

Basic pig iron and scrap steel—The only Indian acid Tropenas convertor entirely depends for its raw material upon selected steel scrap containing very low sulphur and phosphorus. These steel scraps are melted in a cupola with sufficient Fe-Mn, Fe-Si and high coke ratio and then brought to the Tropenas and blown. Formerly, this particular melting unit used to be fed by English hematite pig iron.

At present due to high cost and very short supply of English hematite pig iron, Indian acid furnaces except ordnance factories have to depend entirely on selected steel scrap, as hematite pig iron cannot be produced in India due to bad quality of Indian coke which contains high percentages of sulphur and phosphorus though very good quality hematite iron ore is available. So, the Indian acid openhearth furnaces also have to depend on 100 per cent selected steel

The basic openhearth furnace in one of the foundries generally uses 50 per cent steel scrap and 50 per cent basic pig iron. The Indian basic electric steel furnaces use 100 per cent ungraded steel scrap as this is the cheapest material available in abundance. Cost of power so far as electric furnaces are concerned is comparatively high but is compensated for to some extent by using comparatively cheap raw material. These basic electric furnaces are remelting and refining units.

TABLE 1-ANALYSIS OF INDIAN COKE

		A			В		
Fixed Carbon, %	78	to	82	65	to	70	
Ash, %	14	to	18	20	to	25	
Volatile matter and moisture, %		4			4		

 TABLE 2—ANALYSIS OF INDIAN IRON ORE, %

 SiO2
 Fe2Os
 FeO
 Al2Os
 MnO
 CaO
 P3Os

 0.90
 94.2
 0.47
 0.98
 0.18
 0.26
 0.072

Iron Ore and Mn-Ore

Iron ore and Mn-ore required for the Indian foundry melting units are available very cheaply as huge deposits of these ores are available in our country.

Fuels—The quality of Indian metallurgical coal and coke is not very good due to the high ash content and the high percentage of sulphur and phosphorus. They are however available in abundance in our country. The openhearth furnaces of our country are fueled by producer gas which is made in producer plants near the furnaces using Indian steam coal and Indian hard coke and an attached boiler. The composition of producer gas is as follows (by volume):

CH₄—3 to 5 per cent, CO₂—8 to 10 per cent, H₂—13 per cent, O₂—2 to 6 per cent, CO—18 to 22 per cent, and N—Balance.

The electric furnaces obtain the power supply from city electric supply companies at industrial unit rates. None of these foundries have their own power plants.

Refractories and Fluxes

Refractories both acid and basic such as ganister, silica sand, magnesite, dolomite, etc. are available in India. Silica bricks, fire-bricks, magnesite bricks, etc. are now manufactured for the lining and roof making of these furnaces. Fluxing material such as limestone and lime is available in India in abundance. Good quality fluorspar however is not available in India.

Ferro-Alloys

Ferro-alloys such as Fe-Mn and Fe-Si are now manufactured in India and new plants for manufacture of Fe-Mn are going to be set up in our country. This will not only make India self-sufficient so far as this important item is concerned but also a major portion of the same will be exported to earn more dollars for our national industrial development. As regards Fe-Si, we are not fully self sufficient and have to depend partially on Canada for the supply of this important material.

Other ferro-alloys such as Fe-Ti, Fe-Va, Fe-Cr, etc. are not manufactured in India and our foundries depend entirely upon import from foreign countries.

Carbon Electrodes for Electric Furnaces

For the supply of carbon electrodes for electric furnaces our country is dependent entirely on foreign supplies particularly from U. S. A. and small percentages from Germany and Japan.

The Furnaces

The capacities of the furnaces in Indian foundries varies from one to ten tons. All except one are hand charged units. One 10-ton unit however is provided with top charging arrangements by overhead crane. The foundry industry has not been developed in India in a planned way as a result of which there has been surplus melting capacity in each foundry producing liquid steel by electric furnaces. This surplus liquid steel is cast as ingots for rolling and invariably every foundry producing steel by electric furnaces had had to put up rolling mills to consume the ingots thus produced. As a result the rolling mill industry has been developed side by side with the foundry industry where electric furnaces are used as melting units. The foundries with Tropenas and acid and basic openhearths as melting units utilize the whole of the liquid steel produced, for the consumption of their foundries only.

In the Field of Modern Development

The latest development in the field of melting practice of steel, such as injection of oxygen for minimizing oxidation period, raising bath temperature without loss of time, minimizing the loss of oxidizable costly metals such as chromium in case of alloy steel making, is not generally employed except in the field of experiment in some of the foundries.

Alloy Steel

In addition to ordinary plain carbon steel castings some of the foundries produce austenitic Mn-steel castings and Ni-Cr steel castings according to the demands of the specification.

Pattern Making

Pattern making is an integral part of steel founding in India as it is in all countries but commercial pattern making is not so popular yet. Many good carpenters were available in India during the period when foundries were being developed in the country. The European technicians of Indian railways and ship yards trained many of these carpenters to produce patterns as per blue prints. At the present time good pattern makers are abundant in the country. The former Indian pattern shops were not well equipped and the pattern makers used simple tools such as hand-saw, chisels, jack-planes, etc. for their bench work.

Now, at the present time, almost all the foundry pattern shops are well equipped with band-saws, circular saws, cross-cut saws, planing and thicknessing machines, lathes, etc. Recently, there has been some commercial pattern shops whose business it is to supply patterns to customers as per specifications. The bulk of the work done by these pattern makers is for iron foundries only; but they also undertake jobs for steel foundries too. But reliable commercial pattern makers for steel castings are limited in numbers.

At the present time Indian pattern makers are quite competent to prepare patterns and core boxes as per drawing, to mount these patterns on pattern plates, avoiding cross-joints, making arrangements for the methods of gating and feeding and contraction and finishing allowances.

Pattern materials: Materials used for pattern making are mainly wood but, metal patterns are also used in cases when large numbers of castings justify the use of metal. Cement, plaster and plastics are very rarely used. Generally seasoned Burmah teak timber is used for pattern making in India. If seasoned properly, the expansion and contraction due to seasonal variation of temperature and humidity is almost negligible. Patterns and core boxes made from this timber can be retained for 15 to 20 years without much disintegration or crack. Presently however due to the limited supply of Burmah teak timber, seasoned C.P or Assam teak are also used for pattern making. Metals used for pattern making are generally duralumin, brass and cast iron depending on the nature of the work and the method of molding to be employed.

Coloring of Patterns: There is no Indian standard specification for painting and varnishing patterns but every foundry has its own color code. Generally small patterns are varnished and larger ones are painted red to indicate machined surfaces, yellow for core-prints and black or gray for other surfaces.

Pattern stores: In every foundry, a proper pattern store is maintained for storing the old patterns and core boxes with a simple method of indexing.

Foundry Sand

As in every other country, in India also foundry sands are produced by weathering and disintegration of massive igneous rocks. Sources and qualities of sands for steel foundries are Jabbulpore silica, Rajmahal quartz, river sand and Damodar rock sands. These sands differ in chemical compositions, grain size and clay contents. The latter contain high percentages of clay and are generally mixed with the former varieties for different types of work. Plasticity, green and dry strength, permeability, refractoriness and grain size and shape are the qualities of foundry sand. Bentonite powder, full bond, molasses, dextrin, linseed oil, etc. in different proportion are also used in foundries for preparation of different sands for green and dry sand molding and coremaking.

ANALYSIS OF SAND

Source	SiO	Fe ₂ O ₂ , Al ₂ O ₂ etc
	%	% %
Taljuri	90	9
Jubbulpore	95	5
Yellow Sand	70	30

GRAIN SIZE OF SAND

				1	Mesh			
Source	8	10	16	22	30	44	60	100
Taljuri			0.6	10	28	37	13	6
Jubbulpore	6	2.6	16	17	19	13	14	8

In India, there was no sand testing equipment for scientific control of foundry sands and the molders used to guess the properties of the prepared sands simply by feeling and pressing the same in their palms.

Presently, almost every steel foundry is well equipped with a modern sand laboratory containing equipment for making sand test specimens, green and dry permeability, green and dry strength, shatter test, moisture tester, sand drying ovens, laboratory sand mill for making new batches of sands for standarizing the foundry sands, clay determining equipment, physical balance, riddles for grading sands and arrangements for chemical testing of sands. Silica sands and quartz sands containing 95 per cent to 98 per cent SiO₂, and rock or yellow molding sands containing 10 per cent to 14 per cent of clay substance are generally used for molding purposes, whereas river sands with comparatively small grain size and uniform grain shape are used for core-making. Previously, in Indian foundries, only roller type pug mills used to be employed for sand mulling. At present improved types of plow scraper, kneeding roller, and aerating type mixers have been installed in various foundries with better sand mulling practice. For mixing core sand, core sand mixers have been installed. Dry sand molding practice is widely employed in all the steel foundries in India with only one foundry successfully carrying out green sand molding. This improved equipment made it possible to have better control of

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foundry sand which has ultimately made the green sand practice a success at least in one foundry in our country.

					1	Dry
	Perme- ability	Flow- ability	Strength, psi	Moisture %	Perme- ability	
Green Sand						
Facing	150 to	70 to	11 to	4 to		
	200	72	15	4.5		
Backing	120 to	72 to	10 to	3.5 to		
	160	75	14	4.5		
Dry Sand	60 to	74 to	3.5 to	6 to	80 to	180 min
Facing	70	80	6	8	100	
Backing	36 to	68 to	5 to	6 to	52 to	180 min
	48	72	8	8	60	
Core Sand	48 to	68 to	6 to	6 to	52 to	160 min
	52	72	8	7.5	60	
Core Sand					-	
Dextrine & Linseed Oil	119	75	3.8	7.5	226	340 min

Sand Handling

Prior to World War II, manual handling of foundry sand was the only handling process. Now, after the termination of the War, labor problems are becoming more and more complicated. This situation coupled with high wages, is inducing the Indian industrialists to introduce various labor saving devices and the mechanization of foundries particularly sand handling devices such as belt conveyor, magnetic separator, sand elevator, rotary screen, bucket loader, storage hopper, sand mills, core sand drier, and sand distribution arrangements are being introduced in Indian foundries.

Molding and Core-Making

After the patterns and core-boxes are made they are finally checked by making sample molds and sample castings with proper molding sand. From these sample castings the accuracy of the patterns and core-boxes is checked before they are finally handed over to the molding and core-making shops for mass production purposes. If necessary, corrections and rectification of patterns and core boxes are made at this stage after comparing the patterns, core boxes and sample castings.

Mold Boxes or Flasks

Selection of pairs of mold boxes (flasks) is another important factor as slight defects in flasks may cause cross-joints and ultimately the whole lot of castings may be rejected due to this. The difficulty of getting supplies of mold boxes in India is very great. Every Indian foundry has to make its own mold boxes. These boxes are made either from cast steel or fabricated, machined and drilled in their own shops. There is not a single firm in India specializing in the manufacture of mold boxes. In extreme cases it is found economical to import them from Europe. There is no general standardization of mold boxes and as such every foundry has different types of mold boxes for the same type of job.

Molding

Previously all the steel foundries of India were accustomed only to dry sand molding practice. Floor molding by hand ramming and board or plate molding in case of small castings by dry sand process including sweep molding in case of symmetrical circular castings and gear molding were known to Indian steel founders. Later on, simple types of hand ramming pattern draw type hand-operated molding machines were installed in our steel foundries.

At the present time, Indian foundries will be mechanized and in the meantime some foundries have completed mechanization and modernization for increased production from the same floor space with comparatively reduced number of workers. In these mechanized and semi-mechanized foundries, jolt squeeze turnover pattern draw type pneumatic molding machines, sand slingers and plain bumping type pneumatic molding machines have already been installed. In addition pneumatic rammers are widely used.

Previously, whatever may be the shape and size of the job, the molder had to rely on his skill for making proper molds in two or three parts, according to the need of the job, by hand ramming process. Only recently, machine molding has become popular and many foundries are installing latest type molding machines. Demands for mass produced items in large quantities have justified introduction of machine molding.

So far, only in one foundry, green sand molding has become successfully introduced. This foundry is putting up complete mechanization units consisting of roller conveyors, knock-out grids, underground belt conveyors for knocked-out sand, magnetic separators, elevators, rotary screen, sand storage hopper, bucket loader, overhead sand conveyors, feed hoppers with gate valves for feeding sands to the various types of pneumatic molding machines for green sand molding practice.

The selection of method of molding to be employed for making a particular casting depends on the size, shape, weight and intricacy of the casting and the total number of castings to be made. Indian foundries are gradually realizing the value of scientific methoding. In some foundries this is done on the blueprints but in majority of cases the selection of method of molding, gating and feeding are left to the discretion of the foreman in charge of molding shops. It is hoped that in course of time, every foundry will realize the value of scientific methoding for the standardization of these processes and introduce this methoding arrangement in their own foundries.

Core-Making

Formerly, cores used to be made by hand in Indian steel foundries but presently, some of our foundries have installed core-blowing machines for standard jobs, though the old method of manual core making is still practiced in all of our foundries. For comparatively larger cores, the use of core irons and bent or circular rod insertion to give better mechanical

support to the cores, provision of vent holes and cinder packing inside the cores are also done to obtain better permeability of the cores to produce sound castings. Core making is sometimes practiced according to the demand of the job, for example, striking type cores for cylindrical castings in almost all such cases the core is built upon a perforated iron pipe with straw rope winding over which a coating of loam sand mixture is applied and finally finished by

applying facing sand.

The mystery of core blowing or sand blowing is gradually becoming clear to the Indian founders. The principle of blowing consists of applying a stream of air at a pressure varying from 70 to 90 lb through the sand reservoir. The air carries a stream of sand suspended in it at high velocity through the holes in the blow plate into the cavity of the core mold. While the air stream continues on to the atmosphere through the vents in the box arranged so that air is allowed to pass out of the box, sand particles will be trapped and rammed automatically by abruptly arresting the travel of the sand grains. The practical knowledge of sand blowing is confined to the users of these machines in our country, and though the principle of blowing is very simple it is not a popular practice due to lack of knowledge on the part of foundry engineers who are more inclined to the old and orthodox methods and not sincerely interested to the modern developments in these lines.

Mold and Core Drying

Bogie hearth-type coal or coke-fired furnaces are generally used for mold and core drying in Indian foundries. Some foundries also have muffle type furnaces for drying of small cores. The temperature of these furnaces usually varies from 350 F to 400 F. For skin drying of floor molds and the molds which cannot be dried in the bogie hearth furnaces due to their size and shape, our foundries have to depend on sawdust and coal firing and only recently portable oil or coke-fired mold driers are used for drying of these types of molds. The coke-fired driers are provided with blowers and the oil-fired ones are provided with burner arrangements. Only recently, some of the foundries have installed coke-fired vertical core ovens with temperature recorders, etc. for drying of small and medium-sized cores. But for drying of molds and bigger cores, the bogie hearth type ovens are still extensively used.

Assembling and Closing of Molds

After the molds are dried, they are brought to the closing section for proper core setting, checking the correctness of placing of chaplets and cores, the molds are closed on trial before final closing.

Casting and Teeming

During assembling and closing the molds are placed on the floor in such order that these can be cast from the teeming ladle without back or forward movement of the rope-carriage of the casting crane. The casting temperature of molten steel varies according to the sections and dimensions of the job and in certain important cases the teeming temperature is recorded by immersion type pyrometers.

Stripping, Annealing and Fettling

After the castings are solidified and cooled down to some extent castings are stripped from mold boxes and the majority of adhering sands are removed by hammering either by ordinary or pneumatic hammers. Then the heads, runners, risers and extra metal are removed by flame cutting, gouging and sawing. After these are removed the castings are loaded on a truck and sent into the annealing furnace. In our country, both coal and oil-fired annealing furnaces are used. These annealing furnaces are generally provided with temperature recorders for the guidance of the heat treatment supervisors. Also the same type of furnaces are used for the heat treatment of austenitic Mn-steel castings. After annealing the adhering sand particles, scale, etc. are removed by shot-blasting.

After cleaning the castings by shot-blasting they are then brought to the dimensions as specified in the drawing by grinding over grinding wheels if possible or by swing grinder or portable pneumatic grinders then chipping. In India, none of the foundries have methodical arrangements for fettling (cleaning) of the castings. Proper fettling booths are not to be found in any foundry. There are, of course, swing grinders, floor grinders, chippers, welders, etc. which are distinct processes of fettling operations which the castings are subjected to but handling of castings from one operation to the other is tremendously heavy. It is hoped that sooner or later Indian foundries will gradually appreciate the value of having proper fettling booths.

Inspection

Apart from inspection at various stages in different shops, every foundry has its own Works Inspection Department to check quality from the point of knocking out, after fettling and before shipping. Besides these, inspection from Indian Stores Department is very, very severe. No casting can be welded, either major or minor, unless the inspectors have permitted such welding. Our customers are very rigid and even a little surface defect will bring forth a frown and minor welding is looked upon with suspicion. In our under-developed foundries such a strict customer is looked upon as a future investment towards greater production and better quality.

Market Conditions

India with her extensive railway communication system covering nearly 44,602 miles of track and with 200,645 wagons, 12,813 carriages, 7,732 locomotives and with a new building program of 6,000 wagons, 400 carriages and 220 locomotives, the Indian Railways are the largest consumer of steel castings in the country. Besides the large numbers of cement factories consume several thousand tons of Mn-steel castings. Sugar mills, agricultural tractor manufacturers, ship building yard and other engineering industries also require steel castings regularly for their maintenance.

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The productive capacity of Indian foundries has already been given before, which is not adequate to meet the country's demand and the major portion of the requirements are still imported from European countries. There is therefore, greater scope for not only increasing the capacities of existing foundries but establishing new units in the country not only to meet her ever growing internal demands but also to export to her neighboring countries like Burmah, Siam and Indonesia, etc.

Conclusion

I would like to say a few words about Indian foundry workers. As different components of different machines have to undergo different manufacturing processes and treatment, so also human labor in modern foundries has to undergo different processing or training for different jobs. Labor is very cheap in India, but it is very difficult to obtain skilled help. This difficulty is however solved by training Indian boys in different foundry trades such as molding, coremaking, pattern making, etc. by individual foundries. To obtain increased production introduction of some suitable incentives for betterment of the standard of living of the workers is necessary. Also to maintain good relations between employer and employee and in the interest of both parties it is necessary to make the workers conscious of the national need for our future development by properly educating them. At least a few foundries have better co-operation and increased production from their workers by following the above principles.

In conclusion, I must admit that the condition of

the Indian steel foundries are far from satisfactory. The layouts of most of the existing foundries are ill designed. As a result, even the scope of re-orientations of these are restricted from the point of view of increased production from the same floor space by mechanizations and semi-mechanization, improving the working conditions, minimizing handling of the materials. These difficulties, however can be overcome by planning some new model foundries which our country needs along with the expansion and modernization of the existing foundries. Even by remodeling the old foundries with correct plan of layouts and introduction of latest plants and machinery as far as possible we can increase the productive efficiency appreciably.

Now after the attainment of freedom of our Country, our National Government is doing its utmost for rapid industrial development of the country. American technical aid, both by investment of capital in our country and by the services of American experts, is playing a very important part in different fields of industrial development of our country, such as construction of river valley projects, developing a big factory for production of ferromanganese, construction of oil refineries and in every other field of industry including rural development.

Recently, the services of an American foundry expert has also been obtained by the Government. We hope, after careful study of the position of the Indian foundries both commercial and Government-sponsored, this American foundry expert will do much for the improvement and expansion of this particular branch of industry in our country.

PRACTICAL CONSEQUENCES OF SPACE, TIME AND TEMPERATURE RELATIONS DURING CASTING OF METALS

By

J. S. Abcouwer*

Introduction

The object of this paper is to focus the attention of foundrymen upon relations between the variables and more or less constant physical quantities decisive "in groups as new independent variables" for the temperature distribution in the cooling metal body and therefore for the production of sound castings.

This paper would have a logical start in a demonstration of how such a relationship can be deduced, but this would divert interest entirely from the object stated.

The equations, being new, cannot be found in literature, and something has to be told in their favor. For example, before the interface temperature in equation 3 begins to fall, the cosine function has become very flat and Fourier's Theorem (the differential equation of heat flow) is nearly fulfilled and

even better with higher values of
$$\frac{a_1t}{X^2}$$
.

The boundary condition used is the heat flow $W = \frac{1}{\sqrt{\pi^t}}$. B. θ_{ot} for a surface heat transfer coëfficient $a = \infty$ changing for a = a into:

$$W = \frac{1}{\sqrt{\pi^{\mathsf{t}}}} \cdot B. \; \theta_{ot} - \frac{1}{1 + \sqrt{\frac{B^2}{\pi a^2 \cdot \mathbf{t}}}}.$$

This new boundary condition is the backbone of this work and can be deduced by an indirect demonstration. So there is enough sound proof to warrant the introduction of these equations. Using them to check the experimental findings of thermal analyses will convince the skeptic, utilizing Tables 2 and 3.

The Equations

The equation for unidimensional heat flow reads:

$$(1) \quad \sqrt{z} \quad \sqrt{e} \quad \frac{\frac{L}{\sqrt{m}} \sum_{T_1, C_1} \left(\sqrt{c} - \sqrt{\frac{D^T}{m e^{-L}}} \right)}{\sqrt{c}} \cos \left[\frac{z}{X} \sqrt{\frac{\frac{N}{2} \sqrt{T_{C_1}}}{\sqrt{c}}} \right]$$

$$(2) \quad \sqrt{z} \quad \sqrt{e} \quad \frac{\frac{L}{\sqrt{m}} \sum_{t_1 + L_2} \sqrt{\frac{a_1 C}{X^L}} \left(1 - \sqrt{\frac{D^T}{m e^{-L}}} \right)}{\sqrt{c}} \cos \left[\frac{z}{X} \sqrt{\frac{\frac{N}{2} \sqrt{T_{C_1}}}{\sqrt{c}}} \right]$$

$$(3) \quad \sqrt{z} \quad \sqrt{e} \quad \frac{\frac{L}{\sqrt{m}} \sum_{t_1 + L_2} \sqrt{\frac{a_1 C}{X^L}} \left(1 - \sqrt{\frac{D^T}{m e^{-L}}} \right)}{\sqrt{c}} \cos \left[\frac{z}{X} \sqrt{\frac{N}{2} \sqrt{\frac{N}{m}} \sqrt{\frac{N}{m}}} \right]$$

$$(3) \quad \sqrt{z} \quad \sqrt{e} \quad \frac{1}{m} \frac{b_A}{b \cdot b_A} \sqrt{\frac{4\pi a_1 C}{X^L}} \cos \left[\sqrt{\frac{2}{b_A} \cdot b_A} \cdot \frac{z}{X} \sqrt{\frac{N}{4\pi a_1 C}} \right]$$

To use unidimensional heat flow equations for the

TABLE 1-TABLE OF THERMAL PROPERTY UNITS

Symbol	Units	Factor
t	h	Time
γ	kg/m³	Specific weight
C	cal/kg.°C	Specific heat
λ	cal./m.°C.h	Conductivity
θ_{\circ}	°C	Pouring temperature (above room temperature)
Θ	°C	Liquidus temperature (above room temperature)
$H = \frac{\theta_0 - \Theta}{\Omega}$		Superheat
$a = \frac{\Theta}{\lambda}$	m²/h	Temperature Diffusivity
γ.c	/	Temperature Dinustrity
$b = \sqrt{\gamma.c.\lambda}$	cal./mª.°C.h%	Heat Diffusivity
100 β		% liquid and solidification contrac- tion of the casting
Ve, Vr	m^3	Volume of casting or riser
Sc, Sr	m ²	Surface of casting or riser
100 €		% liquid in riser on the moment casting is solid
q	cal/kg	Heat of fusion
b_1b_2		TT - 1:00 - 1:10 - 1:00
B =	_cal/m ⁸ .°C.h %	Heat diffusivity of system 1-2
b_1+b		
A.	cal/m ³ .h	Heat flow
2	cal/m ² .°C.h.	Surface heat transfer coefficient
Pot	°C	Temperature at center at moment t

Foundry Manager, Messrs. Werkspoor N.V., Amsterdam, Holland.

This is an official exchange paper from the Netherlands Foundrymen's Association, Amsterdam, Holland to the American Foundrymen's Society for presentation during the 1952 A.F.S. International Foundry Congress. In the absence of the author this paper was presented by Wm. S. Pellini, Naval Research Laboratory, Washington, D. C.

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TABLE 2—THERMAL PROPERTIES OF METALS

Metal	Temp. Range Near Solidification	Spec. Weight x Spec. Heat, cal/m ³ $\gamma_1 x$ c_1	Thermal Conductivity, cal/m°Ch	Temperature Diffusivity, m^2/h $a_1 =$ $\lambda_1 c_1$	Heat Susceptibility, $\begin{array}{c} cal/m^{s\circ}Ch^{\frac{1}{16}} \\ b_1 = \sqrt{\gamma_1.c_1.\lambda_1} \end{array}$	Heat of Fusion cal/kg
Steel	Liquid	7200×0.20	5			65
Steel	Solid	7500x0.165	25	0.02	176	00
Cast	Liquid	6900x0.20	3			47 C 4.20
Iron	Solid	7100x0.165	25	0.0214	170	54.5 C 2.50

cooling of metals in molds, even for the most simple forms seems to be an unusual proposition but it is not. As soon as the cosine function is flat enough to

write S.W.dt= $\gamma_1 c_1 V.d\vartheta_{ot}$, $\frac{V}{S}$ takes the place of X

in our equations, that can be used for bodies as well.

This has to be done carefully. Whole surface S must represent a "sufficient" mold and therefore we see that even during solidification parts of S will no longer be "sufficient" but more or less saturated with heat (internal thin cores, re-entrant angles).

- tends to become a maximum and isother-

mal surfaces tend to become spheres.

Equation One

In this equation the term $\frac{B^2}{\pi a^2}$ (having the dimen-

sion of time) is the direct result of the temperature difference at surface. There are two distinctly different temperatures at the interface, one for sand side and the other for the metal side. To understand this phenomenon, producing a temperature discontinuity and air gap time, the reason is evident, metals being non-wetting against sand molds as well as in chills.

If during the start of freezing a diminishes by spontaneous widening of an air gap (e.g. casting against

chills) the air gap time $\frac{}{\pi a^2}$ becomes greater, the

cooling slackens, the temperature curve flattens and the interface temperature (metal side) rises accord-

Moreover the term $\frac{}{ma^2}$ tells us that only systems with the same value for this time can produce the same temperature distribution.

Equation Two

 B^2 If molding material and metals are the same

is constant and this equation involves only the influence of:

$$\frac{1}{\sqrt{\pi}} \frac{b_2}{b_1 + b_2} \sqrt{\frac{a_1 t}{X_2}} \text{ and of } \frac{aX}{\lambda_1} \sqrt{\frac{B^2}{\pi a^2}} = \frac{1}{\sqrt{\pi}}$$

$$\frac{b_2}{b_1 + b_2} \sqrt{\frac{a_1 t}{X^2}} \text{ or only of } \frac{a_1 t}{X^2}.$$

Therefore we can treat $\frac{a_1t}{v_2}$ as the only independ-

ent variable:

- (1) during the casting of metals in molds the dimensionless quantity $\frac{a_1t}{X^2}$ tells everything about
- the temperature distribution in the metal body. (2) possessing a successful solution for a casting, an equally successful solution for a similar casting

TABLE 3—THERMAL PROPERTIES OF MOLDS

Material	Temp. Range, Room to °C	Spec. Heat, c₂ cal/kg	Spec. Weight, γ ₂ kg/m ⁸	Thermal Conductivity, \(\lambda_2\) cal/m°Ch	Temperature Diffusivity, $\begin{array}{c} \lambda_{3} \\ a_{2} =$	Heat Diffusivity $b_2 = \sqrt{\gamma_2.c_2.\lambda_9}$ $cal/m^{s_0}Ch^{\frac{s_0}{6}}$
Silica sand	1,000	0.27	1,600	0.40	0.00093	13
dry 50–70 Linoil core sand baked 50–70	1,000	0.27	1,500	0.40	0.001	12.6
Synthetic mold sand, dry 50-70	1,000	0.27	1,600	1.0	0.0023	20.7
Cement sand	1,000	0.27	1,500	0.60	0.0015	15.4
Synthetic mold sand, dry	1,000	0.27	1,600	0.60	0.0014	16

is at hand by calculating all time quantities keeping for this quantities in both castings $\frac{a_1t}{X^2}$ at the same level.

Equation Three

This equation follows from Equation 1 and 2 by (other factors being constant) making the air gap time zero, or making a equal to infinity, neglecting the temperature difference at surface.

This equation can be used conveniently to calcu-

late $\frac{a_1t}{X^2}$ for the moment the interface (metal side

and sand side) reaches solidification temperature ⊕ as follows:

$$\theta = \sum_{x \in X} = \sqrt{e} \quad \frac{1}{\pi} \frac{b_{x}}{b_{x} \cdot b_{x}} \sqrt{\frac{\sqrt{\pi d} \cdot 1}{X^{T}}} cos \left[\sqrt{\frac{2 b_{x}}{b_{x} \cdot b_{x}}} \sqrt{\frac{X^{T}}{\sqrt{\pi d} \cdot 1}} \right]$$

$$\frac{\partial}{\partial J_{x}} = e^{-\frac{1}{\pi} \frac{b_{x}}{b_{x} \cdot b_{x}} \sqrt{\frac{\pi d}{X^{T}}}} \left(1 - \frac{b_{x}}{b_{x} \cdot b_{x}} \sqrt{\frac{X^{T}}{\sqrt{\pi d} \cdot 1}} \right)$$

$$\frac{\partial}{\partial J_{x}} = e^{-\frac{1}{\pi} \frac{b_{x}}{b_{x} \cdot b_{x}} \sqrt{\frac{4\pi d}{X^{T}}}} \left(\frac{b_{x} \cdot b_{x}}{b_{x}} - \frac{b_{x}}{b_{x}} \sqrt{\frac{X^{T}}{\sqrt{\pi d} \cdot 1}} \right)$$

$$\left[1 - \frac{b_{x}}{b_{x}} \left(\sqrt{\frac{X^{T}}{\sqrt{\pi d} \cdot 1}} - 1 \right) \right]$$

$$\ln \frac{\Theta}{\theta_{2}} = \ln \frac{b_{1} + b_{2}}{b_{1}} \cdot \frac{\Theta}{\theta_{0}} = \frac{b_{1} + b_{2}}{b_{1}} \cdot \frac{1}{H + 1} - 1$$

$$\frac{b_{1} + b_{2}}{b_{1}} \cdot \frac{1}{H + 1} - 1 = -\frac{1}{\pi} \frac{b_{2}}{b_{1} + b_{2}} \sqrt{\frac{4\pi a_{1}t}{X^{2}}}$$

$$-\frac{b_{2}}{b_{1}} \sqrt{\frac{X^{T}}{4\pi a_{1}t}} + \frac{b_{2}}{b_{1}}$$

$$-\frac{b_{2}}{d_{1}} \sqrt{\frac{X^{T}}{4\pi a_{1}t}} + \frac{b_{2}}{b_{1}}$$

$$\frac{4\pi a_{1}t}{X^{2}} - \frac{(b_{1} + b_{2})^{2}}{b_{1} \cdot b_{2}} \cdot \frac{H}{H + 1} \sqrt{\frac{4\pi a_{1}t}{X^{2}}} + \pi \frac{b_{1} + b_{2}}{b_{1}} = 0$$

$$With: \frac{4\pi a_{1}t}{X^{2}} = t^{2}$$

$$\pi \frac{b_{1} + b_{2}}{b_{1}} = p$$

$$\frac{b_{1} + b_{2}}{b_{2}} \cdot \frac{H}{H + 1} = \sigma$$

$$\left(\frac{t}{D} \right)^{2} - \sigma \cdot \frac{t}{D} + \frac{1}{D} = 0$$

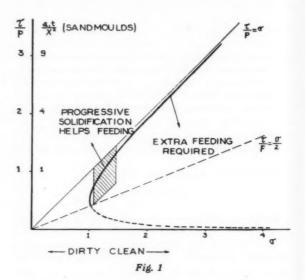
Metals in Sand Molds

The hyperbola, Fig. 1, being the boundary between the entirely fluid state and the start of solidification at the interface with pouring time zero, we can use this same figure to derive certain conclusions about the pouring time of castings:

(1) for the superheat $\frac{b_1+b_2}{b_2} \cdot \frac{H}{H+1} = 1$ is the

lowest limit (or H = 0.1).

(2) to arrive at progressive solidification in a top poured sand mold without denseners or chills



and with H = 12.5 per cent the value of $\frac{a_1t}{X^2}>1.$

(3) for clean sound castings top pouring with superheat and $\frac{a_1t}{X^2}$ the condition of the cross-hatched

- (4) here we have a formula for the pouring time of castings depending only on superheat and wall thickness. To use it for cast iron a₁ = 7mm²/sec, X = half wall thickness in mm and t is expressed in sec.
- (5) using more superheat gives points below the hyperbola and as a consequence more feeding is required.
- (6) using superheat at the low side and a quick pour is not a reasonable answer (dirty castings).
- (7) superheat and pouring time cooperate in producing superior conditions.

Solidification

(a) From Equation 3 we derive for the moment superheat has disappeared (here using $\frac{V}{S}$ in the

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place of X): $\Theta = \theta_0 e \qquad -\frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{\frac{4\pi a_1 t_s}{V^2}} \frac{S^2}{V^2}$ $\ln \frac{\Theta}{\theta_0} = -\frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{\frac{S^2}{4\pi a_1 t_s}} \frac{S^2}{V^2} = -\frac{H}{H + H}$

(b) During solidification we have:

$$S \int Wdt = \gamma_1 c_1 \frac{q_1}{c_1} dV$$

$$S \int \frac{1}{\sqrt{\pi t}} B. \otimes dt = \gamma_1 c_1 \frac{q_1}{c_1} dV$$

$$S \left[\frac{1}{\pi} \cdot \frac{b_2}{b_1 + b_2} \sqrt{4\pi a_1 t} \right]_{t_n}^{t_c} = \frac{q_1}{c_1 \otimes} (1 - \epsilon) V$$

$$\epsilon = \text{fluid part}$$

$$\frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{4\pi a_1 t_e \frac{S^2}{V^2}} = \frac{q_1}{c_1 \Theta} (1 - \epsilon) + \frac{H}{H + 1}$$

$$\ln \frac{\Theta}{\theta_{\circ}} = -\frac{1}{\pi} \frac{\mathbf{b_2}}{\mathbf{b_1 + b_2}} \sqrt{\frac{\mathbf{a_2 c_S^2}}{\mathbf{V^2}}} + \frac{\mathbf{q_1}}{c_1 \Theta} (1 - \epsilon)$$

(c) After solidification we have:

$$\theta = \theta_0 e - \frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{\frac{S^2}{4\pi a_1 t}} + \frac{q_1}{C_1 \Theta}$$

$$\cos \left[\sqrt{\frac{2b_2}{b_1 + b_2}} \cdot \frac{V_x}{V} \sqrt[4]{\frac{V^2}{4\pi a_1 t S^2}} \right]$$

(d) The end of solidification is reached for:

$$\frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{\frac{4\pi a_1 t_c}{X^2}} = \frac{q_1}{c_1 \Theta} + \frac{H}{H + 1} \approx 0.35$$
(c.i. in sand mole $H \approx 0.11$

(e) For the proportion pouring time and solidification time we get:

$$\sqrt{\frac{t_{p}}{t_{c}}} = \frac{\frac{H}{H+1}}{\frac{q_{1}}{C_{1}\Theta} + \frac{H}{H+1}} \qquad \text{Here } \frac{0.1}{0.35} ; \frac{t_{p}}{t_{c}} = 0.08$$

(f) By augmenting the superheat this proportion can be bettered. A slow hot pour promotes self feeding.

Dimensioning Risers and Heads

Making use of this temperature, time and space relation we can deduce an equation for this dimensioning as has been shown at the International Congress at Brussels in 1951¹ by the equation:

$$\frac{V_n}{S_n}: \frac{V_r}{S_r} = 1 - A \frac{V_e}{V_r}$$
See Figs. 2 and 3

This calculation, slightly changed, is repeated here to complete this paper. Writing the equation for the end of solidification for the thickest part of the casting where a head is to be put on, we have:

$$\frac{1}{\pi} \frac{b_2}{b^1 + b_2} \sqrt{4\pi a_1 t} \frac{S_n^2}{V_n^2} = \frac{q_1}{c_1 \Theta} + \frac{H}{H+1}$$

and for the partly solidified head:

$$\frac{1}{\pi} \frac{b_2}{b_1 + b_2} \sqrt{4\pi a_1 t \frac{S_r^2}{V_{\epsilon^2}}} = \frac{q_1}{c_1 \Theta} (1 - \epsilon) + \frac{H}{H + 1}$$

Dividing both equations:

$$\frac{V_n}{S_n}: \frac{V_r}{S_r} = \frac{\frac{q_1}{c_1\Theta} (1-\varepsilon) + \frac{H}{H+1}}{\frac{q_1}{c_1\Theta} + \frac{H}{H+1}} = 1 - \frac{\varepsilon}{1 + \frac{c_1\Theta}{q_1} \frac{H}{H+1}}$$

 ϵ being the fluid part of the riser (volume V_r) 100 β being percentage of fluid and solidification contraction of the casting we have $\epsilon V_r = \beta V_c$ or

$$\epsilon = \beta \frac{V_e}{V_r}$$

$$\frac{H}{H+1} \approx 0.35$$
(c.i. in sand molds
$$H \approx 0.11$$
)
$$\frac{V_n}{S_n} : \frac{V_r}{S_r} = 1 - \frac{\beta \frac{V_e}{V_r}}{1 + \frac{c_1 \Theta}{q_1} \frac{H}{H+1}} = 1 - A \frac{V_e}{V_r}.$$
See Ref. 2.

If the head or riser is insulated (heat susceptibility b₂¹) and the pouring time is not zero but the head is filled the pouring time later with metal of the same temperature of the mold we get:

$$\sqrt{1 - \frac{t_{\rm p}}{t_{\rm c}}} \cdot \frac{\frac{V_{\rm n}}{S_{\rm n}}}{\frac{V_{\rm r}}{S_{\rm r}}} = \frac{b_2}{b_2^{1}} (1 - A \frac{V_{\rm c}}{V_{\rm r}}).$$

Conclusions

- 1. A mathematical treatment, making use of a temperature discontinuity at the metal-mold interface, gives a temperature distribution very nicely in accord with recently published experimental results (thermal analyses of Fifield and Schaum, Schwartz and Bock, Dunphy and Pellini).
 - 2. The dimensionless time $\frac{a_1t}{X^2}$ can be used by

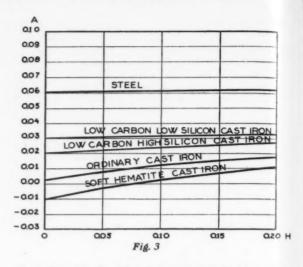
foundrymen to calculate pouring time, ingates, risers, heads for a casting starting from a successful geometrically similar one.

3. Using the value for $\frac{a_1t}{X^2}$ for the start of freez-

ing at the interface for a zero pouring time, this value is a minimum for the pouring time in the case of well distributed top pouring and a maximum in the case of bottom pouring of vertical slender pieces, but mostly can be used with minor changes for both.

4. End of freezing is found as

$$\sqrt{4_{\pi}a_{1}t\frac{S^{2}}{V^{2}}}=\pi\frac{b_{1}{+}b_{2}}{b_{2}}\,\left(\frac{q_{1}}{c_{1}\Theta}+\frac{H}{H+1}\right)$$



5. A formula for the dimensioning of risers and heads follows from this theory.

References

 J. S. Abcouwer, "Dimensioning Risers and Feeding Heads," Transactions of Congrés International de Fonderie, Bruxelles, pp. 1-11 (1951). See Appendix below.
 J. B. Caine, "Risering Castings," Transactions, American

Foundrymen's Society, vol. 57, pp. 66 (1949).

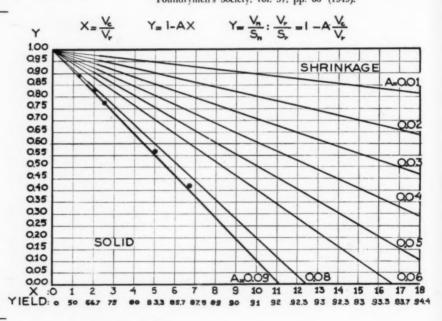


Fig. 2

APPENDIX

DIMENSIONING RISERS AND FEEDING HEADS

If the gating system in the production of sound castings in any metal is inadequate to pour and feed the casting at the same time, risers or feeding heads are introduced. Hereby the casting yield is influenced and becomes less in most instances.

It is good foundry practice to design the pouring system as long as possible for both purposes, as attained successfully in many newer solutions (Connor block, Winte and Barlow whirlgate). Two conditions have always to be fulfilled: the pouring time has to be correct and the feeding has to be sufficient. By this feeding through an ingate or pouring through a riser, there are two advantages: firstly a gain in time, because the casting is already losing heat during pouring and in the same time the metal in the feeding system is replenished and cooling is postponed; and

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secondly a gain in temperature, because the mold walls of the feeding system are heated up by metal filling the mold.

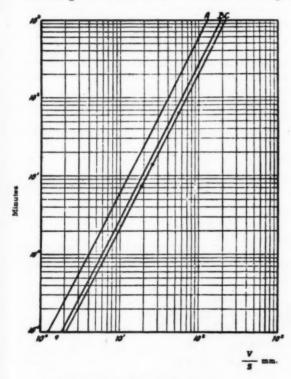
In practice the mold walls are so dimensioned that during casting and solidification no heat from the mold is lost to the atmosphere. In thermal respect the mold walls are therefore infinite. This circumstance makes it possible to treat the subject mathematically. As soon as heat is lost to the atmosphere, the thermal processes are accelerated. This can be expected especially in the case of bigger molds.

Efficient feeding can be promoted by using the metal in the riser to full advantage by correct dimensioning, insulating or exothermal compounds for risers, or chills and denseners in the mold and other means.

We shall treat the simple case that there is a "sufficient" mold (infinite in thermal respect), the whole mold filled with metal of the same temperature (the pouring time is zero) the metal has no freezing range, there are no convection currents in the metal and there is no other heat-flow in the casting or in the head, then perpendicular to the mold walls.

All these assumptions do not reduce our case to a strictly hypothetical one, without any practical value; on the contrary, the solutions make clear, which measures can be taken to improve a situation.

Take e. g. the Williams core: with metal having a



A: Cast Iron with 20 % superheating

Fig. 4—Chworinov's graph and H.M.F.A. end of freeze times.

freezing range, the solidifying metal passes through a mushy state and can be forced into voids that will form inside the casting even at a higher level and therefore for this case is imperative, but the atmospheric pressure improves also the situation if the mushy feeding metal has to flow down.

In expressing the pouring time as a fraction of the solidification time, we find for steel, according to Chworinov:

 $t_{\rm p}=(0.01-0.03)~t_{\rm c}$ { $t_{\rm p}=$ pouring time $t_{\rm c}=$ solidification time This formula can only be used for steel in connection with its small superheat. For cast iron with 10 per cent superheat this formula can be modified into:

 $t_p = (0.03 - 0.20) t_0$

and with 20 per cent superheat into:

 $t_p = 0.20 \ t_c$. In this latter case we have to resort to top pouring and promote by this slow pouring directional solidification. The requirement of feeding metal for the casting will be less than with pouring time zero: as the risers and head are filled at the end of pouring with the hottest metal, the solidification begins late and the largest percentage of feeding metal is produced.

This time lag t_p , known as a fraction of the solidification time t_c of the casting, as well as the use of isolating material for riser or head, can be taken into account in introducing this time lag and different heat susceptibilities b_2 and b'_2 , in the fundamental equation.

A Controversy

In the published literature on heat transfer problems is a controversy to the point whether Chworinov's claims for the solidification times of castings in sand molds are too broad or not.

This controversy has been given rise by the fact that, comparing the results of bleeding tests with the curves found by the Heat and Mass Flow Analyser Laboratory, a good agreement is found, taking the start of freeze curves (see the Reports of the Heat transfer Committee of the A.F.S.).

On comparing the H.M.F.A. end of freeze curves with Chworinov's graph (Fig. 4), there proves to be a complete agreement. Taking e. g. Dr. Paschkis' discussion of the paper: "Dimensioning Risers" by J. B. Caine (Transactions A.F.S., vol. 56, page 500), we read:—

"A sphere, 41/2 in. diameter, has a value of volume

= = R = 0.75. Its solidification time dependsurface

ing on the temperature is 270-330 seconds, against 540 seconds according to Chworinov."

Now 270-330 seconds is the start of freeze, but according to Dr. Paschkis' own curves, the end of freeze time is 500-600 seconds.

A slab, $4\frac{1}{2}$ in. thick and large enough to have no end effects, has a value of R=4.5. Its solidification time is 695-1105 seconds whereas Chworinov gives 324 minutes.

Here Dr. Paschkis makes a mistake: the value of

B: Cast iron with 10 % superheating

C: Cast steel

Values H. M. P. A
 Dr Paschkis

$$R = \frac{\text{volume}}{\text{surface}}$$
 of a slab without end effects is half the thickness
$$= \frac{41/2}{9} = 2.25.$$

The end of the freeze time found with this value in the H.M.F.A. graph is 4200 seconds, the same value as found by Chworinov.

The phenomenon of a difference in start and end of freeze is only found in metals with a freezing range. If this freezing range becomes narrower, the start of freeze time will draw nearer to end of freeze time and in the end they will coincide.

The experimental results of others e. g. of R. W. Ruddle (British Non-Ferrous Metals Research Association) are also in agreement with Chworinov.

The Solidification Equation

In order to have a start, it is necessary to make certain assumptions:—

(a) The contact temperature between metal and mold wall equals the solidification temperature θ from the beginning till the end of solidification.

(b) The heat flow through the mold wall first dissipates the superheat before solidification starts, so

$$\frac{2}{\sqrt{\pi}}b_2\theta\sqrt{t_a} = \frac{V}{S} \times c_1(9_{\bullet} - \theta)$$

or:
$$\sqrt{\frac{4 \pi a_1 t_a S^2}{V^2}} = \pi \frac{b_1}{b_2} \frac{3_0 - \theta}{\theta} = \pi \frac{b_1}{b_2} H$$

For the solidification we get:-

$$S_{c_{s}}^{t} \frac{1}{\sqrt{\pi t}} b_{3} \theta d t = + \int_{Y_{1}} q_{1} d V$$
or:
$$\left[\frac{2}{\sqrt{\pi}} b_{2} \theta \sqrt{t} \right]_{s}^{t} = + Y_{1} c_{1} \frac{q_{1}}{c_{1}} \int_{S}^{d V} \frac{V}{S}$$

$$\left[\frac{2}{\sqrt{\pi}} b_{2} \theta \sqrt{t} - \frac{2}{\sqrt{\pi}} b_{2} \theta \sqrt{t_{4}} \right]$$

$$= Y_{1} c_{1} \frac{q_{1}}{c_{1}} \frac{V}{S} \left[\frac{V - c V}{V} \right]$$

$$\frac{2}{\sqrt{\pi}} b_{2} \theta \sqrt{t} = Y_{1} c_{1} \frac{q_{1}}{c_{1}} \frac{V}{S} \left[\frac{V - c V}{V} \right]$$

$$+ \frac{V}{S} Y_{1} c_{1} (\mathfrak{F}_{s} - \theta)$$

$$\sqrt{4 \pi a_{1} t \frac{S^{2}}{V^{2}}} = \pi \frac{b_{1}}{b_{2}} \left[\frac{q_{1}}{c_{1}} \frac{V - c V}{V} + H \right]$$

$$\sqrt{4 \pi a_{1} t \frac{S^{2}}{V^{2}}} = \pi \frac{b_{1}}{b_{2}} \left[\frac{q_{1}}{c_{1}} (1 - c) + H \right]$$

In this form the solidification equation has also validity for metals with a freezing range. For metals without a freezing range, the formula can easily be transformed in an equation for the thickness of the solidified layer. Such a layer does not form in metals with a long freezing range. Bleeding these metals is impossible.

Omitting the above mentioned assumptions, starting only from a "sufficient" sand mold and the superheat:—

$$9_0 - \theta > \frac{b_1 + b_2}{b_2} 9_0$$
 or : $\frac{b_1 + b_2}{b_2} \cdot \frac{H}{H + 1} > 1$

it is possible to demonstrate on a theoretical basis, that, taking the cooling of the metal until the wall temperature $= \theta$ is reached separately into consideration, this equation has to be changed into:—

$$\sqrt{4 \pi a_1 t \frac{S^2}{V^2}} = \pi \frac{b_1 + b_2}{b_2} \left[\frac{q_1}{c_1 \theta} \frac{b_1}{b_1 + b_2} (1 - \epsilon) + \frac{b_1 + b_2}{b_1} \frac{H}{H + 1} \right]$$

and with a freezing range $\theta_1 - \theta_2$ we can write down the same equation once for $\theta = \theta_1$ and $H = H_1$ and once for $\theta = \theta_2$ and $H = H_2$.

As soon as the mold walls are no longer sufficient (infinite in thermal respect) e. g. losing heat to the atmosphere, the cooling is accelerated and from that moment the equation changes.

Feeding and Risering

If a casting (volume V_c , surface S_c) with a riser or feeding head (volume V_r and surface S_r) is cast in a sufficient sand mold (walls infinite in thermal respect for the time of solidification) with a pouring time zero, the riser must be dimensioned to feed the casting.

In most castings the wall thicknesses are different and also $\frac{V}{S}$ differs from the mean value $\frac{V_{\sigma}}{S_{\sigma}}$.

So we have to seek for a given casting the maximum value or values $\frac{V_n}{S_n}$ and put the risers at these places (Fig. 5).

Writing the solidification equation for the riser and casting at the moment the last metal is setting we get:—

$$\sqrt{4 \pi a_1 t \frac{S_r^2}{V_r^2}} = \pi \frac{b_1 + b_2}{b_2}$$

$$\left[\frac{q_1}{c_1 \theta} \cdot \frac{b_1}{b_1 + b_2} (1 - \epsilon) + \frac{b_1 + b_2}{b_1} \cdot \frac{H}{H + 1} \right]$$

$$\sqrt{4 \pi a_1 t \frac{S_n^2}{V_n^2}} = \pi \frac{b_1 + b_2}{b_2}$$

$$\left[\frac{q_1}{c_1 \theta} \cdot \frac{b_1}{b_1 + b_2} + \frac{b_1 + b_2}{b_1} \cdot \frac{H}{H + 1} \right]$$

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 $\frac{V_0}{S_0}: \frac{V_c}{S_c} = 1 - A \frac{V_c}{V_c}$ $\frac{V_0}{S_0}: \frac{2R_c}{S_c} = 1 - A \frac{V_c/L}{4R_c}$ Fig. 5

(ε being the part of the riser metal not solidified in the riser but in the casting.)

$$\frac{\frac{V_{u}}{S_{u}}}{\frac{V_{r}}{S_{r}}} = \frac{\frac{q_{1}}{c_{1}\theta} \frac{b_{1}}{b_{1} + b_{2}} + \frac{b_{1} + b_{2}}{b_{1}} \cdot \frac{H}{H+1} - \frac{q_{1}}{c_{1}\theta} \frac{b_{1}}{b_{1} + b_{2}} \epsilon}{\frac{q_{1}}{c_{1}\theta} \frac{b_{1}}{b_{1} + b_{2}} + \frac{b_{1} + b_{2}}{b_{1}} \frac{H}{H+1}}$$

$$\frac{\frac{V_{u}}{S_{u}}}{\frac{S_{u}}{S_{r}}} = 1 - \frac{\epsilon}{1 + \frac{c_{1}\theta}{q_{1}} \left(\frac{b_{1} + b_{2}}{b_{1}}\right)^{\theta} \frac{H}{H+1}}$$

At the moment the casting has solidified, all the metal in the casting and riser is solid if the riser has minimum proportions.

If β is the percentage of fluid and solidification contraction of the metal, the amount of metal that during that time has been left in the fluid state in the

riser must be:
$$\beta V_c$$
 . So $\beta V_c = \epsilon V_r$ or $\epsilon = \beta \frac{V_c}{V_r}$

Substituting this value in the preceding formula, we get:—

$$\frac{\frac{V_{0}}{S_{n}}}{\frac{V_{r}}{S_{r}}} = 1 - \frac{\beta}{1 + \frac{c_{1}\theta}{q_{1}} \left(\frac{b_{1} + b_{2}}{b_{1}}\right)^{2} \frac{H}{H + 1}} \cdot \frac{V_{r}}{V_{r}}$$

By replacing the different terms as follows:-

$$\frac{\frac{V_o}{S_o}}{\frac{V_r}{S_r}} = Y \text{ or } = \frac{1}{x};$$

$$\frac{\frac{\beta}{1 + \frac{c_1 \theta}{q_1} \left(\frac{b_1 + b_2}{b_1}\right)^2 \cdot \frac{H}{H + 1}}}{\frac{V_o}{V_r}} = X \text{ or } = \frac{1}{y} : \text{ we can write:} - \frac{V_o}{V_r}$$

$$Y = 1 - AX \text{ or } -\frac{1}{x} = 1 - A - \text{ or } x = 1 + \frac{A}{y}$$
This equation represents the theoretically determined by the statement of t

This equation represents the theoretically determined boundary between shrinkage and solidity in the casting, when pouring time is zero and the casting and riser are in the same way surrounded by molding material.

In the official Exchange Paper of the A.F.S. to the French Foundrymen's Society (Paris 1948). J. B. Caine published an experimentally determined curve from Research Report 13 (1947) Steel Founders Society of America.

Instead of our curve: $x = 1 + \frac{A}{y - A}$, Caine deduced by logic reasoning a curve:

$$x = 1 + \frac{a}{y - b}$$

With the values a = b = A = 0.09 we get a perfect agreement between our curve and the value of the experimentally determined boundary (Fig. 6).

So Caine's hyperbola is reduced to a straight line (Fig. 2).

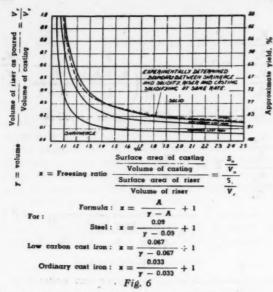
The Value of A

From published literature we have to find the value of

A =
$$\frac{\beta}{1 + \frac{c_1 \theta}{q_1} \left(\frac{b_1 + b_2}{b_1}\right)^2 - \frac{H}{H + 1}}$$
(Fig. 3, Tables 2, 3 and 4).

From: "Volume changes of Cast Iron," by Saeger & Ash (Transactions, A.F.A. 1932) and other sources we deduce for metal in sand molds values in Table 4.

From Fig. 6 we know that the boundary between shrinkage and solidity for steel is found with a value A = 0.09. Knowing that even with atmospheric riser* See Transactions, A.F.S., vol. 57, pp. 66-75 (1949).



ing it will be impossible to drain the last drop of, mostly, semi-fluid metal from a riser a safety factor of 50 per cent is acceptable, the more so, as for cast iron the solidification shrinkage or -expansion also depends upon the cooling rate.

TABLE 4

1 ABLE 4				
Material	Formula	A (theoretically)	A (in practice	
Steel	0.06 + 0.25 H	1		
Steel	$1 + 4.8 \frac{H}{H + 1}$	_ 0.06	0.09	
Low Carbon	0.03 + 0.15 H			
low Silican cast iron	$1+5\frac{H}{H+1}$	~ 0.04	0.06	
Low Carbon	0.02 + 0.15 H			
high Silicon cast iron	$1+5\frac{H}{H+1}$	~ 0.025	0.04	
Ordinary	0.003 + 0.15 H			
Ordinary cast iron	$1+5\frac{H}{H+1}$	~ 0.01	0.02	
	-0.01 + 0.15 H			
Soft Hematite cast iron	1+5-H+1	— 0.005	0.01	

Solving Our Equation

The equation $\frac{V_n}{S_n} : \frac{V_r}{S_r} = 1 - A \frac{V_o}{V_r}$ contains

the factors V, and S, as unknown quantities.

We have to construct our risers or heads in such a

way that ____ becomes a maximum.

To obtain this maximum, height and width of the riser must be the same. The easiest way to solution V_r is by expressing — and V_r both in R_r .

If a riser or feeding head touches a casting in one plane, the dimension of heat flow is 1.5 when this plane has the whole length of the casting and is 2.5 if this is not the case and width, height and length are equal.

Example: Calculate a feeding head for a place $2R \times H \times L$ dimension of head: $2^x \cdot 2^x \cdot L$ (see Fig. 7).

$$\frac{V_{u}}{S_{u}} : \frac{V_{r}}{S_{r}} = 1 - A \frac{V_{c}}{V_{r}}$$

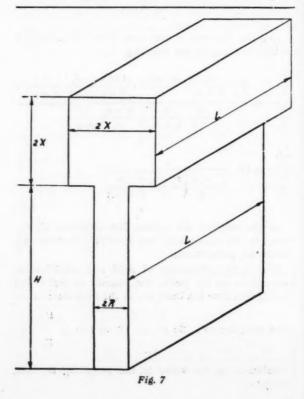
$$\frac{2RH}{2H} : \frac{4X^{2}}{6X} = 1 - A \frac{2RH}{4X^{2}}$$

$$X = \frac{3}{4}R \pm \sqrt{\frac{9}{16}R^{2} + \frac{A}{2}RH}$$

$$= \frac{3}{4}R + \frac{3}{4}R\sqrt{1 + \frac{8}{9}A\frac{H}{R}}$$

TABLE 5

$\frac{H}{R}$	X	V_r	$V_c \equiv 2 RH$	$Yie!d = \frac{V_e}{V_e + V}$
1,000	6.1 R	148 R2	$2,000 R^{z}$	93.5 %
500	4.55 R	$82 R^{z}$	$1,000~R^{z}$	93 %
100	2.6 R	27 Rs	200 R ²	88.5 %
50	2.15 R	19 R2	100 R ^g	84 %
10	1.7 R	12 R*	20 R ²	63 %
5	1.58 R	10 R2	10 Rs	50 %



By winding this plate around a vertical axis, we get a liner or cylinder.

Promoting Efficiency

As the cooling curve of a metal in a mold (b_2) and H constant) solely depends on the dimensionless a.t —, it is clear that linear wall thickness has noth- V^2

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ing to do with feeding requirements. We often find in literature a steeper temperature gradient predicted for small wall thicknesses and in consequence better feeding properties. With small wall thicknesses it becomes however soon impossible to feed a casting properly over a certain distance: the feeding metal cannot run down along the large surfaces in the available short time in narrow channels, normal feeding is impossible and tapering has to be used to prevent centerline porosity. In metals with easy feeding properties, for instance ordinary cast iron, saleable castings, even with larger uniform wall thicknesses, e. g. when the castings are flat pieces, are produced without risers. In the cast iron foundry the pouring system goes a long way to dispense with risers and heads, especially when the sprue, cup or pouring basin can be used as

Top pouring with a pouring time t_p and solidification time of the casting to changes our equation in-

$$\sqrt{1 - \frac{l_p}{l_c}} \cdot \frac{\frac{V_n}{S_n}}{\frac{V_r}{S_r}} = 1 - A \frac{V_e}{V_r}$$

If in connection herewith the riser or head is insulated (heat susceptibility of the insulating material b'2) the formula changes into:-

$$\sqrt{1 - \frac{t_p}{t_e}} \cdot \frac{\frac{V_*}{S_*}}{\frac{V_r}{S_*}} = \frac{b_2}{b_3'} \left(1 - A \frac{V_e}{V_r} \right)$$

Pouring through a riser (riser used as a pouring cup) makes the equation change into:-

$$\sqrt{1 - \frac{2 t_p}{t_c}} \cdot \frac{\frac{V_A}{S_n}}{\frac{V_r}{S_r}} = 1 - A \frac{V_c}{V_r}$$

When pouring through an insulated riser, which is also used as a pouring cup, the equation changes in-

$$\sqrt{1 - \frac{2 t_p}{t_e}} \cdot \frac{\frac{V_e}{S_e}}{\frac{V_r}{S_e}} = \frac{b_2}{b_3'} \left(1 - A \frac{V_c}{V_r} \right)$$

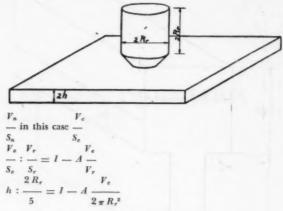


Fig. 8

Example: Calculate the feeding head for the same plate, dimensions 2 $R \times H \times L$, dimensions of the head 2X. 2X. L. The plate is poured at the top in a time t_p . The feeding head is surrounded by insulating material with a heat susceptibility b'2

$$\sqrt{1 - \frac{t_{o}}{t_{o}}} \cdot \frac{\frac{V_{o}}{S_{o}}}{\frac{S_{o}}{V_{r}}} = \frac{b_{2}}{b_{2}^{2}} \left(1 - A \frac{V_{c}}{V_{r}}\right)$$

$$\frac{t_{c}}{t_{p}} = 0.1 \; ; \qquad \frac{b_{2}}{b_{2}^{2}} = 1.25$$

$$0.76 \frac{\frac{V_{o}}{S_{o}}}{\frac{V_{r}}{S_{r}}} = 1 - A \frac{V_{c}}{V_{r}}$$

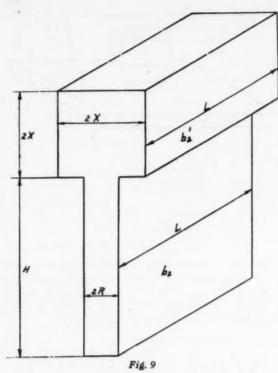
$$0.76 R \frac{3}{2 X} = 1 - A \frac{R H}{2 X_{2}^{2}}$$

$$X = 0.57 \text{ R} \pm 0.57 \text{ R} \sqrt{1 + \frac{3}{2} A \frac{H}{R}}$$

For a low carbon, low silicon cast iron —A = 0.09.

TABLE 6

Н	v	V - 4 Y2	Ve=2RH Yie	V.
R			ve = 2 KII III	$V_e + V$
1,000	6 R	144 R ^s	2,000 Rs	93.5 %
- 500	4.5 R	81 R ³	1,000 Rs	
100	2.4 R	23 R	200 Rs	93 %
50	2 R	16 R*	100 R ^a	86 %
10	1.4 R	8 Rs	20 R*	71 %
5	1.2 R	5.8 Rs	10 R*	63.5 %



The temperature difference in the direction of heat flow of a solidifying casting being very small, a condition aggravated by grain growth at random in the liquid (in the case of low carbon cast iron if still 50 per cent of the metal is fluid, pouring out ceases completely) conditions must be improved by a thermal gradient in the direction of feeding by every conceivable technical means.

If we put a big riser or head on a slender piece (e. g. cylinder or plate) of this cast iron, half of this solidification shrinkage escapes from feeding "per se" and more if the slenderness purpasses a certain value (for parallel walls 1:4 for cast steel) and a smaller head performs the same job, if calculated for the part it can feed maximally.

Incorporating half the solidification shrinkage in our calculation of course is impossible for the riser metal is in the same condition, but we see clearly that computing the riser dimensions is only the whole story for chunky castings and normally only part of the story.

In normal production, things are much more complicated and the question is not wholly answered by: the riser is big enough to feed this casting.

That in the cast iron field it is possible very often to get sound castings without heads or risers, proves the ingenuity of this old trade, but was helped in the past by the solidification expansion of charcoal pig and now with our low carbon mixtures probably by the undercooling associated with superheat.

Conclusions

1. For the same superheat, mold and metal, Chworinov's claims are in accord with the end of freeze times of the "Heat and Mass Flow Analyzer Laboratory."

2. Superheat is of paramount importance on start of freeze time, and in a lesser degree on end of freeze

times and feeding.

3. Risers and heads can be calculated mathematically somewhat simpler as with Caine's equation by:

$$\frac{V_n}{S_n} : \frac{V_r}{S_r} = 1 - A \frac{V_c}{V_r}$$

4. For the relation volume of casting divided by surface of casting has to be taken the maximum value

- and the riser or head has to be put on that place. S_n

5. If feeding proceeds as anticipated depends largely upon the hydrodynamics of the case (ferrostatic and atmospheric pressure, time available and distance to flow), the thermal gradient in the direction of flow and the influence of this gradient on fluidity.

6. In metals with a solidification range, freezing proceeds through the whole mass simultaneously, obstruction at least part of the feeding.

7. It is of little use to calculate riser or head for a whole casting that only partly can be fed. The principles of how to feed a casting first have to be solved.

8. It is good practice to dispense with risers and heads as long as possible: wall thicknesses, height and kind of metal and mold being the limiting factors.

References

- 1. Saeger and Ash, "Volume Changes of Metals and Allovs During Casting," A.F.A. TRANSACTIONS, vol. 38.

 2. Saeger and Ash, "Volume Changes of Cast Iron During
- Casting," A.F.A. TRANSACTIONS, vol. 40, pp. 172-188.
- 3. Briggs and Gezelius, "Solidification and Contraction in Steel Castings I, II, III," A.F.A. TRANSACTIONS, vol. 41, 42, 43.
- 4. Chworinoff, "Theorie der Erstarrung von Gussstücken." Die Giesserei, 1940, Heft 10, 11 and 12.
 5. Taylor and Rominski, "Atmospheric Pressure and the Steel
- Casting," A.F.A. TRANSACTIONS, vol. 50.
- 6. Brinson and Duma, "Studies on Certerline Shrinkage in Steel Castings," A.F.A. Transactions, vol. 50.
 7. Scheurer, "The Solidification of Metals," Institute of British
- Foundrymen, Proc. 1949.
- 8. Ruddle, "The Solidification of Castings," The Institute of Metals, 1950, Monogr. No. 7.
- 9. Ruddle and Mincher, "The Thermal Properties of Some Non-Metallic Mold Materials," The Institute of Metals.
- 10. Caine, "Risering Castings," A.F.A. TRANSACTIONS, vol. 57. 11. Caine, "A Theoretical Approach to the Problem of Dimensioning Risers," A.F.A. TRANSACTIONS, vol. 56.
- 12. Schwartz, "Solidification of Metals," A.F.A. TRANSACTIONS, vol. 53.
- 13. Paschkis, "Heat Flow Problems in Foundry Work," A.F.A. Transactions, vol. 52.
- 14. Paschkis, "Studies on Solidification of Castings," A.F.A. TRANSACTIONS, vol. 53.
- 15. Paschkis, "Studies on Solidification of Aluminum Castings," A.F.A. TRANSACTIONS, vol. 54.
- 16. Paschkis, "Studies on Solidification of White Cast Iron," A.F.A. TRANSACTIONS, vol. 54.
- 17. Paschkis, "Studies on the Solidification of Steel Spheres,"
- A.F.A. Transactions, vol. 54. 18. Paschkis, "Influence of Properties on Solidification of Metals," A.F.A. Transactions, vol 55.
- 19. Paschkis, "Studies on Solidification of Metals," A.F.A. TRANSACTIONS, vol. 55.
- 20. Paschkis, "Influence of Dry Sand Conductivity on Rate of
- Freezing of Slabs," A.F.A. TRANSACTIONS, vol. 57. 21. Paschkis, "Comparative Solidification Studies," A.F.A. TRANSACTIONS, vol. 57.
- 22. A.F.S. "Heat Transfer Committee Reports," A.F.A. Trans-ACTIONS, vol. 53, 56, 57.
- 23. Fifield and Schaum, "Solidification Characteristics of Gray Cast Iron," A.F.A. TRANSACTIONS, vol. 56.
- 24. Bishop and Pellini, "A Contribution of Riser and Chill-Edge Effects to the Soundness of Cast Steel Plates," A.F.A. TRANSACTIONS, vol. 58.

NEW PROCESS FOR NODULAR IRON PRODUCTION AT LOW TREATMENT COST

By

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Cost of magnesium treatment is still the most important obstacle to wide production of nodular graphite cast iron. The present paper shows what was done for making possible the nodulizing treatment with direct use of pure magnesium or high magnesium alloys. We consider that we have gained an economic solution to this problem. More than a year of continuous industrial production confirms the possibility of getting safe and constant results with pure magnesium treatment.

Magnesium contents necessary for complete nodulization of a common gray iron are in the range 0.04–0.08 per cent, relatively low. Due to violent reaction taking place by magnesium introduction in a cast iron bath, it is necessary to use alloys with magnesium content under 50 per cent. Of common use are alloys with magnesium contents about 20 per cent.

Melting point of pure magnesium is 657C (1215F) and its boiling point is 1102C (1835F). Above this temperature the vapor pressure of magnesium increases rapidly, according to results from a diagram of Eucken, Hartmann and Schneider, reported by E. Piwowarsky (Fig. 1).

At mean temperatures of a gray iron submitted to nodulizing treatment, magnesium vapor pressure has remarkable range of 2.5 to 8 atmospheres. This is the reason violent reactions of explosive character occur when lumps of pure magnesium or magnesium-rich alloy are submerged in the iron bath. Immersion trials of pure magnesium in cast iron melts are reported in the technical literature; in every instance magnesium recovery was very low, from 2 to 8 per cent, and inconstant. Hence this treatment was not suitable for industrial application. Only low magnesium alloys then were in use in the foundries, in the form of binary and ternary alloys. At present those mainly used are alloys of nickel-magnesium, copper-

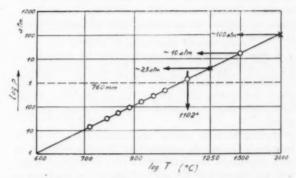


Fig. 1—Relation between temperature and magnesium vapor pressure.

magnesium, copper-silicon-magnesium, nickel-silicon-magnesium and ferrosilicon-magnesium, with magnesium contents about 8–20 per cent.

Magnesium recovery with these alloys was good. For instance, nickel-magnesium 80/20 alloy has a 50 per cent recovery when used for treating an iron melt at 1350C (2465F). This temperature, as other temperatures mentioned in this paper, were read with an optical pyrometer, without correction.

After about three years of experience in nodular iron production and study, the authors feel that low magnesium alloys are the safest alloys to introduce in cast iron melts, but their use presents many drawbacks.

Drawbacks of Low Mg Alloys Treatment

Inconveniences encountered in treating cast iron melts with low magnesium alloys can be summarized as follows:

1. Undesirable alloy elements introduced in the iron melt: If we treat, for instance, a cast iron with 0.1 per cent S, 0.075 per cent nodulizing magnesium plus 0.075 per cent desulphurizing magnesium, 0.15 per cent total Mg, should effectively pass in the melt.

Using Ni-Mg 80/20 alloy, having 50 per cent Mg recovery, 0.3 per cent Mg should be added; that is, 1.5 per cent of Ni-Mg alloy. We thus get in the cast-

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ings a nickel content of about 1.2 per cent.

Using Cu-Mg 80/20 alloy, having 30 per cent recovery, we should add 2.5 per cent of the alloy and we get a final content of 2 per cent Cu in treated cast iron.

2. Temperature drop of treated melts Foreign elements introduced in a cast iron melt with low magnesium alloys are an economic disadvantage, but they are also the cause of temperature drop of treated melt. With sulphur contents about 0.10 per cent and with 20 per cent Mg alloys, temperature drops are of the order of some 10C or 50F with maximum figures experimentally controlled up 80C (175F).

Temperature drop following low-magnesium alloys treatment makes it necessary to treat iron melts at temperatures much higher than temperatures desired for pouring in molds. In this manner increasing percentages of magnesium alloy additions mean treatment at higher temperatures and of course, lower

recoveries.

3. Pure cast irons are required: From the above listed drawbacks appears the necessity of submitting to nodulizing treatment only cast iron melts with sulphur and oxides contents as low as possible. These iron melts must be further delivered at high temperatures.

Having to treat small iron ladles, from 10 lb to hundreds of pounds, temperature must be regarded as very important. As a matter of fact, if sulphur is not low, e.g. above 0.08–0.1 per cent, using low magnesium alloys, cupola iron results often are not suitable for the treatment. Other furnaces should then be required, where melting is more expensive.

- 4. Only some iron quality is suitable to become fully nodular: As reported from technical literature and often confirmed in our own trials, only a few iron mixtures appear to become perfectly nodular if treated with due quantity of low magnesium alloy. It was found that when in a mixture we change only the quality of pig iron and maintain constant the C, Si, Mn, S and P, magnesium treatment results are different, referring to quantity of nodules in treated iron.
- 5. High production cos's Due to the previously mentioned drawbacks, production cost of nodular iron with low magnesium alloys treatment is high. This item can be unimportant when producing small or complicated castings, for which molding and other foundry operations are the biggest factors of total production cost; but it is clear that a high treatment cost for large quantities of castings can be so important as to prevent nodular cast iron from having a competitive position against malleable and steel castings.

Use of Pure and High Magnesium Alloys

From the foregoing considerations it can be concluded that an ideal solution of the problem should be the use of pure magnesium or high magnesium alloys, if that use could be accompanied by a high magnesium recovery and a safe treatment. Particular attention should be paid to prevent explosions or iron emissions out of the ladle.

This problem has been studied jointly by the Isti-

tuto Siderurgico Finsider and Ansaldo Foundries during the first months of 1951 and today it can be considered solved, although further improvements are possible. On the basis of the experimental work, it was then possible since spring of 1951 to do all daily nodular iron production using only pure magnesium or alloys of the "elektron" type.

The explosive effect obtained by introducing and also by simply putting the pure magnesium lumps into the iron melt, is due to the rapid formation of magnesium vapors at relatively high pressure. When plunged in the bath magnesium lumps act like an exploding body with liquid cast iron emissions from the ladle. When magnesium lumps are put on the bath surface we get in the same way a quick powerful magnesium vapor development, without liquid iron emission, but with very big flames due to oxidation of magnesium vapors.

The problem was to find a way to control the formation and development of magnesium vapor in such a manner that we could get a controlling bubbling of magnesium vapors through liquid iron. In other words, our scope was to let magnesium vapors pass through the melt like the air in a converter.

The production of magnesium vapors with a special generator outside of the melt and the following introduction of these vapors through the melt presented difficult problems. It was then thought to produce magnesium vapors within the melt under treatment in the ladle, using latent heat of the melt itself.

Treatment With Magnesium Vapors

First trials to get bubbling magnesium vapors within the cast iron melt contained in the ladle were done with special porous compacts constituted of small sized pure magnesium and inert material, bonded in a suitable manner and dried with care. Pure magnesium was in the form of chips produced by rough turning.

Inert material should possess good thermal conductivity, not react with magnesium at high temperature,

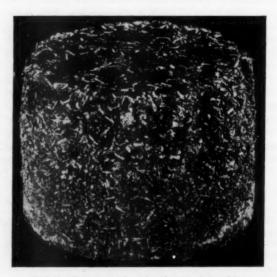


Fig. 2—Compact of magnesium and graphite.

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be refractory and have a thermal capacity as low as possible to prevent considerable heat abstraction from the melt. Graphite was used from crushed electrodes, which was found suitable.

Characteristics of bonding agent should be: no reaction with magnesium, good strength in the compact without sacrificing permeability, and to prevent compact breaking during immersion. Ethyl silicate, as well as sodium silicate, was successfully applied.

The compact slightly pressed in a core box, was dried in an oven like a common foundry core. Usual

form of the compacts is shown in Fig. 2.

There were some cylindrical holes in the compact. One of them was used to insert the extremity of a steel bar for maintaining the compact under the surface of liquid iron in the ladle. Other holes were useful for a better development of magnesium vapors.

Experiments were made with compacts containing various magnesium chip contents. Trials in that way gave interesting information toward final solution.

Ladles containing cupola iron were treated with compacts of graphite-magnesium in the volumetric proportion 50/50. These compacts when plunged in gave a regular continuous development of magnesium vapors, lasting from 60 to 120 seconds. All iron in the ladle received a strong boiling action by magnesium vapors but only small flames were on top of the ladle.

By chemical analysis of the melt before and after treatment, it was ascertained that magnesium from compact was passed in the melt with about 60 per cent

recovery.

An exhausted compact after the above treatment, still retained its original form. Examination of the compact after fracturing it at room temperature, revealed that throughout the entire compact pores were found where previously magnesium chips had been present, the magnesium chips having been entirely vaporized during immersion.

No significant results were obtained in the trials about temperature, because the quantity of iron in the ladle was too small and the time allowed for development of magnesium vapors was relatively long.

On the other hand, one important observation was made, which later proved to be essential, namely, that with a porous mass even though the magnesium content in it was very high, magnesium vapors did not develop either too rapidly or become explosive. The last step was attained when the porous compact of pure magnesium was made, using no diluent or bonding agent. These magnesium chip compacts were submerged in the melt, using a plunger of such form as to assure safe holding of the porous mass up to its complete consumption.

The plungers usually used are made with welded steel sheet, bored throughout, and painted with a refractory wash. They last for about 20 treatments.

Trial Results With Magnesium Compacts

First trials using porous magnesium compacts gave the following results:

 Treating many ladles in which iron was at the same temperature, magnesium recovery was practically constant. 2. When treatment temperature is varied, magnesium recovery varies from 25 to 35 per cent.

3. Magnesium vapor development lasted from 20 to 50 seconds. There was a rather big flame on the ladle top, but no liquid iron projection took place. A shield was sufficient to give safe protection against flames.

4. No temperature drop was discerned after treatment.

Trial results were very promising and allowed conversion of the entire nodular production of our foundry from the low magnesium alloys process to a new

pure magnesium process.

A steel hood with fume-exhausting duct was provided. Under this hood the ladles to be treated are placed. The plunger with porous magnesium charge is operated with lifting devices controlled from the outside of the hood.

Results Obtained with New Process

Characteristics of the process for nodular iron production with the use of magnesium chips are as follows:

- 1. Magnesium recovery: Essential factors governing recovery figures are:
 - Size of magnesium chips or particles used for charging the plunger.

b. Treatment temperature.

in equipment.

 Height of liquid cast iron over submerged plunger.

In order to obtain good uniform results the production of nodular iron with pure magnesium treatment must be standardized in operational phases and

At the Ansaldo Foundry cast iron from electric arc or graphite resistor furnace is treated in ladles containing from 500 to 5000 lb of metal. The ladles were treated under protecting hood at standard temperature. Standard temperature is changed only when particular characteristics of the casting to be poured require it.

Under usual working conditions, magnesium recovery ranges from a minimum of 25 per cent up to 35 per cent, with reference respectively to higher and lower treatment temperatures.

- 2. Inoculation with terrosilicon: Graphitizing inoculation can be made at the same time as magnesium treatment, providing the required quantity (75 per cent) ferrosilicon is put into the magnesium charge in the plunger. The percentage of inoculant varies with the desired characteristics of the castings, but usually it is about 0.6 per cent. The size of ferrosilicon is 0.25 to 0.5 inch. Using these sizes, the effects of simultaneous inoculation were found to be equivalent to those of an inoculation following magnesium treatment.
- 3. Temperature drop: On cast iron melts treated with a plunger containing the entire charge of magnesium chips and inoculant, no noticeable temperature drop was discerned. On the contrary, sometimes temperatures increased slightly (5-10C) after treatment.

It is possible to treat cast iron melts at temperatures below 1300C or 2375F (read with optical pryometer) with the advantage of better magnesium recoveries.

4. Treatment safety and uniform y: The usual mixture used at Ansaldo Foundries for the electric furnaces employed to provide cast iron for nodulizing treatment, is the following:

Pig iron55 per centNodular returns30 per centMild steel scrap15 per cent

When best mechanical properties are required, we take care to get melts with about the following chemical composition:

Total carbon3.3 per centSilicon2.0 per centManganese0.3 per centPhosphorus0.06 per centSulphur0.03 per cent

Using low magnesium alloys for treatment, which was Ansaldo's practice until a year ago, results were strongly affected by quality and composition of the pig. Although apparent chemical identity was present, at least for the elements which are usually determined by analysis, only a few pig irons which were used produced perfect nodular structures. It was common to talk in terms of per cent of nodules obtained in treated irons. Some pig irons, as for example a Swedish and a Russian pig iron which were in our inventory, gave mixtures unsuitable to nodulizing treatment.

With the new treatment process every pig iron quality of our stocks is successfully treated and when treated under standard conditions, graphite structure is always 100 per cent nodular. A year of systematic metallographic control confirms these results: 98 per cent of treated melts became fully nodular. The reason for this is not yet explained and studies are in process.

Consequently the present production of electric furnace cast iron shows a constant and equal nodulizing of cast iron. In the ordinary cupola we can regularly and successfully obtain cast irons which correspond to Meehanite GC and GD, and also special acicular and austenitic cast irons. In any case, the magnesium must be added if satisfactory results are to be obtained.

5. Production Costs

Good savings are effected with this process, because no expensive foreign elements occur in the treatment. Treated cast iron presents a chemical composition nominally equivalent to composition before treatment, except for sulphur and magnesium.

A practical example will show in more comprehensive terms savings which are possible using pure magnesium process instead of the low magnesium alloys process.

Referring to the beforementioned case, we have to treat a cast iron melt with 0.1 per cent sulphur using pure magnesium or copper or nickel alloys with 20 per cent Mg. With this sulphur content 0.15 per cent Mg should be effectively introduced to the bath. For the three cases additions will be as in Table 1.

For a comparative treatment analysis we can omit FeSi inoculation cost because it is constant by the various processes. Treating with pure magnesium there are major costs due to installation of the protection hood, of plungers consumption and to magnesium preparation. As average figures the three terms given in Table 1 have a 1000/ton incidence on nodular iron produced.

For the case under consideration and with nickelmagnesium alloy at official price, treatment costs can be summarized in Table 2.

On a cupola iron, for instance, available in liquid state at 75.000 lire/ton, and treated with 20 per cent magnesium alloys or with pure magnesium treatment costs will have on liquid treated iron cost an incidence as reported in Table 3.

6. Mechanical Properties of nodular cast iron produced with this process are equivalent to properties

TABLE 1-ADDITION ALLOYS

Alloy	% Recovery	% Mg Addition	% Alloy Addition	% Foreign Elements
Ni-Mg 80/20	50	0.3	1.5	Ni — 1.2
Cu-Mg 80/20	30	0.5	2.5	Cu — 2
Magnesium	25	0.6	0.6	_

TABLE 2—TREATMENT COSTS

Alloy	Alloy Cost, Lire/Kg	Extra Costs, Lire/ton	Additions, Kg/ton	Cost, Lire/ton
Ni-Mg 80/20	1000	_	15	15.000
Cu-Mg 80/20	1000	-	25	25.000
Magnesium	400	1000	6	3.400

TABLE 3-MAJOR TREATMENT COSTS IN PER CENT

Alloy	Treatment Cost, Lire/ton	Treated Liquid Iron Cost, Lire/ton	Major Cost per cent
Ni-Mg 80/20	15.000	90.000	20
Cu-Mg 80/20	25.000	100.000	33.3
Magnesium	3.400	78.400	4.55

TABLE 4—MINIMUM GUARANTEED MECHANICAL PROPERTIES FOR THE NODULAR CAST IRON TYPES OF NORMAL PRODUCTION

Туре	Tensile Strength, Kg/mm ^s	Yield Strength, Kg/mm³	Elongation, % on 4 cm	Impact Strength, Kgm/cm ^a	Brinell Hardness	Structure and Treatment
NP	60	40	1_3	0.30	240—270	Pearlitic
NF	00	10	1-3	0.30		As Cast
NPF	55	35	5	0.70	190-220	Semiferritio
NPF	33	33	3	0.70		As Cast
NFR 10	45	28	10	,	150—190	Ferritic
NFR 10	43	20	10	1		Annealed
NFR 15	45	90			1.5 150—180	Ferritic
	45	30	15	1.5		Annealed

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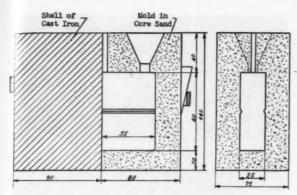


Fig. 3—Chill test for production control.

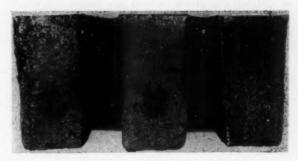


Fig. 4-Fractures of chill tests.

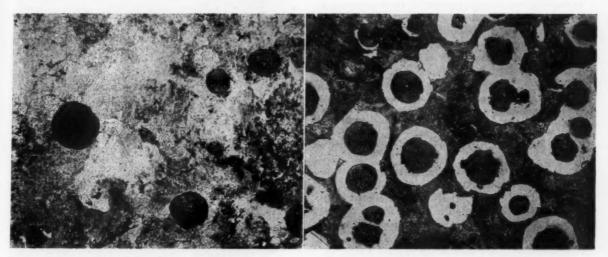


Fig. 5—Nodular cast iron of NP type. Pearlitic structure. 2% Nital etch. Mag. 300x.

Fig. 6—Nodular cast iron of NPF type. Ferritic-pearlitic structure. 2% Nital etch. Mag. 300x.

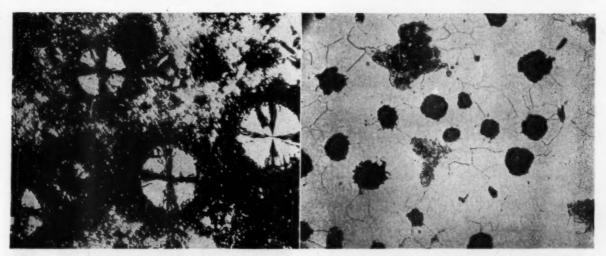


Fig. 7—Nodular cast iron of NFR 10 type, annealed, observed in polarized light. Around the nodules there is a crown of temper carbon deposited during annealing. 2% Nital etch. Mag. 300x.

Fig. 8—Nodular cast iron of NFR 15 type, annealed. 2% Nital etch. 300x.

of nodular iron produced with low magnesium alloys treatment.

It must be noted, however, that the absence of undesirable elements such as Cu and Ni, results in structures desired for metallic matrix of castings.

Common nodular irons produced at Ansaldo Foundries are of four different types, varying in mechanical properties and in metallographic structure. Table 4 reports main characteristics of these irons, Properties shown are minimum properties guaranteed in test bars machined from cast-on bars. Usually, however, mechanical property figures result in testing remarkably higher than guaranteed minimums.

Impact figures refer to notched Mesnager bar tested in a Charpy impact test machine. (UNI Tab. 565 = ASTM - E23 - 47T).

Obtaining mechanical properties corresponding to the four nodular iron types of Table 4, depends mainly on chemical composition of treated iron. In particular, type NFR 15 is easily obtained only when phosphorus and manganese contents are very low (under 0.1 per cent and 0.4 per cent, respectively). When these elements are higher than the above mentioned figures, in annealed castings we get characteristics corresponding to NFR 10 type.

7. Production Control

There are three phases of nodular iron production control.

a. Immediate control: In the melting shop before tapping the furnaces, chemical composition of cast iron is controlled, particularly with reference to silicon and carbon content, with a chill test as reported in Fig. 3. Eventual corrections are made in the furnace after chill test observation in such a way as to get the desired characteristics in tapped metal.

Following the magnesium treatment a new chill test specimen is immediately poured. The specimen after chill testing, is fractured and the fracture is examined and interpreted.

- (1) For fracture color, which if treatment is successful, will be bright and brilliant;
- (2) For fracture conformation, varying from NP to NPF type. Fracture of NPF type is markedly conchoidal.
- (3) For chill depth to ascertain that the characteristics of the iron are suitable for the mean section of the castings to be poured.
- (4) Shrinkage trend. The shrink formed in the chill test, which appears on the fractured section as a dark spot, is a good indication of the shrinkage trend in the treated iron. This nodular iron chill test further indicates two depressed zones symmetrically located on major surfaces resulting from atmospheric pressure during solidification.

In Fig. 4 examples are given of chill tests: at the left untreated cast iron; at the center NP nodular cast iron, and at the right NPF nodular cast iron.

b. Metallographic and chemical control: Structural characteristics of nodular iron are inspected and controlled as routine work in the metallographic laboratory. At Istituto Siderurgico Finsider research is being done in the field of heat treatment and studies are



Fig. 9—Gear covers for tractors.

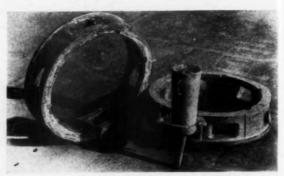


Fig. 10-Clutch rings for tractors.

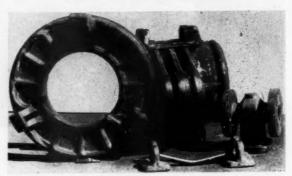


Fig. 11—Clutch plate, railway axle sleeve and high pressure valve body.



Fig. 12—Covers and sleeves for railway car axles.



Fig. 13-Railway axle sleeves.

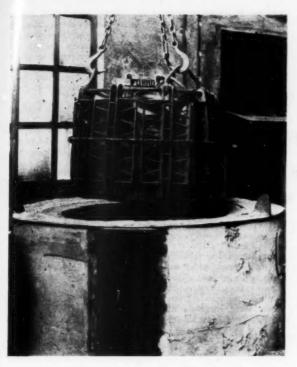


Fig. 14—Charging of an annealing furnace.

being made of nodular graphite formation and growth.

In these studies, improved methods for metallo-

graphic observation are widely used, based upon specimen coloring with heat oxidation.

Characteristic structures of normal production nodular iron are shown in Figs, 5, 6, 7 and 8.

In this control phase chemical analyses are included. Magnesium content in castings is also chemically controlled.

c. Mechanical properties control: Control of mechanical properties of castings is usually done on test bars machined from cast-on bars poured with the castings. Special care is taken in positioning and risering these appendices to get them entirely sound. Only exceptional test specimens are machined from keel blocks poured separately from the castings.

8. Applications: Production of the nodular cast iron department of Ansaldo Foundry consists of small and medium castings, but when required, castings up to about 2 tons were successfully produced.

Nodular cast iron is largely used in substitution of castings, which formerly were made of steel, and these castings are, at present, the major part of our production.

In Fig. 9 are some tractor gear covers, and in Fig. 10, clutch rings. The casting at the left in Fig. 11 is a clutch plate for tractors, in the center there is a railway axle sleeve, and at the right a high pressure valve body.

Two other types of railway axle sleeve are shown in Figs. 12 and 13. Figure 14 shows an annealing box of axle covers, during its charging in one of the electric annealing furnaces.

Experimental production of nodular iron ingot molds began some months ago. Fig. 15 is a nodular iron slab mold for 1-ton slabs. This type of ingot mold is now being used in steel work and the results appear to be promising.

The new process using pure magnesium, due to its low treatment cost, will undoubtedly find large application in the production of nodular iron ingot molds, if trial run results confirm superiority of nodular iron in ingot mold application.

Acknowledgments

Authors are indebted to Istituto Sigerurgico Finsider and Ansaldo S.A. Directors, who permitted publication of the present paper.

DISCUSSION

Chairman: J. S. Vanick, International Nickel Co., New York. Co-Chairman: W. S. Pellini, Naval Research Laboratory, Washington, D. C.

Recorder: Chas, Mooney, Olney Foundry Div., Link-Belt Co., Fhiladelphia.

A. P. GAGNEBIN: 1 Dr. Longaretti and Dr. Noris are to be congratulated for their work in connection with developing a method for the direct addition of magnesium to melten iron. The Inco-Mond patents, which are based on the discovery that spheroidal graphite can be produced in gray cast iron by the introduction and retention of magnesium in the iron, make it clear that magnesium can be introduced into molten iron either in elemental or alloyed form and that it is not necessary to introduce the magnesium in the form of any particular alloy as has been sometimes believed by others.

¹ The International Nickel Co., Inc., New York.



Fig. 15-Nodular iron slab mold.

We have recognized that there is room for improvement in the methods of adding magnesium and developments in this direction will reduce the process cost and obviate the introduction of alloying elements which sometimes might not be required. Our attention has been drawn to an abstract of Italian Patent No. 457,374 which proposes the use of iron and steel containers having calibrated holes and filled with magnesium particles which are then plunged into the molten iron. If the authors are familiar with this patent, they might wish to comment on it in relation to their own work since the methods appear to resemble each other.

We have a few questions that we would like to direct to the authors. What size ladles do the authors prefer to treat with their method? What is the maximum size that can be treated? Can they tell us how the magnesium recovery varies with the temperature of the molten iron? For example, what magnesium recovery is achieved at 2800 F (about 1540 C), 2700 F (about

1480 C) and 2600 F (about 1430 C)?

Messas, Longaretti and Noris (Written Reply to Mr. Gagnebin): In this paper we made no reference to patent questions. We are not familiar with the details of the patent Mr. Gagnebin refers to. We are certain that the process described in this paper is at the present time the only one used successfully on an industrial scale.

We have no preference as to the size of ladles to use for the treatment with our method. We commonly use 700-lb capacity ladles, but ladles of some tons capacity have been used in this treatment.

We find in the American literature cast iron temperatures as high as 1450-1550 C (2640-2820 F). We think the temperature

readings in American practice are usually considerably higher than the European readings. This is probably due to the fact that we record the temperatures just as we read them with the optical pyrometer, without correction. In every case in the temperature range where we treat a cast iron melt to get it nodular, the magnesium recovery varies from 20 to 30 per cent. Owing to the absence of a temperature drop due to the treatment, our temperature range is between 1340 and 1300 C (2445 and 2375 F).

In answer to a question on the difficulty to see chill in nodular iron chill tests, in the chill test samples shown in Fig. 4, it is difficult to see on the fracture a clear difference between the chilled and non-chilled parts. That depends only on the photographic reproduction; when observed by eye the chill depth is

very easy to control.

In answer to the question regarding the percentage of recovery and how it is calculated, the figures on magnesium recovery referred to in the paper are calculated with the same method and formulas, which are reported in the literature as commonly used in this connection.

In answer to the question whether we tried Mg-Si and what the recovery was when this was used, we did not try this method of addition. Usually the charge of the plunger is made by magnesium chips (about ½x1/2 in.) mixed with some 75 pct ferrosilicon. added as inoculant. Often there is room in the plunger to hold at the same time magnesium chips and all ferrosilicon; sometines only magnesium chips are introduced in the plunger, while ferrosilicon is added later in the ladle. Under these conditions the magnesium recovery was the same as with the compound charge.

GASES AND NATURALLY OCCURRING (CONGENITAL) BLOWHOLES IN FOUNDRY PRACTICE

By

Albert M. Portevin*

The problem of gas and blowholes in the foundry frequently has been discussed and numerous papers have appeared on this subject e.g. by C. H. Lorig, Budgen, etc. But the majority of these papers deals either with special processes or is restricted in scope to certain alloys, be these steels, cast irons, copper alloys, aluminum and light alloys. Thus it seemed to be in order to present the topic in such a way that all aspects of foundry practice are covered, drawing amply on the work of metallurgists and steel specialists on ingot blowholes, especially those in agreement with our long-time teachings at the "Ecole Superieure de Fonderie." †

Furthermore this Exchange Paper shall serve as a gesture of sympathy, collaboration and gratefulness toward the American Foundrymen's Society.

This paper is subdivided in the following headings; Definition of blowholes;

State of the gas in the metals;

Origin of blowholes in the foundry;

- (1) Not originating in the metal: exogenous blowholes
- (2) Originating in the metal: endogenous blowholes, or naturally occurring ones.
- A. Dissolved gases—Solubility of gases in metals: Factors: Temperature, pressure, composition of the alloys, nature of metal, supersaturation, agitation and nuclei

Segregation at solidification.

- B. Gases due to the reaction— Two classical examples:
 - (1) Copper, Cu2O and H (S and C)
 - (2) Steel, FeO and C. Blowholes in steel ingots and in rimmed steels.

C. Gases in suspension-

Evaluation of the inclination for naturally occurring blowholes:

(1) Gas content

(2) Frothing under change of pressure.

Character of the blowholes: Microblowholes and microshrinkage-holes (pipes).

General procedures for combating naturally occuring blowholes (degassing processes).

- (1) Chemical ones
- (2) Mechanical ones
- (3) Physical ones.

Definition of Blowholes

Blowholes are cavities resulting from entrapment or retention of gas bubbles from the start of solidification of the metal.

Thus the cavity or the blowhole at its very beginning of formation is occupied by a gas the pressure of which is equal to or greater than the sum of the pressures of the surrounding liquid metal and of the atmosphere above the melt,* and of the surface tension; the shrinkage hole, on the other hand, from its very beginning is evacuated, therefore gases penetrate these holes with a pressure lower than the sum of the pressures of the liquid metal and of the atmosphere over the melt.

The blowholes represent a state of mechanical retention or inclusion of gases in metals.

$$P = P_D + \frac{2T}{r}$$

where T is the surface tension of the metal. One should however note that according to this formula the pressure inside a bubble at the formation (r = 0) should be infinity, which, of course, is never the case.

Thus this equation can no longer be applied for very small values of r; Allen (Jour. Inst. of Metals, vol. 49, 1932, p. 317) expressed the opinion that the first resistance which a bubble has to overcome at its formation is the surface tension, therefore he prefers to write $P = P_D + \sigma \ (r)$ where σ (r) is a function of the surface tension, which in the case of nuclei, unrestricted within the range of thermodynamic equilibrium conditions, may for practical purposes be set equal to zero (see Herty and collaborators).

[•] Pressure P in the interior of a bubble with the radius r, under the hydrostatic pressure P_D , is equal to

[•] Member of the "Institute de France," Honorary President of the "Association Technique de Fonderie" and Honorary Director of the "Ecole Superieure de Fonderie" of Paris, France.

This is the Official Exchange Paper from the Association Technique de Fonderie of France to the American Foundrymen's Society. It was translated from the French.

[†] About 600 of our former students have been placed in the foundry industry, therefore it is not at all surprising that certain parts of the present paper duplicate papers presented and/or published in France.

States of Gases in Metals

A. At the surface: superficially adsorbed gases;

B. In the interior of the metal.

(a) dissolved gases; example: H in Al at saturation or supersaturation:*

(b) combined gases; example: FeO, Cu₂O, etc., compounds either dissolved (example: Cu₂O in liquid Cu), or not in solution or precipitated (inclusions; SiO₂, silicates, Al₂O₃).

(c) gases in inclusions, or mechanically entrapped

(sometimes called occluded gases)

This complete classification is applicable both to

the liquid and the solid states.

In the solid states the gaseous inclusions (Class C) form the macro- or microblowholes; these are the only ones which interest us here. Thus we deal with the gas of gaseous inclusions or blowholes only, not the action of combined or dissolved gases, or of solid inclusions. Thus we shall not discuss the influence of oxygen, nitrogen etc. on ferrous metals especially on their grain size, formation of cementite, the capability of being tempered, etc. Blowholes originate from gases which exist in the liquid metal in one of the three above-named categories; we shall point out in this discussion which category we are dealing with.

The term "porosity" may refer to both microblow-

holes and micro-shrinkage-holes.

The gases contained in industrial metals and encountered in the foundry may originate from:

(1) Methods of Fabrication—This concerns metals in the liquid state, viz. the amount of nitrogen and oxygen (in steels) may depend upon the production method (i.e. Bessemer converter, open hearth, electric furnace) and for each method the actual course of the operations accounts for great differences.

Among the methods used for the production of metals in the solid state, mention should be made of electrolysis, because here the deposits may contain a

considerable amount of hydrogen.

(2) Treatments of the Solid Metals—Thus the rolling of aluminum in a moist atmosphere considerably increases the amount of hydrogen (Chaudron).

The pickling e.g. of iron in sulphuric acid or of aluminum in soda may considerably increase the

amount of hydrogen.

But all this may be changed by a foundry operation, and this operation itself introduces new gases which give rise to foundry blowholes. Usually blowholes may occur when the gases are liberated in the solid state.

Generally, if the gases cannot leave the ingot by diffusion, they develop into discontinuities in metal, i.e. physical defects such as *blowholes* if this inhibition of diffusion takes place during solidification or even in the solid state at temperatures high enough that the steel is still plastic (bubbles in the old-fashioned blister steels), and flakes (in steel) if this inhibition of gas diffusion takes place in the solid state and at low

enough temperatures. (This becomes especially noticeable as consequence of the $\gamma \to \alpha$ transformation which takes place at the temperature Ar''' during quenching of certain special steels). Also intergranular microblowholes may be brought about in this way.

Origin of Foundry Blowholes

The origin of the gases which form the blowholes may be:

- (1) Extrinsic to the metal, exogenous blowholes; they depend upon the design and construction of the ingot mold and on the pouring of the metal into the mold.
- (a) Mechanically entrained air, entrained by the liquid metal from sources outside the charge. Adherence phenomena between gases and liquids used in water-blast. One may avoid this by keeping the runner full (or choked). The amount of gas entrained depends upon the rate of pouring; more gas is retained by the metal when the fluidity diminishes.

(b) Air contained in the mold which could not escape when the liquid metal was poured into it.*

(c) Gases and vapors given up by molds, cores and coatings upon heating by the molten metal.

(d) gases created in a reaction between the molten metal and certain parts of the mold. Example: cast iron reacting with a rusty core chaplet and thus developing CO.**

(2) Interior of the metal—Endogenous or naturally occurring blowholes depend upon the metal itself, on its state, and its prehistory during solidification. We shall discuss in the present paper only such blowholes. With respect to the state of the gas in these metals, we distinguish:

^{*} It is of importance that the mold must not be filled too rapidly. The entrainment of air while filling the mold becomes of special importance in die casting, where the metal is injected into the mold and somewhat atomized. The gates are restricted to a thickness of 0.3 to 0.6 mm (0.01 to 0.02 in.); thus the fluidity and not the "castability" of the metal, plays the important role.

Also in the "gocseneck" hot chamber machines in which a pressure of 375 to 675 psi (24-45 kg/cm²) is exerted upon the liquid metal by means of compressed air, and even in the cold chamber machines where a pressure of 1050 to 1800 psi (70 to 120 kg/cm²) is applied by means of a steel piston, it is almost impossible to get castings which are absolutely free from blowholes. In spite of all the mold vents, part of the air therein cannot escape when the metal is poured in. This air is compressed and its volume, distributed over a great number of minute blowholes, is thus greatly reduced. This air remains in the finished casting practically at the pressure under which the metal was forced in the mold.

Thus blowholes in an alloy cast at low pressure will be greater for the same amount of air, as compared to the same alloy cast with a cold chamber machine. If such castings be subjected to subsequent treatment at elevated temperature such as in baking on of varnish or other thermal treatments for certain light alloys, the problem of air inclusions becomes accentuated. With rising temperature the creep strength diminishes and the pressure of the entrained gas rises; at a certain moment the gas pressures will overcome the resistance against flow in the blowholes near the surface, blisters are now formed rapidly. Between two deeper situated blowholes a crack may now be formed, and therefore the mechanical properties of such a casting after heat treatment may not be improved at all, but might even have become worse, because the internal flaws have increased, and these are not apparent at the surface.

^{*} In the solid state (crystallized) the H atoms contribute to the distortion of the lattice; this causes an increase in hardness. More generally speaking the gases are in the interstices of crystalline structures, mcsaic structures, micrographic structures (intercrystalline microblowholes).

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(A) Gases dissolved in the liquid metal, being discharge during the cooling or the solidification of the metal.

(B) Gases due to chemical reactions which may take place in the molten metal; such reactions are influenced by temperature, pressure, concentration of the everal elements and by the solidification phenomena (as these bring about changes in concentration of the components). Such bases also may originate from the dissociation of compound, one element of which, is a gas.

(C) Gases in suspension or in emulsion which are due to the two causes cited above; and whose elimina-

tion is delayed.

These absorbed or entrapped gases are either eliminated from the metal or are retained in it as blowholes.

(A) Dissolved Gases

Solubility of the gases in metals and alloys

(a) Metals—The solubility or absorption of gases and their precipitation or expulsion in and from solid and liquid metals depends upon temperature and pressure. Therefore this solubility is either plotted as curves at constant pressure, mostly 1 atmosphere (isobars), or at constant temperature (isotherms). These curves furnish as a function of either temperature or pressure, the amount of gas, q or m, respectively, which is dissolved or absorbed. The quantities are expressed either by volume (cm³ at 0° C and 760 mm) per unit volume of metal, or in mg per 100 g of metal, or in mole per mole of metal, respectively.

The greatest number of investigations deals with hydrogen. There are some papers dealing with oxygen, nitrogen, sulphur dioxide, the oxides of carbon. The rare gases are practically insoluble in all the metals.

(1) Pressure and state of the gas—The solubility increases with the square root of the pressure: $q = k \sqrt{p}$, or in the more general form $q = k \cdot p^{1/n}$ with n = 2 prevalent.

That means that the dissolved gases are in the atomic state, and this explains the rapid absorption of atomic gases by molten metals, e.g. the absorption of nascent H derived from the chemical reaction of liquid metals with their contact with water or ammonia.

(2) Temperature—The solubility of gases increases with the temperature in molten metals. This is just the opposite of what takes place with water.

At the moment solidification sets in, the solubility of gases decreases abruptly: a typical example thereof is the sprouting of silver, known since antiquity, having been studied by Lucas 1819, Chevillot 1820, Gay-Lussac 1830, and J. B. Dumas. Here the volume of discharge oxygen represents about 100 times the volume of the silver at the melting point. The same phenomenon may be observed for platinum and palladium, if both metals are molten in the presence of oxygen.

If m is the solubility (in cm³/g of metal) of hydrogen in metals, then according to Borelius

 $\log m = f(1/T),$

where T = the absolute temperature; this equation is plotted (Fig. 1) for all the metals as a family of straight lines which converge toward a narrow limit of the I/T axis.

And furthermore, in the case of hydrogen, the change of solubility at the transition solid-liquid is

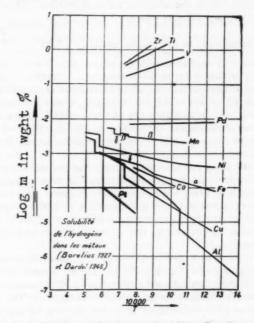


Fig. 1—Solubility of H₂ in metals (After Borelius and Dardel).

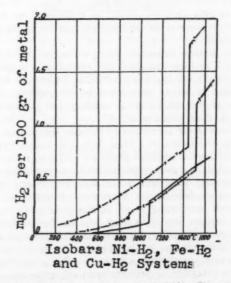


Fig. 2—Absorption of H₂ by metals (After Sieverts).

^{••} In the same way the blowholes near the surface of steel ingots, situated several mm up to 1 cm deep, are often due to the reaction of the carbon containing metal with the oxidized walls of the ingot mold (poor maintenance) or with the moisture condensed thereon (ingot mold was too cold, in general blowholes occur for this reason at the upper part of the ingot only), or they occur if the pouring was interrupted, and the new molten steel reacts with oxidized metal (in this case the blowholes are arranged in a zone at the spot where the pouring was interrupted, and finally they may be due to turbulence of the molten metal if it was poured too fast (blowholes at the bottom of the ingot).

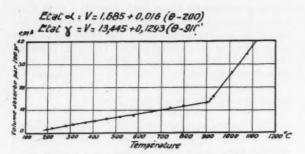


Fig. 3—Absorption of H_2 by reduced iron and H_2 . (Keizo Iwase).

larger, where the absolute value is smaller. Thus at the melting point we have:

Solubility m_1 in the liquid metal Fe Al 2.93

Solubility m_s in the solid metal Log solubility H₂ in solid metal 2.922 5.28

Hydrogen shows this abrupt solubility change at the melting points of Fe, Ni, Cu (Fig. 2 and 3), also at the $\alpha \rightleftharpoons \gamma$ transformation of iron, (which shows the same phenomenon for nitrogen, Fig. 4). Thus these metals show sprouting, and they become porous if the gas cannot be discharged.

We know on the other hand, that the diffusion of hydrogen into certain solid metals at high temperatures may bring about important changes: thus steels are decarburized with the formation of hydrocarbons, and oxidized copper forms water vapor which is insoluble in the metal, and accounts for its deterioration.

Certain rare, precious metals, or those of high melting point (Pd, Ce, Th, Zr, Ti, Ta, V) absorb in the solid state large amounts of hydrogen (1 volume of Ti dissolves 1800 volumes of H₂; 1 volume of Th dissolves 1700 volumes of H₂); the amount of gas absorbed diminishes with rising temperature (See Fig. 5). Hydrides or solid solutions of hydrogen are obtained with a density lower than the one of the metal and of gray or black color. In palladium we find a rapid decrease of the hydrogen solubility at the melting point.

(b) Alloys—In molten alloys the solubility of gases varies with their composition. Therefore it is possible to increase the hydrogen solubility in copper by the addition of nickel, decrease it by the addition of tin or aluminum, and leave it uninfluenced by the addition of the addition

tion of silver (See Fig. 6).

However the formation of definite compounds may change the solubility. Thus such compounds like Al-Cu and Mg-Cu addition alloys are prepared by uniting the metals in a ratio approximating the formation of a compound; these alloys are very brittle and easily pulverized due to the formation of numerous fine channels originating in the discharge of gases.

As regards the solid state, the studies concerning hydrogen solubility in palladium and ferrovanadium show different results (See Fig. 7); sometimes there is a decrease, sometimes an increase up to a maximum.

It is necessary to perform synthetic experiments to be sure that the porosity or lack of compactness of a metal is really due to gas.

Thus G. L. Bayley (Rev. Metal., Abstracts 1929; p. 204) has studied for 70:30 brass the action of hydro gen, nitrogen, and sulphur dioxide by comparing the compactness with the apparent density; he drew the conclusions that under varying conditions of casting the gases have nothing to do with the porosity of 70:30 brass. This may be explained by the fact that the vapor pressure of zinc in the liquid brass is rather high (about 200 mm at 1000 C or 1830 F, it reaches 760 mm at 1140 or 2085 F, a temperature at which occurs a rapid increase of the discharged vapors). Thus presumably a solution of the gases in the alloys is made impossible. We know that the solubility of a neutral gas in a solvent decreases as the vapor pressure of the solvent increases, becoming often equal to zero at the boiling point.

One may modify the solubility of gases by additions. Thus, e.g., the solubility of hydrogen in aluminum diminishes slightly if copper is added (Budgen), and diminishes greatly if manganese is added (Chaudron).

Summarizing, the absorption of a gas at the fusion of a metal depends, for any given metal upon:

(1) The chemical nature of the atmosphere of the furnace particularly reducing or oxidizing. A good example is the melting of manganese under hydrogen when the magnesium industry was still in its infancy;

(2) The state of the gas, i.e. molecular (O₂, H₂, N₂) or atomic. The absorption of atomic gases proceeds much easier than the absorption of molecular ones (by bubbling through). Thus aluminum absorbs (Chaudron):

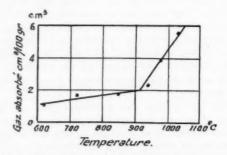


Fig. 4—Absorption of gas by reduced iron and N₂ (Keizo Iwase).

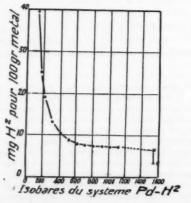


Fig. 5—Absorption of H2 by Pd and H2 (Sieverts).

(a) The hydrogen from its action upon water vapor (moisture in the air); if droplets of water are added in evenly spaced time intervals to crucible furnaces, a very gaseous metal is obtained;

(b) Nitrogen, by the action of ammonium salts;

(c) Carbon monoxide, by the action of carbon dioxide.

Therefore it is often advantageous to maintain a slightly oxidizing atmosphere. One must always avoid the chance that the metal should come in contact with atomic gases. Water vapors and ammonia come under this heading, as they furnish hydrogen and nitrogen respectively, by dissociation.

The alternative is to protect the metal with a flux

or a slag

(2) The temperature and the length of the melting. All melting should be done in as short a time as possible, and overheating of the melt shall be avoided.

(3) The *pressure* could be *total* (melting in vacuo or *partial* (dilution of the absorbable gases by inert gases, such as fusion under argon).

Supersaturation, Agitation and Nucleation

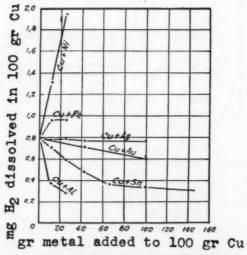
The dissolved gases remain frequently in a state of supersaturation after solidification has occurred. In such a case agitation relieves the supersaturation, a phenomenon well known from carbonated beverages; but also solid particles (inclusions) present or gas bubbles may act as nuclei for the gas development. Thus we have a means to degas a metal.

Degassing may be done by bubbling a gas through the metal, different from the one dissolved in it; a double action takes place:

(1) The liquid is agitated mechanically,

(2) Gas bubbles are created with an internal atmosphere where the partial pressure of the dissolved gas equals zero.

From these two actions the second one plays perhaps the more important role, as the hydrogen treatment of steels proves.



Alloys Prepared by Melting Copper with other Metals.

Fig. 6-Solubility of H, in alloys (Sieverts).

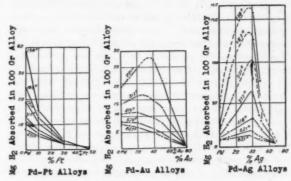


Fig. 7-Solubility of H2 in alloys (Sieverts).

As we shall see below, in the case of steels the boiling and the blowholes are due to the reaction gas, i.e. the carbon monoxide which is created due to the simultaneous presence of dissolved carbon and dissolved oxygen in the metal. This boiling may be reduced by killing the steel, i.e. reducing agents are added which form irreducible oxides, irreducible by the carbon. But it also may be enhanced by carburizing an oxidized heat or oxidizing a carburized one.

But there are other gas developments, one sees them frequently in deoxidized metals; they are due to a hydrogen absorption (the hydrogen might come from the moisture of the raw material, or from leaks into the cooling units). In this case the addition of deoxidizers of course is without effect. One must treat the heat in such a case:

(1) By recarburizing;

(2) By adding ore. The metal boils up and thus

becomes degassed.

This is what one has to do in the case of hydrogen if simple stirring is ineffective. By the same token the addition of limestone flux in place of CaO gives similar effects due to the development of CO₂. Thus we draw the conclusion that the mechanical stirring is of secondary importance.

Segregation at Solidification

Just as for all the other dissolved elements, the concentration of gas also changes during the solidification, and as the gas dissolves less in the solid, more gas concentrates in the liquid as solidification proceeds. Thus a metal or an alloy which was not saturated with gas when completely liquid, may now become saturated and discharge gases which at the second step or definitely at the third step cause blowholes (microblowholes).

Gas Content in Metals

Much progress has been accomplished in controlling the gas contents, especially in steel. For all special steels the amounts of O, H, N are determined in the shop practice of steel mills.

But the determined amount of gases has never shown conclusively that gases do damage; as regards physical defects, i.e. blowholes and steel flakes. Thus far no relation has been established between gas content and proportion of such defects.

Therefore it does not make much sense to determine

the amount of gas in the liquid or in the solidified ingot. We would have to know the damaging or dangerous amount, and this amount depends not only upon the alloy, but upon the overall foundry practice. Furthermore the solubility of a gas per se is of no great importance. The change of solubility as function of temperature and solidification together with the retardation phenomena such as the supercooling, crystallization, etc. become important. Thus one deals here with a problem of physico-chemical kinetics. We find in the literature (G. Chaudron, Bull. A.T.F. p. 410, 1937) reports to two samples with almost identical gas content one evolves gases, the other does not

Furthermore it was never possible to establish from basic open hearth heats any relation between the hydrogen content and the boiling of the bath. There was also no relation found between the change of the hydrogen content and the speed of decarburization.

However it was established at the basic open hearth which furnished steels with a 4 to 6 per cent Cr that rimmed steel was produced if the amount of hydrogen was 9 to 10 cm³/100 g, and that the steel was always satisfactory if this amount dropped to 7.5 cm³/100 g.

Nowadays we must admit, based on the work and experiments by Houdremont and his collaborators, which were verified independently by Musatti and Reggiori, that hydrogen is the most important cause for "flakes" in steels. If one examines the quantity which the latter authors applied (Rev. Met. vol. 32, p. 550, Nov. 1935) one finds no differences for steels with flakes and steels without flakes. The authors make the following statement: "Analyses of samples rich in flakes and analyses of those completely devoid thereof, do not allow us to reach conclusions as to the amount present in either case." Furthermore they state "that our values force us to draw the conclusion that hydrogen is almost completely eliminated with slow cooling, for the avoidance of flaking and with fast cooling for the promoting of flaking."

One must keep clear in mind that there are two properties, one is a physical one, simple and elementary, such as the solubility of a gas in a given metal; the other is a complex technological one, the ability to get full of blowholes, ("blowholiness"), which is a function of the above-mentioned simple property and its variation with the temperature. The gas solubility is determined by chemical analysis at different tem-

peratures.

But the second property, though being a function of the first, can only be determined by performing sprouting experiments at appropriate pressures.*

Thus one must do experiments, like solidification under reduced pressure or "incubation in vacuo," to determine the blowholiness.

(B) Reaction Gases

Equilibrium reactions with the dissolved materials take place inside the molten metal or alloy and these reactions give rise to gaseous compounds. We shall give some classical examples which are of fundamental importance:

(1) For steel: (Fig. 8)
FeO + C
$$\rightleftharpoons$$
 CO \uparrow + Fe
 c_1 C_2 p

(2) For copper (Allen, Fig. 9):
$$Cu_2O + 2H \rightleftharpoons H_2O \uparrow + 2 Cu$$
 $c_1 \qquad c_2 \qquad p$

this reaction also may take place in brasses, where ZNO and SnO₂ take the place of Cu₂O.

The equilibrium reaction under isothermic conditions are governed by the law of mass action $c_1 \cdot c_2 = Kp$. Under isobar conditions this law is replaced by Van t'Hoff's isochore $\partial \log K / \partial T = Q/RT^2$.

If one wishes to avoid the gaseous discharge, one must remove either the oxidant or the reducing agent from the reaction, or at least lower its concentration considerably, below the equilibrium concentration in the molten metal.

This equilibria are shifted by changes of tempera-

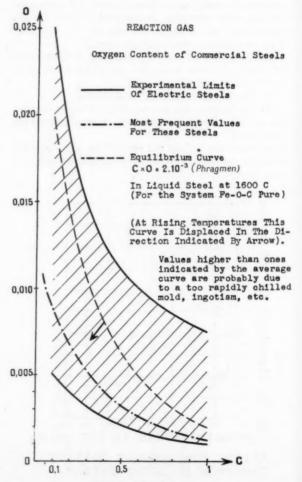


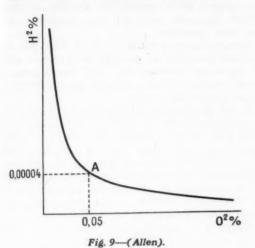
Fig. 8-Reaction gas.

^{*}An analogy is presented by the castability. If we want to know if a heat is suitable for pouring we cast it experimentally in a spiral, and this tells us if the heat is satisfactory. This ability too depends upon a simple physical property, the fluidity, which is a function of temperature.

In the same way we determine piping and shrinkage by direct experiments. We cannot deduce them from the basic physical property, the thermal expansion upon melting which is one of the factors.

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tune, pressure and phase concentrations. Upon solidification both concentrations c_1 and c_2 rise simultaneously in the liquid phase. This gives rise to a very important gas development, called sprouting resembling effervescence in carbonated beverages.



These reactions must be thoroughly watched both in copper foundries and in steel mills. We shall talk about this now more in detail.

(1) Copper—The removal of impurities from copper (called scrubbing, purifying, refining) is done with slags. Oxidation occurs, and the partly oxidized bath is reduced by stirring with a pole ("poling") until a certain amount (about 0.05 per cent) of oxygen remains. If the oxygen is reduced still further, defective ingots would result which are full of blowholes and the mechanical properties would be affected.

The same difficulties present themselves if one pre-

pares cathodes of electrolytic copper.

Therefore every commercial copper ingot contains numerous microblowholes interspersed between the dendrites and at the interspace between the grains. Therefore its apparent density is only 8.3 to 8.7, whereas pure rolled copper has a density of 8.95. Any annealing in a reducing atmosphere changes its mechanical properties considerably.

According to Allen (J. Inst. Metals 1930, p. 43-81) the defects of copper ingots are due to the equilibrium

reaction:

 $H_2 + Cu_2O \rightleftharpoons 2 Cu + H_2O$

This reaction gives rise to water vapors from the beginning of solidification due to traces of residual hydrogen.

As scon as the amount of oxygen exceeds 0.1 per cent (Fig. 9), the amount of hydrogen becomes less than 0 00002 per cent, and this amount can no longer do any damage. The ingots have a pipe, but during poling the oxygen diminishes, hydrogen increases, the pipe becomes smaller and smaller until one reaches point A (Fig. 9) with ingots without a pipe, and with a smooth surface. If oxygen is cut down beyond this point, hydrogen increases tremendously. Thus overpoling is obtained, the ingots puff up inside in the molds and they are useless.*

One may remove the oxygen completely with a

flux of boric acid anhydride or of metaphosphoric acid, as these fluxes dissolve Cu_2O and thus prevent oxide absorption. Melting removes the major part of the hydrogen. Upon tapping oxygen is absorbed by additions of Mn, Mg, B, Ca, e.g. one uses 0.02 per cent Mg. These small additions do not influence the conductivity; it is quite remarkable that up to 0.1 per cent oxygen shows hardly any influence upon the conductivity.

Even copper which has solidified with formation of a pipe may not have a perfect compactness in the

density test.

A calcium addition protects the copper in the liquid state (tapping) as well as in the solid state (heating for rolling).

An addition of 20 gr Ca/ 100 kg Cu suffices to cause

cracking of the copper at rolling.

If a copper wire is heated in air preceding the rolling, oxygen is absorbed. But a gas development may also take place due to the action of Cu_2O upon either sulphur or carbon. This is because in the reaction $Cu_2S + 2$ $Cu_2O \rightleftharpoons SO_2 + 6$ Cu develops an equilibrium pressure of 1 atmosphere for SO_2 at 750C (1382F). Sprouting or violent gassing occurs at this temperature. The carbon which is insoluble in copper may give rise to the reaction $Cu_2O + CO \rightarrow CO_2 + 2$ Cu, after the molten copper has been agitated with a stick of graphite.

Thus in summarizing we might say that Cu2O may be removed either by the action of reducing substances like P, Si, Al, B, Ca, Be, etc., or by flux absorption. The flux must have an acid base, like boric, silicic, metaphosphoric acid. One may apply both things simultaneously. Hydrogen is removed by the action of oxidants (metallic oxides, peroxides, etc.) forming water vapor which is almost insoluble in both the liquid and the solid metal. If this reaction takes place in the hot solid metal.** internal formation of water vapor takes place with the formation of blisters. Guichard has studied this phenomenon in the laboratory; Artignan observed it in actual field conditions in the fire-box plates of locomotives. Andoyer has pointed out that the loss of weight occurring this way could be used as simple analytical tool, but it is of course easier to observe the change in mechanical strength upon bending.

We were the first ones to point out 40 years ago how oxidizing agents may be used to combat blowholes. We did this at the Paris Foundry Congress in May 1913 (A. Portevin, "The Role of Oxidants

^{*} Furthermore the discolved hydrogen causes blowholes all by itself: G. Lobley and D. Jepson (Rev. Met. Abstract, p. 409, 1926) fused and chilled copper in a hydrogen atmosphere between 1100 and 2200 C (2012 and 3992 F), they found that at solidification gas is discharged. The duration in seconds of this period and the volume per cent of blowholes in the solidified metal were at a minimum at 1350 C (2462 F) and at a maximum at 1750 C or 3150 F.

^{**} When hydrogen causes damage (by the formation of water), the defect looks like cracks, whereas CO causes blisters: the metal becomes spongy due to the development of CO₂, without rupture. The change of density due to the speed of diffusion is thus much greater with hydrogen and water than with CO and CO₂.

in the Copper Foundry Practice and Their Application," Rev. Met., vol. X, p. 944, 1913).

(2) Steel—The purifying or refining of pig iron is also done by oxidation, either with ore (open hearth, electric furnace), with gas (open hearth), or with an air blast (Bessemer converter); all this gives rise to FeO which is dissolved in the liquid steel. If one thereafter adds carburizing agents, like cast iron, spiegel etc., CO is formed. The development of CO also causes the elimination of hydrogen and nitrogen which were present in a state of supersaturation.*

If casting is done at this moment, an ingot with blowholes is obtained, and even outbursts of liquid metal occur. But we shall show that it is possible to regulate the gas development in the ingot mold so that the volume of blowholes compensates for the pipe ("open" or "rimmed" steels with greatly re-

duced crop ends to cut).

If one wants to diminish the formation of CO, which is the reaction gas, one cannot do it by the elimination of carbon. Carbon is a very important element in the steel, it is responsible for all the mechanical properties and for the tempering process. Therefore FeO is the compound worked on. It is acted upon by reducing agents or deoxidizers: Al, Si, Ti when reacting with FeO furnish Al₂O₃, SiO₂, TiO₂, and these cannot be reduced with carbon under the prevailing conditions. In other words, they do not act on CO. The FeO may also be removed by slags which are low in FeO and acids.

Pourcel in 1878 discovered the action of silicon. Starting with this discovery it was for the first time possible to obtain steel castings free from blowholes. We know now that aluminum is more efficient (as found by Brinell) and may be used both for steel

and cast iron.

There might be an occurrence of hydrogen also in the case of iron. However it can be completely avoided. It does damage (flakes) or at least does not serve any purpose. This hydrogen cannot be removed by oxidation in the liquid metal, as in copper, because at the temperature of molten steel (1600C or 2412F) the water vapor is almost 100 per cent dissociated.

Agitation, by means of a neutral gas effects the discharge of the unwanted gas in excess of its saturation, just as the discharge of reaction gas at the outset of solidification, eliminates that part of the dissolved gas which is in a state of supersaturation. Thus in copper the water vapor that is formed discharges the excess hydrogen simultaneously. In

* Gas liberation from heats, at the cooling of steel and cast iron has been studied by Muller (1883), Ruhfuss (1897), Munker 1904), Herwig (1913), Baradac-Muller (1916), Piwowarsky (1920), Karhrs (1925), Herasymenko and Dombrowski (1941), Hare, Peterson and Soler (1937). The developed gases are overwhelmingly carbon monoxide with some hydrogen and nitrogen and

steel the discharge of CO carries the hydrogen with it; thus the discharged gas contains both H and CO. Let us pause for a moment and remember the famous controversy between Pourcel and Miller over reaction CO gas against dissolved hydrogen gas. Each one was alleged to be responsible for the blowholes. Finally LeVerrier gave the correct answer as we just have indicated.

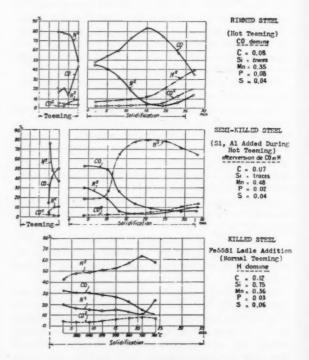
Thus it is to elimination effects that one must attribute the effect of coatings upon the blowholes of ingots and castings; viz, pitch decomposes and volatile matter evolves, and this gas causes the dis-

solved gas to blow out.

As the gas develops during the steel solidification (experiments made by Kahrs in the Krupp Works 1925, Fig. 10) one observes that first the hydrogen appears, to make place later on for CO. This is because the supersaturated gas, i.e. the gas not in equilibrium, is released first under the trigger action of the solidification, whereas the gas resulting from an equilibrium reaction is developed slowly as the equilibrium changes due to solidification.

Moreover in certain heats, especially of special steels, one cannot kill the casting by the action of the common deoxidizers Si and Al, and here hydrogen blowholes occur. It is necessary to destroy first the excess of reducing agent in the casting by oxidation, then one must carburize the metal, heat it until it sprcuts or evolves gas violently and then

finish as usual.



Basic open hearth, scrap used, teemed directly into ingots of 650 kg in an ingot mold which was covered with a lid carrying an iron pipe for the gas collection. (In order to measure nitrogen, the ingot mold was first filled with hydrogen).

with very little carbon dioxide and methane.

In timmed steel, carbon monoxide dominates, and hydrogen in killed steel. The maximum amount of both gases is reached at almost the same moment in both cases.

The time which passes between pouring in the ladle and casting in the mold does not influence the gas composition, at least not if this time varies from 8 to 10 min to 22 to 25 min.

Fig. 10—Composition of the gas discharged during teeming and solidification of steel (Kahrs, Krupp, 1925).

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Brinell has studied the most commonly used deoxidizers Mn, Si, Al; we shall talk about them further.

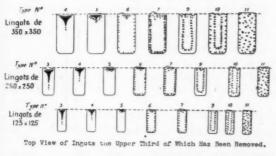
Aluminum is the most active deoxidizer. It is even suitable for the rapid and easy determination of dissolved oxygen in liquid steel. The right amount is found by adding aluminum in increasing amount to every ingot of one teeming operation, until one ingot will be completely killed, i.e., no gas evolves during the solidification (a deep pipe forms continuously under a charcoal layer).

In summary, the factors which influence the equilibrium of the reaction gases (such as the pressure, temperature, concentration, segregation at solidification) are the same which influence the equilibrium of the dissolved gases. Thus the solution for overcoming blowholes are the same, irrespective of the origin of the gas and the attempt to accentuate ability to form blowholes by avoiding conditions for violent gassing as explained below, will be equally applicable for both cases.

The gas discharge may operate in correlation with the non metallic inclusions which thus might become of tremendous importance. They might react chemically and thus produce a reaction gas. They act either per se or by aid of this reaction gas to initiate the release of the supersaturated gases (compare the release of steam from superheated water when nuclei are thrown in).

Conclusions from the Studies of Brinell, concerning Blowholes in Steel Ingots (Fig. 11).

According to Brinell the most important factors



bype 4. type 5 Types 6et 7 type 8 type 9 type 11

Classification according to compactness figures, which appear on top of every cross-section; appearance of the ingot with the upper third removed; each type is associated with a definite compactness coefficient.

For 250x250 ingots (walls 50 mm acid open hearth, hot teeming (P $= 0.02 \cdot 0.03\%$).

Secondary Factors: Amount of C and P, temperature of teeming, initial temperature of the mold, wall thickness of the mold, ingot cross-section, feeder.

Fig. 11—Blow holes in steel ingots (according to results of Brinell's experiments on exhibit in Paris, 1900).

are the percentages of Mn, Si and Al, because they influence the time at which the discharge of gas from the steel starts. After all the free surface has been covered with a solid crust the gas development must cease, and now a high gas pressure results from the discharge of gases in a closed space.

If gas is discharged violently from such an ingot, the crust cannot form, thus the gas discharge con-

For 250 x 250 mm ingots, wall thickness of the mold 50 mm, of soft steel from the acid open hearth (P = 0.02 - 0.03 per cent), the ingot type is de-

K = Mn + 5.2 Si + 90 Al

the values of K for the different types are

fined by the compactness coefficient K.

The type of compactness once established, according to Brinell, K has a value smaller than the one calculated from the amounts of Al, Si, Mn, if

(1)—pouring temperature is higher;

(2)—initial temperature of the mold is higher;

(3)—wall thickness of the mold is lower;

(4)—amount of P is higher

(5)—cross-section of the ingot is smaller

(6)—there is no refractory hot top (sinkhead).

The presence of a sinkhead does more damage than good, especially if ingots of the type 5, 6 or 9 are wanted.

According to a note published by the Gathmann Engineering Co., if at tapping a heat of killed steel the aluminum is added to the ladle when it is filled to one-third or up to two-thirds, one must wait about 8 min for a casting of 50 tons and 12 min for a casting of 100 tons, before one may cast in the molds.

In a rimming steel the ingot rises generally 75 mm to 125 mm in an ingot mold of 1.80 meters height.

In semi-killed steels (semi-deoxidized steels) the blowholes form especially at the upper outer part of the ingot. There is also a pipe which is not very voluminous but reaches very deep. One notes differences which can not only be attributed to the degree of deoxidation, but to the manner of pouring or to the temperature of both the heat and the ingot mold.

An ingot of steel which was killed in the ingot mold, after the teeming, may show simultaneously a secondary pipe of a semi-deoxidized steel and deep seated blowholes in the lower part of the ingot just like a rimmed steel; thus these two defects are not restricted to the classes of steel they are supposed to be characteristic of.

To summarize what we have said thus far there is an equilibrium in reactions between an oxidizing agent (Cu₂O, FeO) and a reducing agent (C, CO, H, S) said equilibrium being influenced by temperature, concentration and pressure. The discharge of gases may be prevented by two procedures:

The reducible oxide must be eliminated:
 (a) either by dissolving it in a flux or slag, which must be acid if the oxide is basic, and vice versa;

(b) or by the addition of a reducing substance which sees to it that the metal oxide is replaced by an irreducible oxide (in the common copper alloys, like brasses and bronzes, Zn and Sn are the reducing agents for Cu₂O).

(2) The reducing agent must be eliminated by addition of an oxidizer. This can only be done if the reducing agent is not needed in the final product. Thus carbides cannot be removed from steel.

One should keep in mind that these oxidations and reductions are governed by thermodynamic facts, like heats of reaction, changes of free energy, etc.

As in the open hearth processes, one may for the sake of greater efficiency and speedup use slags and

deoxidizers simultaneously.

The second process described above must be used, for copper and even for iron, if the reducing agent is a gas, like hydrogen. If hydrogen had been used in the first process and an excess thereof remained dissolved, blowholes result, as has been described above. This explains the occurrence of a metal full of blowholes if Fe is used to deoxidize copper. Blowholes also result if SO₂ is not removed completely.

If the reducing agent is sulphur, it may be removed with magnesium, cerium (Ce₂S₃), beryllium or lithium. As all of these are simultaneously deoxidizers, oxygen and sulphur are removed, thus SO₂ in solution is de-

stroved.*

As it is not possible to add exactly the correct amount of oxidizer, one uses an excess thereof and follows up with a deoxidizer to remove the remainder of the oxide.†

Or one may also deoxidize or desulphurize by a slag, i.e. with a phase outside the metal proper. Or one may fix both the oxygen and the sulphur into an inactive combination, i.e. one which is innocuous with respect to the properties sought. This is a neutralization and not an elimination.

Rimmed Steels

According to the amount of FeO which remains at the moment of teeming, a more or less open steel is obtained, which upon pouring into any one given ingot mold furnishes all types of ingots starting with a completely killed steel without blowholes and shrinkage holes (Type 1, Fig. 12) until a completely open ingot which creeps out of the ingot mold, forming the "leg of a boot" (Type 7).

If the steel is kept open, the pipe is suppressed, and lots of blowholes become evenly distributed through the mass. They become closed at the rolling operation; but still one must avoid overflowing and see to it that the gases escape at a certain time after the teeming of the ingot when a solidified layer without blowholes has been formed (compare with the series of

the Brinell types for the ingots from the basic Bessemer process). The blowholes which break through to the surface or which are cracked during the rolling do not weld together anymore due to oxidation (ingots with slivers), thus furnishing a metal which must be scrapped. Furthermore the upward motion resulting from the effervescence of the metal causes the formation of a cornice at the upper part of the walls which finally seals in the top of the ingot thus for forming the solidification to proceed in a closed space.

Therefore, in order to obtain this result, one needs in the ingot an amount of FeO which moves only within very narrow limits, and this correct amount is obtained by the judicious addition of aluminum in

small portions,

The last stage of the ingot "closing" described above is sometimes obtained by forming a "cold-top," i.e. either cold metal or a fine spray of water; but this is

a very dangerous undertaking.

In open steels it was found that the blowholes of the first outer zone contain hydrogen, the ones in the interior carbon monoxide. This might be due to the fact that carbon monoxide is an equilibrium reaction gas, which discharges slowly during solidification, as its concentration in the liquid increases, whereas hydrogen is released immediately at the first agitation from a supersaturated solution, due to the discharge of CO or due to solidification.

If one lets the solidification take place under pressure by closing the ingot mold, as was done by Hultgren, Phragmen, Wohlfahrt and Ostberg (Jour. of

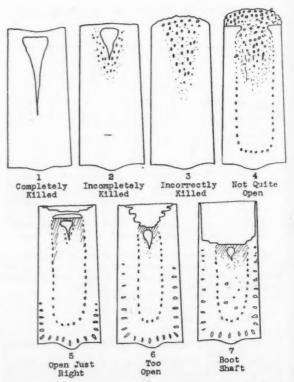


Fig. 12:-Types of ingots from the basic Bessemer. (Pipes and blow holes, killed steels and rimmed steels). 3-ton ingots 500x500.

^{*}One should here neither use the expression deoxidation nor desulphurization, but talk about a fixation of oxygen and sulphur into an inactive combination. The word inactive refers to the chemical as well as the thermal properties. One should keep in mind that Mg, Ce, Be, Li are added to pig iron in order to obtain spheroidal graphite, but the action of these elements in the presence of both sulphur and oxygen warrants new studies.

[†]Thus e.g. with cupronickels (Land and Rolfe, 1939) one adds first an oxidizer, like copper oxide, nickel oxide, manganese dioxide, afterwards magnesium is added to remove and sulphur and finally silicon as a deoxidizer.

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Metals, Feb. 1951), then, if one keeps in mind that there is such a thing as an equilibrium pressure (which is a function of the equilibrium, position of the blowholes, the structure), it should be possible to obtain at this pressure a steel which solidifies just like a killed steel. And this was indeed what was obtained.

(C) Gas in Suspension

Either gas bubbles, the ones from the dissolved gas or the ones from the reaction gas, rise to the surface with a speed which is governed by Stokes' Law. They are discharged if the progressing solidification does not entrap them, and this takes place at the second state of solidification. In addition to this, one has in the light alloys the phenomenon that the escape of the hydrogen bubbles is inhibited by "films" of alumina which float in the metal.*

Masing now found that a treatment of Y-alloy with ZnCl₂ is effective (0.5 per cent of it), whereas MgCl₂, CaCl₂ and NH₄Cl show no effects. If ZnCl₂ is thrown in as one big lump, a rapid reaction occurs with lots of gas bubbles rising, which explode upon contact with air. A dry skin is formed, consisting of AlCl₃ and ZnO, which is easy to skin off.

ESTIMATION OF THE BLOWHOLINESS (Ability to get blowholes).

EXPERIMENT OF SOLIDIFICATION IN VACUO OR SPROUTING (INCUBATION) IN VACUO**

Advantages and Critique—To find out how much a given metal is inclined to form blowholes one must produce, in agreement with the principles which we have stressed repeatedly (e.g. Fonderie, Feb. 1951, no. 62, p. 2347) the causes in an exaggerated way. In the case on hand we exaggerate the conditions by lowering the external pressure—and one may simultaneously make use of an agitation by ultrasonic or an electromagnetic means. This causes gas development. The amount of gas developed increases, being always proportional to \sqrt{p} . Its volume increases too, being inversely proportional to the pressure (Mariotte's Law).

To do this one pours a sample of the liquid metal into a crucible which is placed into an inclosed space which is now evacuated (this accounts for the name: bell experiment); Thus one has simultaneously lowered the pressure and cooled to metal solidification. The result thereof is shown graphically in Fig. 13.

If no gas develops, one obtains a little ingot, showing an external recession due to a funnel-shaped pipe. If gas develops (sprouting) the metallic mass will show blowholes and a more or less blisterly surface. The result is thus estimated by examining the total free surface and center cross-section (Fig. 14 and 15).

This essentially qualitative result may be expressed in a quantitative way if the compactness index is derived from the apparent density (A. Portevin, Bull. A.T.F., vol. XI, p. 218, June 1937). A semi-quantitative evaluation is obtained if the size of the blowholes in the central cross-section is measured with a planimeter.

One thus may in addition to the two criteria (appearance of the surface and examination of a cross-section) determine the apparent density, especially if the gas evolution is very weak.

Applications of this work have been thoroughly

studied (Fig. 14 and 15).

This test is very simple and fast, and is well related to the phenomenon we are after. It shows us the defects i.e. blowholes which will result.

There is some adverse criticism also, dealing with its value as a research tool:

- (1) A minor criticism is the one which objects that we have here simultaneously three kinetics: the kinetics of gas development, the kinetics of cooling and the kinetics of expansion.
- (2) The temperature of the metal worked varies from surface to center and furthermore, the solidification progresses only from the walls towards the center with a simultaneous gaseous segregation.

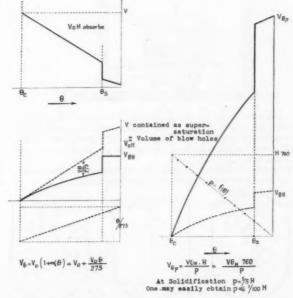


Fig. 13—Curves showing instantaneous volume of blow holes as a function of θ for the case of complete saturation in θ_c with two assumptions (1) external pressure of H_2 constant and (2) external pressure diminishing at cooling according to p: $f(\theta)$. At solidification there is

$$V = \left[k(\theta_c - \theta) + V_s \frac{(1 + \alpha \theta) H}{760}\right]$$

NOTE: As the velocity with which the bubbles rise depends upon Stoke's Law, and thus upon the temperature, the test which the distillers give to brandy for testing the alcohol proof (how long it takes shaken-in air bubbles to rise in the bottle) is of course nonsensical.

^{*} That these "films" really exist can be shown with X-ray pictures of aluminum alloys (thin slices thereof) containing lead globules. The alumina catches the lead globules like a net catches fish, and from the arrangement of the lead globules we conclude the presence of such an alumina net. The lead here of course takes the place of the hydrogen bubbles.

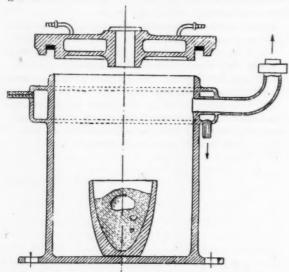
^{**} This experiment has sometimes been referred to, in the name of the apparatus used (bell experiment) and sometimes with the name of the inventor (Pfeiffer's experiment).

(3) Furthermore the phenomenon might have to run a competition. If there are gaseous nuclei at the walls which were placed there by the steel when it was poured into the crucible, an objection voiced especially by Moreau, (Bull. A.T.F., p. 446, 1936), that accelerates the gas development up to sprouting.

It seems that the study of this phenomenon is related again to the processes used for the study of other systems out of equilibrium, especially of the quenched

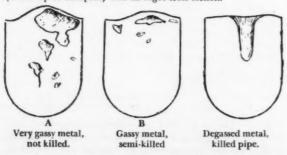
states of steel:

(a) The development of gas under isothermal conditions (crucible electrically heated) as a function of decreasing pressure. Y. Dardel devised this method to determine the pressure when the first bubbles would appear, at the surface of the metal, and thus he would draw conclusions about the amount of hydrogen contained in the aluminum. Dardel claims that there is no retardation of the gas development, as the small amount of alumina operates to initiate reaction and that the surface layer of alumina is impervious to the gas.



Example of the Moreau apparatus. The crucible is placed in a receptacle with lid and rubber gaskets; it is water cooled. The observations are down through a pyrex opening in the lid. The vacuum might be zero, i.e., no vacuum, in the case of certain stee.s.

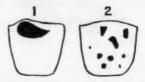
One observes simultaneously the appearance of the surface (blown up or collapsed) and an ingot cross-section.



Solidification under reduced pressure or sprouting experiment in vacuo. The pressure drop causes a gas discharge on account of the pressure drop proper and because of the agitation connected therewith; also the volume of blow holes becomes larger.

Fig. 14-Measurement of the ability to form blowholes.

To appreciate how much gas was developed one may photograph the ingot cross-section and measure the holes with a planimeter and thus determine the per cent of the area covered with gas. But one must also take into consideration the distribution of the cavities. Thus from the two shown, ingot 1 is acceptable and 2 is not, the area covered with gas being equal.



COMPARISON OF DIFFERENT METHODS OF PURIFICA-TION AND DEGASSING AT ALUMINUM FUSION (AFFER J. CZOCHRALSKI AND K. MIGURSKA, POLISH FOUNDRY CONGRESS, 1938).

Treatment	Porosity, % Bell Test		Tensile Strength, R _t Kg/mm	tion
Without Purification	18.1	2.656	12.5	1.1
Slow Presolidification	14.6	2.743	13.7	1.2
Kept Liquid for a While		2.747	10.7	1.3
Kept Liquid for a While				
(With Flux)	6.9	2.805	16.5	2.3
Fusion with Flux*	2.2	2.831	17.1	3.0
Chlorine Treatment	0	2.838	16.9	3.1
 Fluxes were mixtures Na. Analysis: Cl, 32; F, 1 				

Fig. 15-Blow holes. Bell experiment.

(b) Study of the gas development under isobaric conditions has been made at decreasing temperatures, namely the influence of the speed of cooling upon the delay of the gas discharge.

These studies would be achieved by an examination of the role of stirring and of the role of pre-existing gaseous nuclei. This might conceivably lead to a study of the relaxation times under isotherm and isobar conditions; to the study of nuclei and the surface tension.

As a special case, we have the case where the outer pressure is atmospheric, and the discharge of gas at solidification, no matter if reaction gas or dissolved gas, is sufficient to produce violent gassing or at least is clearly visible. Therefore it is used in the control tests for the killing of steel (of the Brinell types), it is used in the foundry before teeming (an examination of the free surface suffices). The same applies to the test by Gurtler for alpax. He prepares a test casting which has the form of a disk in a thick electrolytic carbon mold, which is heated to 100 to 200 C (212 to 392 F) (Fig. 16 and 16a). The test has even been used for aluminum by observing the surface of a specimen similar in shape to the one used by Gurtler. The diameter of the blistered part of surface is measured. The less the diameter, the less gas there is as the vigorous gas development starts only when the pressure of gas in the central liquid region reaches the atmospheric pressure, during solidification.

General Character of Blowholes

(1) Blowholes of external origin (exogenous blowholes), appear at oxidized surfaces, and are sometimes of very irregular shape, because the gas had to fight its way through the metal in course of solidification.

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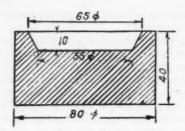
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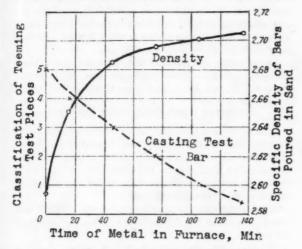
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TEST PIECES FROM TEEMING

In the mold shown consisting of electrode carbon, heated to 100-200 C, one teems about 100 g of metal; the surface is observed after solidification. Small bubbles, i.e. the discharged gases, lift up the thin oxide layer at the surface and give rise to growths giving the impression of craters the size of pin heads. According to the number thereof one evaluated the gas content evolved by assigning a value of 1 to 5.



Left to cool at 750 C in a neutral atmosphere, the Al containing H loses it slowly giving the broken curve shown; it is in agreement with the logarithmic growth of the compactness derived from the apparent density from the Boulon test sample:

Relation between test piece classification and specific gravity. The Boulon Test Piece has a cylindrical shape, is teemed horizontally and carries a feeder at one end, large or small. A round piece is cut out from a spot in one-third of the length of the sample and x-ray inspected. Results are given the number 1 to 5. With but little gas, the feeders show large blowholes; for much gas they show very fine blow holes in great number.

Fig. 16—Test pieces used for Alpax (G. Turtler, 1938).

One pays attention in this case to slivers, laminations, etc. (in the casting there are scabs); was it a big-end-up or a big-end-down mold, etc.

(2) Naturally occurring blowholes (endogenous ones): The walls of the ingot are as a rule shiny, they look satisfactory only sometimes there is blistering at the surface. These blowholes are sometimes arranged in a regular pattern. They are round or egg-shaped or even like a flattened out pear (Fig. 17) in the peripheral regions of the parallel crystallization. The larger blowholes have occasionally dendritic outcrops on their surface.

While the shape of the pipe may be deduced from the form and the distribution of the isotherms together with the action of gravity (their position agrees with the last zone of solidification R, Fig. 18), the form of the blowholes depends upon the surface tension,

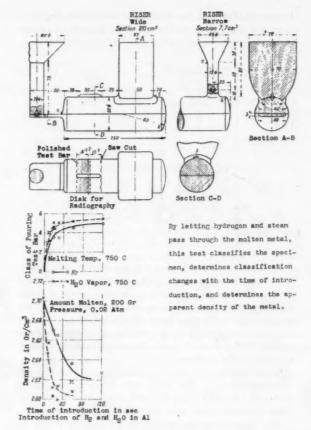


Fig. 16A

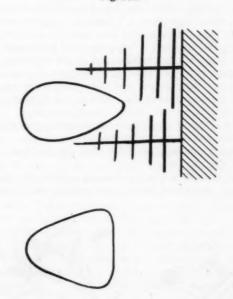


Fig. 17

because these are gas bubbles with a total curvature of 1/R + 1/R' which is constant. The distribution of the blowholes is in relation to the shape of the iso-

[•] In a free gas bubble, the pressure $p \equiv A (1/R + 1/R^2)$, i.e. the total curvature is a constant.

therms (S, Fig. 18) because their formation depends upon the saturation of the residual liquid, i.e. upon the segregation phenomena (isochemical surfaces).

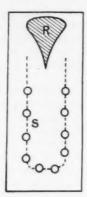


Fig. 18

In the case of a cylinder we can always clearly differentiate between pipe and blowholes; but should the gas discharge take place at the end of the solidification period, the blowholes and the pipe would coincide. For both pipe and blowholes one must say that they are always somewhat out of shape, because crystallization at their walls always gives them a somewhat irregular outline.

Now talking of microblowholes and micropipes (microshrinkages), the difference becomes still more obscured. But at least for cast steel one can distinguish between these two kinds of cavities because:

(a) the micropipes have a very irregular shape; in their inside one finds dendritic outcrops;

(b) the microblowholes are of a somewhat rounded form; sometimes there is a striation at their walls.

Generally they are both interdendritic or intercrystalline, the micropipes being enclosed by crystals and the microblowholes, being round at the beginning and becoming deformed with the crystalline solidification. Therefore their shapes and distributions are similar. Ignoring the origin of these defects, it is wiser to apply a name which covers both: namely porosity.

The shape of the blowholes depends upon the gas content and upon the solidification structure.

Thus according to Gurtler 1938, for the alpax alloy (Fig. 19), one finds:

 (a) if there are many primary crystals, the jagged cavities are located between the primary dendrites;

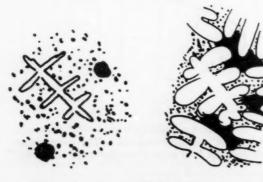


Fig. 19

(b) if there is quite an amount of eutectic, the blowholes are spherical.

At high gas contents, the gas pressure becomes the dominant factor, round pits are in this case also

formed in hypo-eutectic alloys.

In aluminum alloys the pits of gas-rich alloys are always rounder than the ones of low gas content. It is probable that there is a series of intermediate steps starting with a micropipe and ending up with an interdendritic blowhole, if the amount of gas in the liquid metal rises or if the pressure during the solidification drops. The empty micropipe draws the gas in, if there is just a little of it. If there is much gas, bubbles form of a more rounded shape, and they rise to the top before the micropipes are even created*

The empty micropipes appeal to the oxidizing gases (or to sulphurizing ones, like reaction SO₂), and the microcavity is covered by an oxide film which in the case of copper alloys furnishes a darkened color which appears in the fracture (that is what Bolton and Veigand call "incipient shrinkage").

In each case the porosity is apparent from a determination of the apparent density (see A. Portevin, Bull. A.T.F., vol. XI, p. 218, June 1937).

Eliminating Naturally-Occurring Blowholes

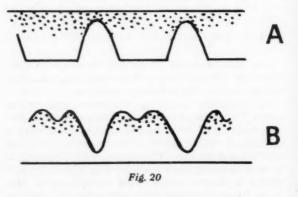
First of all it is necessary to avoid any absorption of gas after the metal has once been heated to the liquid state, and this depends greatly on the nature of the gas (mostly CO, CO₂, O, N, H₂O, H, CH₄, C_nH_m, CO₂) and upon the temperature of the metal, upon the surface, and the length of the contact.

If we scrutinize the gases with which a metal may come in contact at the ordinary industrial smelters, we find:

(a) every metal absorbs hydrogen, which gives rise to blowholes due to dissolved gas;

(b) oxygen and the oxidized gases (CO₂, CO, SO₂) furnish oxides, and simultaneously they furnish sulphides and carbides; thus these blowholes are always due to reaction gases;

(c) nitrogen is not absorbed by copper nor by alum-



^{*} In the subdivided aluminum ingots, which are obtained by pouring in chills, the pits and the pipes are always in the upper part (Fig. 20). Thus if pouring is done in the "normal" position, the constrictions along which the ingot shall be broken, are already preweakened, as compared to those of ingots cast in the B position, where the constructions contain good metal.

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inum at its melting point. Thus it may be used for degassing by flushing.

Therefore Berbero (Milano Foundry Congress, 1931) recommended a limit of 820 C (1475 F) when melting aluminum and the avoidance of moisture in the heating gases (the example of aluminum was cited by Pontremoli, *ibid*)

It is especially important to avoid melting pieces of metal by means of a current of combustion gas going right through the charge.* This happens sometimes in crucible melting of copper alloys if one uses a sink-head on top of the crucible; the foundryman fills both the crucible and the sinkhead at the same time, the liquid metal runs from the sinkhead into the crucibles and comes thus in contact with the hot gases. The correct way to operate is first to melt the metal in the crucible, then push the metal from the sinkhead into the crucible as soon as it is red.

The permeability of graphite crucibles permits gases to pass through in a highly superheated state.

Finally the amount of gas picked up depends upon the length of time of the metal-atmosphere contact. This is true both for dissolved gases and for reaction gases like sulphur and oxygen compounds. Thus the speed of melting plays a role. Of course one always wishes to pour the liquid metal as soon as possible (see the example for bronzes by A. Brizon, below).

The gas-metal contact may be prevented by the use of fluxes which must have an appropriate melting range and viscosity.

A. Brizon (Bull, A.T.F., Jan. 1934, p. 7) has given a practical example which shows clearly the influence of this factor for bronzes cast in green sand and molten in crucibles of 75 kg. The crucibles were placed in a furnace under draught which was heated with coke. If the melting required over 40 to 45 min, then 80 to 95 per cent of the ingots were rejected on account of blowholes discovered when machining. Almost all of these were in the upper part of the castings. By reducing the melting time to 25 min and by pouring at a lower temperature, the defect was completely eliminated. Thus it is obvious that the melting time must always be controlled.

We can say that gas absorption by metals increases the rate of defect occurrences; because the more pieces that are defective, the more rejects must be recast. Thus the smooth operation of a foundry is interrupted.

In addition to examining all the foundry factors (a general examination), we must make a special mention of:

(1) Additions Which Act Either Upon the Dissolved Gases or Upon Gas-Producing Reactions.

(a) action of deoxidizers (Si, Al, Ti) prohibiting the evolution of gas from steels.

(b) additions which diminish the solubility or the absorption of the dissolved gas; that is why Mg, Zn and Ca is added to aluminum (Mg reduces CO and CO₂ at the melting point), and why Ca is added to magnesium (nitrite formation is avoided).

In the case of the light alloys, their porosity, the size and shape of pits depend upon the alloy; therefore Budgen has arranged the alloys in a series, the porosity decreases in the series:

(1) Y-alloy (4 Cu, 2 Ni, 1.5 Mg); (2) 12 and 8 Cu; (3) 13 Zn + 3 Cu; (4) 9 and 14 Si; (5) pure Al.

(c) direct chemical action upon the dissolved gases. One should cite here the effect of oxidants (lead oxides) upon bronzes molten in reducing atmospheres. The Tullis process (see Rev. Met. XXVI a, 416, Sept. 1929) devised for the elimination of hydrogen by chlorine; chlorine is passed for a few minutes through molten aluminum alloys.*

Efforts have been made to replace chlorine by dissociable chlorides like ferric chloride, tin tetrachloride, titanium chloride, silicon tetrachloride, carbon tetrachloride, and boron trichloride. The boron and titanium chlorides refine the grain of the metal (Rosenhain, Grogan and Schofied, Institute of Metals, Sept. 1930).

(2) Physical Processes

(a) Degassing of molten metal by mixing or stirring; for example: mechanical agitation of magnesium molten under hydrogen; ladles and crucibles shaken with ultrasonics and with rotating motion; flushing with an inert gas of low solubility like nitrogen in the case of copper and in the case of aluminum which has absorbed hydrogen (the nitrogen is almost insoluble in the aluminum and drives out the hydrogen which burns off at the surface of the molten metal) or like argon in the case of all the metals. It is obvious that the flushing gas must be completely dry, otherwise one simply adds fresh hydrogen to the solution

(b) Degassing by slow cooling prior to pouring and even during solidification; for example: aluminum absorbs gas when superheated (1000 to 1500 C or 1830 to 2730 F) to be able to decant sodium and get rid of gas inclusions from the electrolysis; if one cools slowly one notes at about 900 C or 1650 F a vivid gas development which causes the metal to foam; when the temperature of pouring is reached there will be no blowholes. Other examples: Degassing of copper containing hydrogen and sulphur dioxide by slow solidification (Pryterch, Inst. Metals, March 1930);

^{*} In the case of aluminum one must avoid both hydrogen and water vapor; therefore one uses fuels which do not produce these gases. One also separates the combustion unit from the melting unit proper. But thus of course the thermal efficiency is reduced, therefore an attempt was made to reduce the solubility of the gas in the metal by keeping the latter at a lower temperature (750 C, 1380 F).

Metal from an electric furnace is much better, it contains less

Therefore in steel making one must take certain precautions such as drying the charges (also the hearth must be dry, as well as any added slags or other additions). One must watch carefully any cooling unit using water circulation. It is of special importance that limestone be replaced by flux, and that the ferrosilicon is heated prior to use.

^{*} If degassing is done in aluminum by flushing, N and Cl, both act in the same way. They reduce the partial pressure of the hydrogen. One uses the gases which are most available. We might as well add that chlorine furnishes gaseous AlCla which acts in the same way.

The action of zinc chloride and similar fluxing compounds must be explained as lowering the apparent viscosity of the metal, as Al₂O₄ is dissolved. Thus the mechanical discharge of the suspended gases is facilitated.

Archbutt has devised a method of slow solidification applied to aluminum. The solidification takes place in crucibles, right after the solidification a rapid refusion takes place and the metal is teemed, thus it has no time to reabsorb too many detrimental gases.

Presolidification in crucibles has also been applied to bronze; thus an improvement of the elongation has been obtained without increase in density.

This explains why the second fusion may be done without superheating.

(c) Finally fusion in vacuo and solidification under

pressure.

Fusion in high vacuum (1/100-1/1000 mm) at 1600 C (2910 F) results, as Hochmann has shown for the case of high chromium steels (Thesis, Paris, 1950) in a notable reduction of the carbon content (0.03-0.003), of the oxygen (0.05 to 0.002), of the nitrogen (0.05 to 0.005), but the hydrogen does not increase. Even on a laboratory scale, this physical process does not get rid of the hydrogen, but gives a good deoxidation (melting in the high vacuum in presence of carbides is the method for oxygen determination in steel and cast irons hence of total elimination of oxygen).

(d) The speed of solidification opposes the discharge of gases and the formation of blowholes (not the difference of pin holes in aluminum teemed in

chills and in sand).

Blowholes may be mistaken for micropipes; both may occur side by side and success of a combating agent gives information on the reason for the defect. For example, aluminum and its alloys show pin holes, i.e., small cavities of varying shapes and dimensions, of 0.4 to 3 mm and more in diameter. One encounters these especially in the thick parts of castings made

in sand, they hardly occur in pieces teemed in chills.

Gases play a role in their production; this becomes obvious if teeming is done in vacuo: the gas bubbles now take up more space, and they disappear if solidification is accomplished under elevated pressure. But the pipe plays a role too, because Budgen points out that the fewer pinholes an ingot shows (vacuum treatment, teeming in chills, teeming at low temperatures), the more the surface pipe is pronounced.

Influence of Metal Impurities

Thus impurities in the metal may influence the formation of blowholes, and there are three kinds of action to be distinguished:

- (1) The dissolved impurities may change the gas solubility;
- (2) The soluble impurities may extend considerably the solidification interval (in commercially pure aluminum the silicon and iron present as impurities cause already the binary and ternary eutectic, the gaseous impurities segregate and collect in the portions which remain liquid longest).

(3) The non-dissolved impurities or inclusions may act as nuclei for the formation of blowholes, i.e. gas bubbles are thus formed in a liquid which is already

supersaturated with respect to gas.

In closing we wish to point out that the facts given and the principles enumerated here are known to the steel producer and to the metallurgist, especially those working in the fields of iron and copper. But the foundryman lacks this information, and therefore we have presented these things here in the hope they might be helpful.

A STUDY OF THE PRINCIPLES OF GATING AS APPLIED TO SPRUE-BASE DESIGN

By

K. Grube, J. G. Kura, and J. H. Jackson*

Introduction

This is the third of a series of papers describing the results of an extensive research program on the principles of gating. The research was sponsored at Battelle Memorial Institute by the Research Committee of the Aluminum and Magnesium Division of the American Foundrymen's Society. Battelle contributed some financial assistance to the program.

Two previous papers^{1,2} contained the results of research on pouring methods, design of the pouring basin, sprue, runners, and gates. In the second paper,² an optimum design of sprue and gating system was described. It was indicated, however, that these gating systems employing a wedge in the runner at the base of the sprue or restrictions in the runners at each side of the sprue operated favorably only with certain specific geometries of runners and gates. This led to the present work which was a study to develop a gating system applicable to a wide range of runner and gate geometries.

As in the previous studies, this investigation was conducted by pouring water in a transparent mold and recording the flow by means of high-speed photography. Entrainment of air in the water was indicative of dross formation during the pouring of molten metal.

In the present work, emphasis was placed on the investigation of the design of the sprue base and the effect of various gating ratios on casting quality. The results described herein were also used to prepare a color motion-picture film which was exhibited at the 1952 A.F.S. Convention.

Experimental Method

A drawing of a Lucite model prepared for the study of the sprue-base problem is shown in Fig. 1. The model represents an end section of a mold which contained the sprue and runner. For experimental purposes, it was termed an end runner and consisted of a cope section and a drag section. The cope was machined to accommodate any one of four sprues that were used in the investigation. Gate cavities were machined in the ends of the cope section.

The drag portion of the model contained a runner cavity 2 in. wide and 2 in. deep. The central portion of the runner was machined to $2\frac{1}{2}$ in. in diameter. Because the sprue opening was centrally located in the cope to provide a two-runner system, blocking off one-half of the runner cavity at the sprue permitted the investigation of a one-runner system with the same model. The major emphasis, however, was placed on the two-runner system.

Using various Lucite inserts, 17 different geometric combinations were available in the runner. Four round, tapered sprues having a bottom cross-sectional area of $\frac{3}{8}$, $\frac{1}{2}$, $\frac{3}{4}$, and $\frac{11}{8}$ sq. in. were employed. The condition of the runner under the sprue could

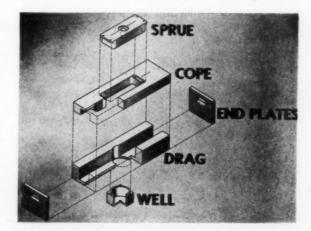
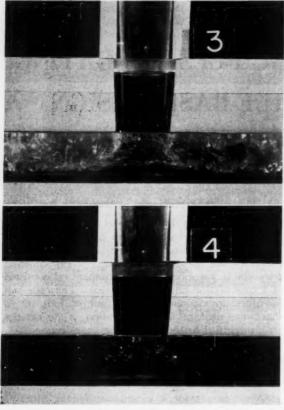


Fig. 1-Exploded drawing of Lucite sprue-base model.

This is the third progress report on a research project conducted at Battelle Memorial Institute, initiated and sponsored by the Aluminum and Magnesium Division of the A.F.S. with some financial assistance of Battelle Memorial Institute. The project was under the general guidance of the Research Committee of the Aluminum and Magnesium Division which consisted of the following members: R. F. Thomson, Chairman; W. Bonsack, Co-Chairman; W. J. Klayer, C. E. Nelson, T. D. Stay and W. E. Sicha.

Research Engineer, Assistant Supervisor, and Supervisor, respectively, Battelle Memorial Institute, Columbus, Ohio.

After 1 sec-Note severe turbulence.



After 15 sec .- Note continuous aspiration of air bubbles.

Fig. 3—Turbulence and aspiration in inverted-T sprue base with a wide, shallow runner. (Runner 1½ in. wide by 1 in deep. Gating ratio 1:4:4.)

be changed to produce enlargements of different diameters, and wells of various depths could be formed directly beneath the sprue. Although a wide range of gating ratios was available with this model, the major portion of the investigation was confined to sprue-to-runner-to-gate ratios within the limits of 1:2:2 and 1:6:6.

End plates containing an opening to simulate the cross-section of the gates were attached to ends of the assembled model. The opening in the end plates was located above the parting line. Thus, the gate in the cope was fed by the runner in the drag. To study the effect of back-pressure on the condition of the

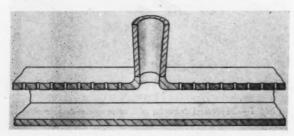
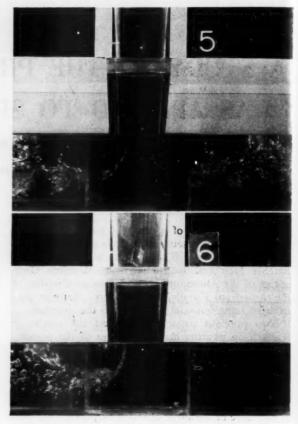


Fig. 2-Drawing illustrating inverted-T sprue-base design.

After 1 sec-Note severe turbulence.



After 15 scc—Note localized turbulence. Aspiration was intermittent.

Fig. 4—Turbulence in inverted-T sprue-base with narrow, deep runner. (Runner 1 in. wide by 1½ in. deep. Gating ratio 1:4:4.)

runner, the end plates were eliminated and in their place drain pipes turned upward were employed. The height of the discharge pipe controlled the degree of back-pressure on the system.

Sprue-Base Designs in Two-Runner System

The Inverted "T"—The inverted "T" design of sprue base illustrated in Fig. 2 was investigated for all of the runner geometries. This is a common type used in the foundry and is easily made with the ordinary type of sprue cutter.

In the laboratory tests, aspiration of air through the cope side of the runners was very severe with this

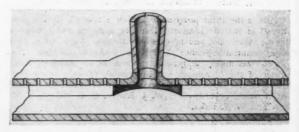
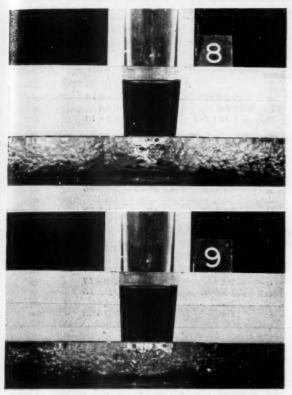


Fig. 5-Drawing of enlargement type of sprue-base design.

After 1 sec_Note severe turbulence.



After 5 sec-Note severe turbulence is still present.

Fig. 6—Turbulence at enlargement of a wide, shallow runner. (Runner 1½ in. wide by 1 in. deep. Gating ratio 1:4:4.)

sprue-base design. This was attributed partially to the fact that the liquid emerging from the sprue entered the flow channel at its highest velocity. In 36 of 45 runner and sprue combinations investigated, aspiration and turbulence occurred with this spruebase design. In the other nine combinations, aspiration did not occur, but long periods of time were

[•] The action pictures were taken by means of a Kenny Speedlight and a Speed Graphic camera. Exposures of 1/6000 of a second were made after the lapse of one or more time intervals measured from the beginning of the pour.

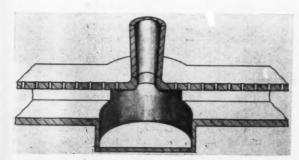
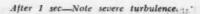
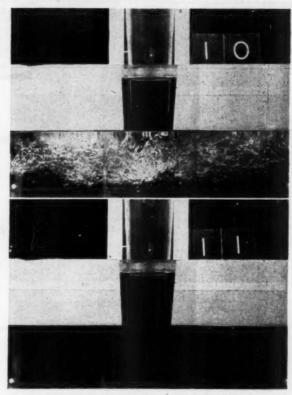


Fig. 8-Drawing of well-type of sprue base design.

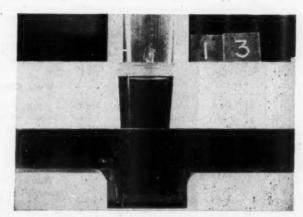




After 5 sec-Turbulence eliminated.

Fig. 7—Turbulence at enlargement of narrow, deep runner. (Runner 1 in. wide by 1½ in. deep. Gating ratio 1:4:4.)

required to wash the entrapped air out of the system. The data for these tests are presented in Table 1. The action in the runner,* 1 and 15 sec after the beginning of the pour, is illustrated by Figs. 3 and 4. A wide, shallow runner was used in Fig. 3, and a narrow, deep runner was employed in Fig. 4.



After 1 sec-No turbulence or aspiration.

Fig. 9—Elimination of turbulence by use of a well in a wide, shallow runner. (Well ¾ in. deep, 2 in. diam; runner 1½ in. wide, 1 in. deep. Gating ratio 1:4:4.)

TABLE 1—EFFECT OF INVERTED "T" SPRUE-BASE DESIGN ON CLEAN-UP TIME IN A TWO-RUNNER SYSTEM

		3/8-Sq-I	n. Sprue	1/2-Sq-Ir	n. Sprue	3/4-Sq-In	. Sprue	11/6-Sq I	n. Sprue
	nner sions, In. Depth,	Gating Ratio	Clean- Up Time, Sec	Gating Ratio	Clean- Up Time, Sec	Gating Ratio	Clean- Up Time, Sec	Gating Ratio	Clean Up Time Sec
				Square	Runner				
3/4	3/4	1:3:3	00	_	_	_	-	-	-
1	1	1:5.3:5.3	90	1:4:4	00	1:2.7:2.7	00	1:1.8:1.8	00
11/2	11/2	\ _	amus .	1:9:9	75	1:6:6	00	1:4:4	00
2	2	_	******	1:16:16	35	1:10.7:10.7	00	1:7.1:7.1	00
				Wide, Shall	low Runne	•			
1	3/4	1:4:4	10	1:3:3	00	1:2:2	00	1:1.3:1.3	00
11/6	3/4	-	-	1:4:4	55.5	-	_	-	-
11/6	1	_		1:6:6	91	1:4:4	00	1:2.7:2.7	00
2	1	-	_	1:8:8	00	1:5.3:5.3	63	1:3.5:3.5	00
2	11/8		-		-	_	_	1:4:4	00
11/2 2 2 2 2	11/2	_	_	1:12:12	151	1:8:8	56	1:5.3:5.3	00
				Narrow, De	ep Runner				
3/4	1	1:4:4	00	1:3:3	00	1:2:2	00	1:1.3:1.3	3
3/4	11/3	-	-	1:4:4	57	_	-	-	_
3/4 3/4 3/4	2	_		1:6:6	00	1:4:4	00	1:2.7:2.7	90
1	11/2	1:8:8	00	1:6:6	00	1:4:4	90	1:2.7:2.7	00
1	2	_	_	1:8:8	00	1:5.3:5.3	00	1:3.5:3.5	00
11/8	2 2 2	-	_	_		-	_	1:4:4	00
11/2	2	-	_	1:12:12	00	1:8:8	00	1:5.3:5.3	00

The Enlargement
The second sprue-base design investigated was termed the enlargement. As illustrated in Fig. 5, the enlargement consisted merely of an expansion in the

area of the runner directly under the sprue.

The earlier work2 on the Lucite molds indicated that this sprue-base design was beneficial in reducing the amount of turbulence in the system. Conse-

TABLE 2—EFFECT OF ENLARGEMENT SPRUE-BASE DESIGN ON CLEAN-UP TIME IN A TWO-RUNNER SYSTEM

			3/8-5	iq-In. S	prue			1/2-	Sq-In.	Sprue			3/4-	Sq-I	n. S	prue		11/8-S	q-In.	Spru	e
							Gating t Ratio					Gating Ratio 1				ime,		Gating Ratio	Ti Di		
Width	Depth		l In.	11/2 In.	2 In.	21/2 In.		l In.	11/2 In.	2 In.	21/2 In.			I In.	11/2 In.	2 In.	21/2 In.		11/2 In.	2 In.	21/2 In.
	1						***		Sq	uare'.	Runne	r									
3/4 1 11/2 2	3/4 1 11/2 2	1:3:3	80 — —	10.0	22.2		1:4:4 1:9:9 1:16:16	_ _ _ _	6.7 — — Vide,	22.3 — Shall	6.5 92.0 64.7 ow Ru	1:2.7:2. 1:6:6 1:10.7:10 nner				- 8 -	∞ 28.3 30.0	1:1.8:1.8 1:4:4 1:7.1:7.1		- 8 -	8 8 8
11/8	3/4		_	-	-	-	1:4:4 1:6:6 1:8:8	_		79.0 ∞	00 00 00	1:4:4		_	=		80	1:2.7:2.7	=	_	8 8
2 2	11/6	=	_	=	_	=	1:12:12	_	=	_	86.0	1:8:8		=	_	_	72.0	1:4:4 1:5.3:5.3	=	_	73.1 43.3
								Λ	Varrov	v, De	ep Ru	nner									
3/4 3/4 3/4 1 1 11/6 11/2	1 11/3 2 11/2 2 2	1:4:4	8	10.4	2.6	37.5 — — — —	1:3:3 1:4:4 1:6:6 1:6:6 1:8:8	8.6 ∞ ∞	5.1 4.0 ∞ 11.0	1.8 5.5 3.7 19.0	3.5 2.8 4.6 6.4 9.0 —	1:2:2 1:4:4 1:4:4 1:5.3:5.3		× -	8 8 8	2.2	3.0 5.5 5.6 45.0	1:1.3:1.3 1:2.7:2.7 1:2.7:2.7 1:3.5:3.5 1:4:4 1:5.3:5.3	8 8	1 8 8 8	14.0 1.8 4.0 3.8 ∞

quently, it was desirable to investigate this design with a wide range of runner dimensions and shapes.

Action pictures, taken at two different time intervals in both the wide, shallow runner and the narrow, deep runner, are illustrated in Figs. 6 and 7. The improvement effected by this design over the inverted "T" design is evident by comparing this group of figures with Figs. 3 and 4.

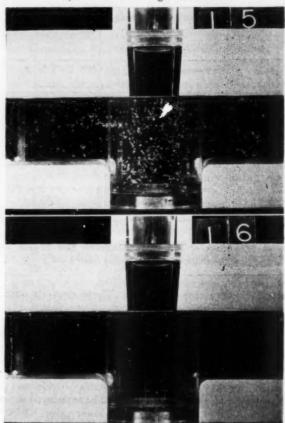
An attempt was made to correlate the diameter of the enlargement with the dimensions of the sprue and runner. For this purpose, those systems which produced a satisfactorily short clean-up time were tabulated. The trend indicated by these data in Table 2 was that an enlargement with a diameter approximately $2\frac{1}{2}$ times the width of the runner produced the shortest clean-up time in a narrow, deep runner. The sprue size did not affect this trend.

With the enlargement, the wide, shallow runner was more prone to aspirate than the narrow, deep runner. In a large number of tests, the enlargement appeared to be an effective design for reducing turbulence under the sprue and for preventing aspiration of air or mold gases. Its effectiveness, however, was restricted to a narrow, deep runner or a square runner. Because it was not universally applicable to all runner shapes and gating ratios, effort was directed to the study of the third sprue-base design, i.e., the well.

The Well

The schematic drawing in Fig. 8 illustrates a spruebase design employing a well. As implied by the term, the well consisted of a sump located directly beneath the sprue and extending below the level of the runner. Some earlier work¹ had been done with a well which showed that it was a faulty design. A long clean-up period was experienced during pouring because the sprue was projected into the well. In the present work, no such extension of the sprue was used.

After 1 sec-Note slight turbulence.



After 2 sec-Note minimum turbulence.

Fig. 10—Elimination of turbulence by use of a well in a narrow, deep runner. (Well 1 in. deep, 2½ in. diam; runner 1 in. wide, 1½ in. deep. Gating ratio 1:4:4.)

The optimum well dimension was determined for a large number of runner combinations, using the

TABLE 3-WELL DIMENSIONS PRODUCING THE SHORTEST CLEAN-UP TIME IN A TWO-RUNNER SYSTEM

			3/8-Sq-	In. Spr	ue		1/2-Sq-1	In. Spr	ue		3/4-Sq-1	In. Spr	ue	*	11/8-Sq	In. Spi	rue
mensio			Diam,		Clean- Up Time, Sec.	Gating Ratio	Diam,		Clean- Up Time Sec.	, Gating Ratio			Clean- Up Time Sec.	, Gating Ratio			
								Squa	re Runi	ier							
3/4	3/4	1:3:3	11/2	3/4	1.5	_	-	_		_	-	-	\$100 mg	_	_	_	West or
1	1	-	_	_		1:4:4	2	1	1.5	_	-	-	-	_	_	_	****
11/2	11/2	_		_	-	*****	-	-	-	1:6:6	21/2	2	2.5	1:4:4	21/2	1	2.0
1.5												- 1					
							W	ide, Sl	hallow F	Runner							
1	3/4	1:4:4	11/2	1	1.5	1:3:3	2	11/2	1.5	-	-	_	****	-	-		***
11/3	3/4	_	-	-		1:4:4	11/9	1	2.0	_	_	_	Militia	_	-	-	-
11/2	1	_	-	-	_	1:6:6	2	11/2	2.0	1:4:4	2	3/4	1.5	_	_	-	-
2	1	-	-	_	_	-	-	-	-	1:5.3:5.3	21/9	11/2	2.5	1:3.5:3.5	21/9	11/2	2.0
2	11/8	_	_	-	-	_	_	-	****	_	_	-	****	1:4:4	21/2	11/2	
							N	arrow,	Deep R	unner							
3/4	1	1:4:4	11/4	34	1.5	1:3:3	11/9	11/2	1.5	_	-		*	_	-		Marcon .
3/4	11/3	_	_		-	1:4:4	2	1	1.5	-	_	-	-	_	_	-	***
3/4	2	_	-	_	_	1:6:6	11/4	1	2.0	1:4:4	2	1	2.0		-	-	
1	11/2	_	-	_	_ =	1:6:6	2	: 1	2.0	1:4:4	2	11/2	2.0	-	_	-	******
1	2	_	_	_	-	_	_ '	_	-	1:5.3:5.3	21/2	1		1:3.5:3.5	21/2	3/4	1.5
11/8	2	-				-	_	-	-		-72	-		1:4:4	21/2	2	3.0

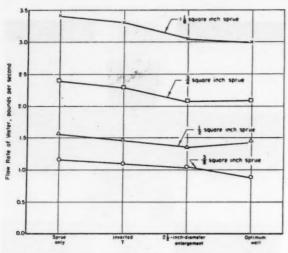


Fig. 11—Effect of sprue-base design on rate of flow in wide, shallow runner (gating ratio 1:4:4.)

Od rol circu do circu

Fig. 12—Relationship between sprue size and clean-up time for various gating ratios in a two-runner system employing a well under the sprue.

four different sizes of sprues. These data are summarized in Table 3.

The clean-up time obtained with the well having optimum diameter and depth ranged from 1 to 3 seconds. It must be remembered that during this investigation there was no mold attached to the runner. Consequently, very little back-pressure was exerted on the system. Undoubtedly, back-pressure created by an attached mold would have reduced these clean-up times to a significantly lower value.

The photographs in Figs. 9 and 10 show the action in the well when used with the wide, shallow runner and the narrow, deep runner. Note that a well functions satisfactorily in both types of runners.

From the data in Table 3, there appeared to be an empirical relationship between the dimensions of the well and the area of the sprue. An examination of the data showed that the area of the well should be about five times the area of the sprue, and the depth of the well should be about equal to that of

the runner. The data obtained on the effect of various depths and diameters of wells are recorded in Table A of the Appendix.

The use of a well or a sump is not new to the foundry industry. The well has been employed by a number of investigators^{3,4} as a means of providing a steady flow of molten metal to a fluidity spiral casting. The magnesium foundries have recently adopted a well at the base of the sprue as a place in which to place screens that are employed in the gating system. In the present investigation, the well produced a flow pattern which markedly reduced turbulence. Thus, clean-up of the system was effected in a very short time period for a large variety of runner and gate geometries.

As a result of this investigation, it was learned that a well under the sprue will (1) prevent aspiration of air into the liquid, (2) reduce the clean-up time to a reasonably low value, and (3) function in a wide range of runner geometries and sprue sizes. The only

TABLE 4—EFFECT OF INVERTED "T" SPRUE-BASE DESIGN ON CLEAN-UP TIME IN A ONE-RUNNER SYSTEM

Ru	nner	1/2-Sq-1	n. Sprue	3/4-Sq-1	n. Sprue	11/8-Sq-	In. Sprue
	sicns, In.	Gating	Clean-Up	Gating	Clean-Up	Gating	Clean-Up
Width	Depth	Ratio	Time, Sec	Ratio	Time, Sec	Ratio	Time, Sec
			Squa	re Runner			
1	1	1:2:2	00	1:1.3:1.3	3	1:9:9	00
11/2	11/2	1:4.5:4.5	00	1:3:3	00	1:2:2	00
2	2	1:8:8	10	1:5.3:5.3	3	1:3.6:3.6	00
			Wide, Sh	allow Runner			
1	3/4	1:1.5:1.5	2	1:1:1	2	1:.67:.67	2
11/2	1	1:3:3	3	1:2:2	00	1:1.3:1.3	2 2 3
2	1	1:4:4	25	1:2.7:2.7	00	1:1.8:1.8	3
2 2	11/2	1:6:6	6	1:4:4	00	1:2.6:2.6	90
			Narrow,	Deep Runner			
3/4	1	1:1.5:1.5	10	1:1:1	5	1:.67:.67	3
3/4	2	_	_	1:2:2	00	1:1.3:1.3	00
1	11/2	1:3:3	00	1:2:2	00	1:1.3:1.3	6
1	2	1:4:4	00	1:2.7:2.7	00	1:1.8:1.8	10
11/2	2	1:6:6	6	1:4:4	00	1:2.6:2.6	00

exception when the well failed to clean up the system of entrapped air was when large gating ratios were employed with the 1½-sq-in. sprue. However, limitations of equipment precluded the use of a well larger than 2½ in, in diameter. It may be possible that a well of larger diameter than 2½ in. would have resulted in a satisfactory clean-up time with the larger gating ratios and larger sprue.

A comparison of the effect of the various spruebase designs on the rate of flow is shown graphically in Fig. 11. It is interesting to observe that the well did not appreciably retard the rate of flow beyond that effected by the enlargement. The effectiveness of the well in reducing turbulence and the time required to wash out entrapped air bubbles must be attributed to some factor such as a positive flow pattern.

The plot shown in Fig. 12 was prepared by averaging the data obtained for the various types of runners with the optimum well dimensions. The data plotted for each of the four different sprue sizes indicate a small gating ratio with a small sprue is desirable if the shortest clean-up time is to be real-

ized. However, a gating ratio of the order of 1:4:4 is recommended to minimize turbulence in the mold cavity.

Three Sprue-Base Designs in Single-Runner System

A limited amount of effort was spent in investigating the effect of the various sprue-base designs in a single-runner system. A single runner was obtained by blocking off one-half of the system in the model shown in Fig. 1.

The inverted "T" design, which was very unsatisfactory in the two-runner system, was less prone to turbulence and aspiration in the single-runner system. Of 35 combinations of runners and sprues, only 18 aspirated air. These data are shown in Table 4.

As in the two-runner system, enlarging the runner under the sprue was most beneficial in the narrow, deep runner. The enlarged zone in the wide, shallow runner had a tendency to aspirate air into the system. The limited data presented in Table 5 on the effect of the diameter of the enlargement on the time required to wash out the entrapped air did not permit any recommendation to be made as to the optimum

Table 5—Effect of Enlargement Sprue-Base Design on Clean-up Time in a One-Runner System

			1/2-Sq-11	n. Spru	e		2	4-Sq-In. S	prue			11/8-Sq-In. Sp	rue
R	unner		Clean	-Up Ti	me, Sec			Clean-Up	Time, &	ec		Clean-Up	Time, Sec
	sions, In.	Gating	Diamece	r of Er	nlargemen	t Gating	Di	ameter of	Enlarger	nent	Gating	Diameter of	Emargemen
Width	Depth		1½ in.			Ratio		11/2 In.	2 In.	21/2 In.	Ratio	2 In.	21/2 In.
						Square	Runi	ner					
1	1	1:2:2	1.6	5.0	9.6	:1.3:1.3	_		_		1:.9:.9	_	90
11/2	11/2	1:4.5:4.5		4.0	9.1	1:3:3	_	_	00	00	1:2:2	1.6	2.2
2	2	1:8:8	-		11.0	:5.3:5.3	_	_	_	5.0	1:3.6:3.6	_	00
					W	ide, Sha	allow 1	Runner					
1	3/4	1:1.5:1.5	_	-	00	1:1:1	_	_	_	00	1:.67:.67	_	00
11/2	1	1:3:3	_	25.1	00	1:2:2	_	-	5.5	00	1:1.3:1.3	_	00
2	1 .	1:4:4		_	102.0	:2.7:2.7	_	_	-	00	1:1.8:1.8	-	00
2 2	11/2	1:6:6		_	20.0	1:4:4	_	-	-	6.5	1:2.6:2.6	-	1.8
					Ne	arrow, L	Deep F	Runner					
3/4	1	1:1.5:1.5		_	8.5	1:1:1	_	-	_	9.0	1:.67:.67	_	. 00
3/4	2	_	_	_	_	1:2:2	4.4	1.2	1.3	1.7	1:1.3:1.3	-	1.2
1	11/2	1:3:3	1.4	1.5	3.1	1:2:2	_	1.4	1.2	1.9	1:1.3:1.3	-	1.1
1	2	1:4:4	1.9	2.2	2.7	:2.7:2.7	_	1.6	1.2	1.5	1:1.8:1.8	_	1.3
11/2	2	1:6:6	-	6.9	23.7	1:4:4	_	_	2.4	3.1	1:2.6:2.6	ào	90

TABLE 6—WELL DIMENSION PRODUCING THE SHORTEST CLEAN-UP TIME IN A ONE-RUNNER SYSTEM

			1/2-Sq-I1	n. Sprue		3,	4-Sq-In. S	prue			11/8-Sq-I	n. Sprue	
	nsions, n.	Gating	Well Diam,	Well Depth	Clean- Up Time,	Gating	Well Diam,	Well Depth,	Clean- Up Time,	Gating	Well Diam.	Well Depth,	Clean- Up Time,
Width	Depth	Ratio	In.	In.	Sec	Ratio	In.	In.	Sec	Ratio	In.	In.	Sec
						Square	Runner						
1½ 2	11/2	1:4.5:4.5	2	3/4	3	1:3:3 1:5.3:5.3	$\frac{21/_{2}}{21/_{2}}$	3/4	2 4	1:3.6:3.6	21/2	3/4	31/2
					W	ide, Shall	ow Run	ner					
1½ 2 2	1 1 11/2	1:3:3 1:4:4 1:6:6	2 21/ ₂ 21/ ₂	3/4 3/4 3/4	2 2 31/2	1:2:2 1:2.7:2.7 1:4:4	21/ ₂ 21/ ₂ 21/ ₂	3/4 3/4 3/4	1 1½ 1½	1:2.6:2.6	21/2	3/4	11/2
					N	arrow, De	ep Runi	ner					
3/4	2 2	_	_	= ,	=	1:2:2 1:2.7:2.7	11/2	3/4	1	=	_	=	_
11/2	2	1:6:6	21/2	3/4	4	_	-	-	-	-	-	-	-

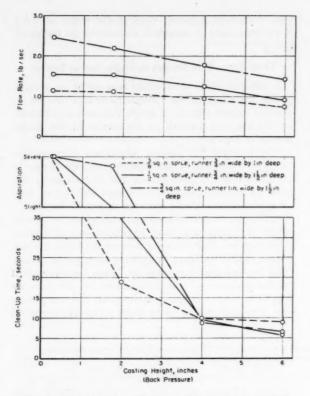


Fig. 13—Effect of back pressure on flow rate and clean-up time. A 1:4:4 geometry used for each sprue.

diameter. However, there appeared to be an indication that the diameter of the enlargement should not be greater than two times the width of the runner.

The use of the well in the wide, shallow runner substantially reduced the time to wash out the entrapped air. The well was most beneficial in those geometries susceptible to aspiration. Besides preventing aspiration, clean-up time was held to a low value.

The data for those systems in which the well produced a short clean-up time in the single-runner system were taken from Table B in the Appendix and summarized in Table 6. These data indicate that if a wide, shallow runner is used, then the well should have an area approximately five times that of the sprue. The depth of the well should be equal to that of the runner. If a narrow, deep runner is employed, there is no advantage in using a well. In this latter type of runner, an enlargement with a diameter approximately twice the width of the runner appeared to be satisfactory in obtaining the minimum clean-up time.

The flow rates for the four sprues when employed with the three sprue-base designs are recorded in Table C of the Appendix.

Back-Pressure and Sprue-Base Problem

Prior to this time, very little attention was given to castings having much vertical height. Therefore, a few laboratory tests were conducted in which the height of a theoretical casting was varied from 1/4 in to 6 in. The objective was to study the effect of backpressure, created by a casting, on aspiration in the runner.

Employing the sprue-base model shown in Fig. 1, a few changes were made in the discharge end of the runners. The Lucite end plates were eliminated and in their place a right-angle drain pipe was attached with its open end up. Circular pieces attached at the desired level inside the drain pipe established the height of the casting.

Because the inverted "T" design was the most susceptible to aspiration, it was selected for evaluating the effect of back-pressure on the runner system. This design was employed in a narrow, deep runner with three different sizes of sprues. A sprue-to-runner-to-gate ratio of 1:4:4 was maintained for all tests.

As in all of the previous experiments, a total sprue height of 7 in. and a head of 2 in. in the pouring box were maintained in each test. Thus, as the height of the casting increased, the difference between the height of the sprue and that of the casting decreased.

The chart in Fig. 13 shows that, as the height of the casting increased, the degree of aspiration decreased and the clean-up time was reduced. In spite of the decrease in velocity of the liquid as the casting height increased, the minimum time required to clean up the system of entrapped air was entirely too long when the inverted "T" sprue-base design was employed. For example, a clean-up period of 6 seconds was obtained when pouring a 6-in. high casting with a ½-sq-in. sprue. In this time period, 24 lb of aluminum would have passed through the system and the casting probably would have been severely contaminated with this damaged metal.

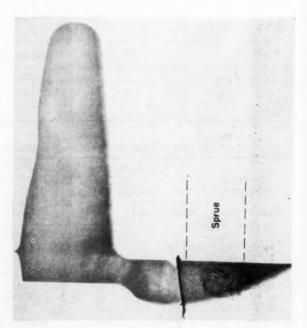


Fig. 14—Radiograph showing function of a vertical screen in runner. Black streaks on sprue side of screen are steel wool.

Inasmuch as the tendency toward aspiration and the clean-up time were reduced as a result of backpressure created by a vertical casting, it may be assumed that clean-up times tabulated in this paper for the systems employing an enlargement or well would also be reduced considerably by this backpressure effect.

Use of Screens in the Gating System

It is common practice in the light-alloy foundries to employ screens in the gating system for the purpose of holding back dross particles. Some of the methods of placing the screens are as follows:

1. A cylindrical screen placed in a well at the base of the sprue, the diameter of the screen being less than the diameter of the well.

2. A vertical screen placed across the runner at some location between the sprue and the first gate.

3. A horizontal screen placed at the parting line of the cope and drag. To employ this type of screen, the gates and a portion of the runner are placed in the cope.

Portions of runners containing screens were obtained from commercial castings. Radiographs were made of sections taken from these runners to examine the effectiveness of the screens.

The radiograph in Fig. 14 reveals the condition that existed when a vertical screen was used in the runner. Delineation of dross was not clear enough to draw a conclusion as to whether or not dross was formed and that it was trapped by the screen.

Large quantities of dross were contained within the cylindrical screen in the well illustrated in Fig. 15. The sprue, which was removed before radio-

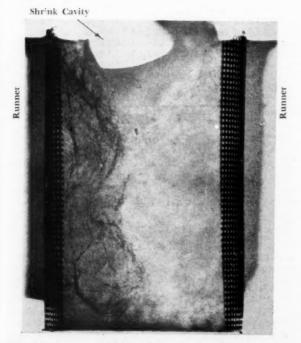
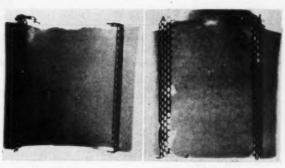


Fig. 15—Radiograph illustrating effectiveness of cylindrical screen placed in a well under sprue. Black streaks inside the screen are steel wool.



A—Tapered sprue with large end up.

B—Tapered sprue with large end down.

Fig. 16—Radiograph showing effectiveness of a screen in removing oxides from metal poured with (A) good gating practice, and (B) poor gating practice.

graphing, was directly over the center of the well. The runners had been attached to the top of the well. Thus, all of the metal had to pass through the cylindrical screen before entering the runner. It may be concluded from Fig. 15 that, if dross enters the system, it can be retained to some extent by the use of a screen.

Foundry tests were made to determine the effectiveness of a screen when a good gating practice was changed merely by substituting a tapered sprue with its large end down for one with its large end up. A cylindrical screen was placed in a well at the base of the sprue of each sand mold. The molds were poured with an aluminum-10% magnesium alloy at approximately 1400 F.

Radiographs taken on sections through the screens are illustrated in Fig. 16. No dross is evident in the radiograph in Fig. 16a, because a good sprue design was used. However, as would be expected, the reversed tapered sprue promoted the formation of the dross which was trapped by the screen, as shown in Fig. 16b. These results indicate that a screen is of value when the gating design is inferior. It should not be assumed, however, that screens are a good substitute for good gating practice, but rather that screens may be beneficial in supplementing good gating practice.

Attempts were made to evaluate screens with the water-transparent-mold techniques. It became evident, however, that additional development work was required before this approach could be used to yield information on how screens function.

Conclusions

Of the various types of sprue-base designs investigated, the well appeared to be universally applicable. Laboratory tests indicated that the well will prevent aspiration and provide a very low clean-up time for a wide range of gating ratios and runner shapes. This improvement in the condition at the base of the sprue was obtained without a marked reduction in the flow rate of the liquid, as compared with that effected by the enlargement type of sprue-base design.

APPENDIX

TABLE A—Effect of Well Diameter and Depth on Clean-up Time in a Two-Runner System

		Wel	78	-Sq-I	n. Sp	orue		y	-Sq-I	n. Sp	rue		3/4	-Sq-I	n. Sp	rue			S-Sq-I			
		Deptl	n, Clean	n-Up	Tin	ie, Se	ec	Clea	n-Up	Tim	ie, S	ec ec	Clean	n-Up	Tim	ie, So	ec	Clean				ec
		In.	11	ell D	Diam	eter		V	Vell I)iame	eter		11	Vell 1				W	ell I			
Width	Depth		Gating Ratio	l In.		In.	21/2 In.	Gating Ratio	l In.	11/2 In.	In.	21/2 In.	Gating Ratio	l In.	11/2 In.	In.	2½ In.	Gating Ratio		11/2 In.		21/2 In.
										Squ	ıare	Runn	er									
3/4	3/4	3/4	1:3:3	2.6	1.3	2.0	2.4	_	_	_	_	_	-	_		_	_		_	-	_	-
3/4	3/4	1	1:3:3	2.6	1.4	2.8	2.5	-	_	-	_	-	_	_	_	_	-	_		-		_
3/4	3/4	11/2	1:3:3	4.2	1.7	3.1	6.4	_	_	-		-	_	-	_	_	-	substitut.	-	_	-	-
3/4	3/4	2	1:3:3	4.2	2.0	2.9	8.5	1:4:4	7.2	3.7	1.9	2.2	_	dismile	-	_	_	_	_	_		_
1	1	3/4		_		_	_	1:4:4	8.5	2.4	1.6	2.7	_	_			_	_	_	_	_	_
1	1	11/2	_	_		_		1:4:4	12.8	1.8	1.6	4.2		_	-	_	-	-	_	_	_	-
1	1	2	-		_	_		1:4:4	11.1	2.0	1.7	4.8	_	-	_	_	_	_		-	_	_
11/2	11/2	3/4	-	_	0'-	_	_		_		_	-	1:6:6	_	00	00	5.4	1:4:4	_	00	90	2.5
11/2	11/2	1		_	-	-	_	-	_	_	-	-	1:6:6	_	00	00	3.1	1:4:4	_	00	00	2.0
	11/2	11/2	_	-	_	_	-	_	_	-	-	_	1:6:6	-	00	00	2.6	1:4:4	-	00	00	2.4
11/2	11/2	2	_	-	-	-	_		No	-		on P	1:6:6 unner	_	00	3.2	2.5	1:4:4	_	00	90	4.3
		-/	1.4.4		1.0	0.0	0.0	1.9.0					unuier									
3/4	1	3/4	1:4:4	00	1.3	3.3	3.3	1:3:3	8.0 16.0	2.4	1.9	2.4		_	_	_	_	_			_	
3/4	1	1 11/2	1:4:4	00	2.3.	3.9	6.2 9.8	1:3:3	4.5		1.9	6.3	_	_	_		_	_	_	-		-
3/4	1	2	1:4:4	00	2.4	3.3	7.8	1:3:3	46.0	1.8	2.1	4.2	_	_	_	_		-	-	_	-	_
3/4	11/3	1	_	_	_	_	_	1:4:4	6.7	2.4	1.7	6.0	-	_	_	_	_	_	-	_	_	-
3/4	11/3	11/3	_	_	_	_	_	1:4:4	5.3	1.8	1.9	5.5	_	_	-	_	-	_	_	_	_	-
3/4	11/3	13%	-	_	-	_	_	1:4:4	3.8	1.9	2.0	4.6		-	-	_	-		_		_	-
3/4	11/3	2	-	_	_	_		1:4:4	5.0	2.0	2.0	4.7		_	_	_	-	_	_	_	_	-
3/4	2	3/4	-	_	_	-	-	1:6:6	00	2.3	2.5	4.4	1:4:4	90	00	2.0	2.4	_	_	_	_	-
3/4	2	1	-	_	-	_	-	1:6:6	00	2.0	2.2	4.9	1:4:4	00	00	1.9	2.7	_	_		_	
3/4	2	11/2	_	_	-	-	_	1:6:6 1:6:6	00	2.7	2.9	4.4 5.0	1:4:4	00	00	2.4	3.2		_		_	
3/4	2	2	-	_	_	_	-	1:6:6	00	5.3		10.2	1:4:4	00	3.6	2.6	3.0		_	_		_
	11/2	1 3/4	_	-	_	ofession	_	1:6:6	00	3.3	2.0	6.4	1:4:4	00	2.5	2.3	2.7	-	_			_
	11/2	11/2	_	_		_	_	1:6:6	00	3.3	2.3	4.8	1:4:4	00	3.7	2.2	2.9	-	_	_	_	-
1	11/2	2	-	_	_	_	_	1:6:6	00	2.7	3.1	4.8	1:4:4	00	3.2	2.5	2.7	_	-	_	_	_
11/8	2	3/4	_		-		_	_	-	-	_	_	-	-	-	_	_	1:4:4	_	00	00	00
11/8	2	1	-				_	-	-	_	_	-	_	_	-	_	-	1:4:4	_	00	00	3.2
11/8	2	11/2	_	-	_	_	-	_	-	_	Automic	_		_	-	_	_	1:4:4	_	00	00	3.3
11/8	2	2		_	-	-	-	_	_	_	-	-		-	-	_		1:4:4	_	00	00	3.0
1	2	3/4	-	_	_		minuments	-	enterale	-	_	_	1:5.3:5.3	00	00	3.3	2.3	1:3.5:3.5	00	00	5.2	1.7 2.3
1	2	1			-		-	-	-		_	_	1:5.3:5.3	00	00	3.4	2.1	1:3.5:3.5 1:3.5:3.5	00	00	2.0	2.4
1	2	11/2	-		_	Marcon.	_	-	_	_	_	_	1:5.3:5.3 1:5.3:5.3	00	00	2.2	2.8	1:3.5:3.5	00	00	2.3	2.7
11/	2	2		_	Street	-	-					_	-	00	00	4.4	4.0	1:5.3:5.3		00	00	00
116	2	1 3/4	_					-	_	_	_	_		_	_	_	_	1:5.3:5.3	_	00	00	00
116	2	11/2			-		_	_	_	_		_	_	_	_	_	-	1:5.3:5.3	_	00	00	00
11/4	2 2	2	-	-					_	_	_	-	_	_	_	-	_	1:5.3:5.3	_	00	00	00
									Wie	de, S	hall	ow R	unner									
1	3/4	3/4	1:4:4	2.9	1.7	4.9	3.5	1:3:3	3.7	3.8	1.9	2.3	_	_	_	-	-		_	_	-	_
	3/4	1	1:4:4		1.5	3.4	3.5	1:3:3	5.5		1.8		-	_	_	_	-		_	-	_	-
		11/2	1:4:4			3.3		1:3:3		2.1			_	_	_		_	-	_		_	_
11/	3/4	2	1:4:4	3.0	2.2	3.1	9.7	1:3:3 1:4:4	4.0	3.0	3.5	4.1	-		_	_	_	-	_		-	_
11/3	3/4	3/4	_			_	_	1:4:4	_	2.1	2.3	6.6	_	_	_	_	_		_	-	_	_
11/3	3/4	11/2	_	_	_	_	_	1:4:4	_	2.2	3.3	6.8	_	_	_	_	-	****	_	_	-	_
1/3	3/4	2		_	_		_	1:4:4	- Second	3:3	2.3	6.9	_	_	_	-	-	property.		_	_	_
	1	3/4	_			_	_	1:6:6	_	3.9	2.6	3.9	1:4:4	_	00	1.7	3.3	more.	_	_	_	_
	1	1	-	-		_	-	1:6:6		3.6	2.7	5.9	1:4:4	-	00	2.7	2.9	-	_	-	_	_
	1	11/2		-	-	-	-	1:6:6/	-	3.0	2.1	5.1	1:4:4	-	00	2.4	2.5		_	_	-	-
1/2	1	2	-	-	-	-	_	1:6:6	_	3.3	2.3	4.8	1:4:4	_	00	2.3	2.7		_	-	_	-
	1	3/4	-	-	-	_	-	-	-	_	_	_	1:5.3:5.3	_	_		5.2	1:3.5:3.5	_	-	00	3.0
	1	1	-		-	_	-	-	_	_	_	-	1:5.3:5.3	-		4.3	3.9	1:3.5:3.5 1:3.5:3.5	_	_	00	2.1
	1	11/2	- 4	_	_	_	_	-	-	_	_	_	1:5.3:5.3	_	_	3.0	2.7	1:3.5:3.5	_	_	00	2.4
	1	2	-	_	_	_	-	_		_	_	_	1:5.3:5.3	_		2.9	2.1	1:4:4	_	_	00	3.6
	11/8		_	-	_		-	_		_	_	_	_		_		_	1:4:4	_	_	00	2.2
	6.00	1 11/2	_	_	_	- Contract		_	-	_		_	_	_	-	_	-	1:4:4	_	-	00	2.1
		2		_		_		_	_	_	_	******	_		_		-	1:4:4	_	_	00	2.3
	11/2	3/4		_	_	_	_		-	_	_	_	. —	-		_	_	1:5.3:5.3	_	_	oc	00
	11/2	1	_	_	_		_	_	_			_	_	_		_	_	1:5.3:5.3	-	_	oc	00
																		1:5.3:5.3	_	-	00	00
	11/2	11/2	-	-	-	_	-	Mercento	-	_	_	_	_	_	-	-	_	1:5.3:5.3	_		00	00

TABLE B-EFFECT OF WELL DIAMETER AND DEPTH ON CLEAN-UP TIME IN A ONE-RUNNER SYSTEM

					In. Spru					n. Spru				Sq-In.		
Rur	nner	Well		Cle	ean-Up	Time,	Sec		Cle	an-Up	Time,	Sec	Cl	ean-Up		
Dimensi	ions, In.	Depth,	Gating	V	Vell Dia	neter		Gating	M	ell Dia	meter		Gating	W	ell Dia	meter
Width	Depth	In.	Ratio	l In.	11/2 In.	2 In.	21/2 In.	Ratio	l In.	11/2 In.	2 In.	21/2 ln.	Ratio	11/2 ln.	2 m.	2½ In
							Squa	re Runn	er							
1	1	3/4	1:2:2	00	2.1	1.7	2.1	-	_	_	_	_	-	_	-	_
-	-	1	1:2:2	00	1.9	2.6	1.9		_	_	_	_	-	_	_	-
_	_	11/2	1:2:2	00	2.6	8.3	3.8	roman	-		_	_	_		-	_
_	_	2	1:2:2	90	3.4	7.0	5.6	-	_	-	_	_		_		_
11/2	11/2	3/4	1:4.5:4.5	-	00	3.1	3.0	1:3:3	-	00	6.3	2.2	1:2:2	00	8.8	2.2
-	-	1	1:4.5:4.5	_	00	5.1	3.7	1:3:3	_	90	10.0	3.0	1:2:2	00	00	2.2
_	_	11/2	1:4.5:4.5	_	00	5.8	8.0	1:3:3	_	00	3.8	4.1	1:2:2	00	2.8	13.8
-	-	2	1:4.5:4.5	-	6.1	6.3	9.3	1:3:3	_	00	4.5	7.5	1:2:2	90	3.8	6.0
2	2	3/4	-	-	_	-	_	1:5.3:5.3	_	_	4.0	4.1	1:3.6:3:6	_	90	3.3
-	_	1	-	-	_	-	_	1:5.3:5.3	_	_	4.5	5.0	1:3.6:3.6	-	00	7.0
	-	11/2	_	-	_	_	_	1:5.3:5.3	_	_	5.0	7.0	1:3.6:3:6	_	00	00
-	-	2	_	-	_	-	_	1:5.3:5.3	_	_	6.4	8.2	1:3.6:3:6	_	00	00
						W	ide, St	nallow R	unner							
11/2	. 1	3/4	1:3:3		2.7	1.8	1.8	1:2:2	-	2.2	1.8	1.0	_	-	_	
-72	_	1	1:3:3	_	2.7	2.1	2.0	1:2:2	_	00	2.1	1.1	_	-	_	_
_	_	11/2	1:3:3	_	3.0	5.5	2.8	1:2:2	_	00	10.0	3.8	_	_	_	-
	_	2	1:3:3	_	3.5	7.7	9.7	1:2:2	_	00	5.8	5.8	_	_	_	-
2	1	3/4	1:4:4	_	_	2.9	2.1	1:2.7:2.7	_	_	00	1.5				
_	_	1	1:4:4		_	3.5	2.6	1:2.7:2.7	_	_	5.8	1.9	_	_	_	_
-		11/2	1:4:4	_	_	00	4.0	1:2.7:2.7	_	_	00	3.1	-	_	_	
-	-	2	1:4:4	_	_	10.6	6.2	1:2.7:2.7	-	-	00	6.1	_	_	_	_
2	11/2	3/4	1:6:6		_	3.6	3.3	1:4:4	_	_	3.6	1.4	1:2.6:2.6	_	00	1.4
-	- /2	1	1:6:6	_	_	4.2	6.8	1:4:4	_	_	3.1	2.3	1:2.6:2.6	Name of Street, or other Designation of Street, or other Desig	00	2.1
_	-	11/2	1:6:6	_	_	5.7	4.7	1:4:4	_	_	4.1	2.6	1:2.6:2.6	_	90	2.3
_	_	2	1:6:6		_	5.9	6.7	1:4:4		-	3.4	5.2	1:2.6:2.6	-	90	4.5
								Deep R	unner							
2/	2	9/						1:2:2	7.8	0.9	1.5	2.5				
3/4	_	3/4	_	_	-	_	_	1:2:2	6.7	2.0	1.9	4.0	_	_	_	_
_	_	11/2	_	_	_	_	_	1:2:2	6.5	2.3	3.5	5.2	_	_	_	_
_	_	2	_	_	_	_	_	1:2:2	6.7	1.9	3.6	7.2	_	_		_
1	11/2	3/4	1:3:3	00	2.2	2.8	2.6	1:2.2	00	90	2.8	1.9	_		_	_
	- 1/2	1 74	1:3:3	00	2.4	3.3	3.8	1:2.2	00	1.9	4.0	3.4	_	_	_	_
_		11/2	1:3:3	-	2.7	6.6	13.7	1:2:2	00	1.8	5.0	3.7				_
_	_	2	1:3:3	4.3	3.4	5.7	14.3	1:2:2	00	1.9	5.3	8.1	_			_
1	2		1:4:4	9.3	2.4	2.4	3.7	1:2.7:2.7		1.9	1.1	3.0	-			
_	_	3/4	1:4:4	00	2.0	2.3	5.8	1:2.7:2.7	00	2.0	1.7	3.9			_	_
		11/2	1:4:4		3.1	4.9		1:2.7:2.7	00	1.9	2.2	5.0	_	_	_	_
_	_	2	1:4:4	00	2.1	5.7	8.3	1:2.7:2.7	00	1.7	3.7	6.9		_	_	_
11/2	2		1:6:6		27.0	4.8	4.2	1:4:4	-	00	6.1	3.1	1:2.6:2.6	00	00	00
		3/4	1:6:6	_	14.0	6.2	5.5	1:4:4	_	00	3.6	4.0	1:2.6:2.6	00	30	00
_	_	11/2	1:6:6	_	22.0	8.2	6.2	1:4:4	_	00	3.7	5.7	1:2.6:2.6	00	00	00
_		2	1:6:6		10.3	9.8	7.4	1:4:4	_	00	3.7	7.4	1:2.6:2.6	00	00	00

It appears that the minimum clean-up time was obtained when the well area was approximately five times that of the sprue, and its depth was equal to the depth of the runner. A smaller gating ratio, 1:4:4 rather than 1:6:6, produced shorter clean-up times for a particular sprue size and a particular shape of the runner. For a given gating ratio, clean-up time was decreased by using smaller sprues in conjunction with a well at their base.

The enlargement type of sprue-base design was effective in preventing aspiration and reducing clean-up time in a narrow, deep runner. With this type of runner, the diameter of the enlargement should be about 2½ times the width of the runner. In a square runner and wide, shallow runner, the enlargement usually resulted in a prolonged clean-up period and slight aspiration in some instances.

The inverted "T" type of sprue-base design caused severe aspiration and turbulence in all runner shapes

and gating ratios. This design also produced the highest flow rates which probably intensified the undesirable aspiration effect.

The behavior of the three sprue-base designs was, in general, similar in both the single- and double-runner systems.

Screens appeared to be beneficial in trapping dross particles that entered the gating system. However, screens cannot be recommended as a substitute for good gating practice.

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TABLE C—FLOW RATES FOR THE FOUR SPRUES EMPLOYING THE THREE SPRUE-BASE DESIGNS IN A TWO-RUNNER SYSTEM

				Sprue Rate, I		1/2		Sprue Rate, L	b/Sec			Sprue Rate, L			-	. Sprue	
mensi	ner Di- ons, In.		T De-	2½- In. Diam Enlarg.	mum	Gating Ratio	T De-	2½- In. Diam Enlarg.			verted T De-	2½- In. Diam Enlarg.	mum		T De-	21/2- In. Diam Enlarg	mun
								Square	Runn	er							
3/4	3/4	1:3:3	1.13	1.08	_	_	_	_		-	_	_		_	_	_	_
1	1	1:5.3:5.3	1.12	1.08	_	1:4:4	1.54	1.44	_	1:2.7:2.7	2.33	1.78	-	1:1.8:1.8	2.98	2.33	
11/2	11/2	_	_	_	-	1:9:9	1.33	1.24	_	1:6:6	2.32	1.99	-	1:4:4	3.40	2.74	_
2	2	-	_		_	1:16:16	1.25	1.21	_	1:10.7:10.7	7 2.03	1.92	-	1:7.1:7.1	3.04	2.87	_
							Nari	row, D	eep R	unner							
3/4	1	1:4:4	1.17	1.06	_	1:3:3	1.52	1.31	_	1:2:2	2.03	1.58	-	1:1.3:1.3	2.35	2.02	_
3/4	11%		-	_	-	1:4:4	1.53	1.34	-	_	_	_	_	_	_	_	
3/4	2	_	_	manager .		1:6:6	1.49	1.34	_	1:4:4	2.17	1.96	_	1:2.7:2.7	2.67	2.60	
1	11/2	1:8:8	1.15	1.06	-	1:6:6	1.55	1.47	_	1:4:4	2.42	2.10	2.00	1:2.7:2.7	3.26	2.59	_
1	2	_	_	-		1:8:8	1.48	1.38	_	1:5.3:5.3	2.26	2.07	-	1:3.5:3.5	3.11	2.69	_
11/8	2	-	_	_	-	_	_	_	_	_	-	_	-	1:4:4	3.22	2.95	_
11/2	2	_	_	-	_	1:12:12	1.38	1.31	_	1:8:8	2.24	2.04	_	1:5.3:5.3	3.42	2.89	_
							Wide	, Shal	low R	unner							
1	3/4	1:4:4	1.08	1.05	0.92	1:3:3	1.55	1.27	-	1:2:2	2.22	1.78	_	1:1.3:1.3	2.80	2.19	_
11/3	3/4	_		_		1:4:4	1.45	1.34	1.40	_	-		-	_	_	-	-
11/2	1	_	_	*****	_	1:6:6	1.45	1.31	-	1:4:4	2.32	1.98	2.10	1:2.7:2.7	3.26	2.74	_
	1	_	-	-	_	1:8:8	1.39	1.33	-	1:5.3:5.3	2.24	2.08	_	1:3.5:3.5	3.22		_
2	11/8		_	_	-	_	_	_	_	_	_		-	1:4:4	3.30	3.04	2.99
2	11/2	_	_	_	_	1:12:12	1.30	1.32	-	1:8:8	2.07	2.08		1:5.3:5.3	2.94	2.33	-

sack, *Vice-Chairman*; C. E. Nelson, T. D. Stay, W. J. Klayer, and W. E. Sicha, who assisted in the direction of the project on behalf of A.F.S.

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careful criticisms and suggestions of Dr. C. H. Lorig and many other associates of the authors at Battelle.

References

- 1. "A Study of the Principles of Gating," by R. E. Swift, J. H. Jackson, and L. W. Eastwood; A.F.S. Transactions, vol. 57, pp. 76-88 (1949).
- 2. "A Study of the Principles of Gating," by K. Grube and L. W. Eastwood; A.F.S. Transactions, vol. 58, pp. 76-107 (1950).
- 3. "A Contribution to the Study of Fluidity of Cast Iron," by S. Perin and R. Berger; A.F.S. TRANSACTIONS, vol. 42, pp. 589-682 (1934).
- 632 (1934).
 4. "A Fluidity Test for Aluminum Casting Alloys," by W. E. Sicha and R. C. Boehm; A.F.S. Transactions, vol. 56, pp. 502-507 (1948).

INVESTIGATION OF HOT TEARS IN STEEL CASTINGS

PART 1-TEST SPECIMEN FOR EVALUATING HOT TEARING IN STEEL CASTINGS

By

Gordon W. Johnson*

Foreword

This report describes work carried out under sponsorship of the Steel Research Committee† of the American Foundrymen's Society at the Armour Research Founday, Burnside Steel Foundry, and the American Steel Foundries from May 28, 1948, to February 1, 1952. This report is in three parts. Part I was prepared by the Armour Research Foundation, Chicago; Part II by the American Steel Foundries, East Chicago, Ind.; and Part III by Burnside Steel Foundry Co., Chicago. The program objective has been to determine the influence of mold factors such as mold collapsibility, sand binders, hot strength, density, etc., on the development of hot tears in steel castings.

The influence of these properties has been surmised through observations of castings made under industrial conditions, but too often the real cause of hot tears has been obscured by the multiplicity of factors involved. Therefore, it was obvious that if the influence of only sand properties was to be satisfactorily investigated it would be essential that other contributing factors be controlled. Of these other contributing factors, design of the test piece was the most significant. Hence, the major portion of the program was devoted to the development of a suitable test casting whose hot tearing tendencies could be evaluated.

Part I of this report outlines the work done during the development of such a test specimen and shows the various designs which were tried before a satisfactory method was evolved. Test data are presented, indicating the sensitivity and reproducibility of this specimen over a range of core sands varying in hot strength, collapsibility and density.

The material presented in Part I will be divided,

for the sake of clarity, into four "phases." These phases are chronological in order, and represent time periods in which different specific approaches were made to the problem of the development of a satisfactory test specimen.

Melting Practice

Induction melted steel of a commercial grade was used for all the tests. Its composition fell within the following range.

Carbon	0.20 - 0.25	pct
Manganese	0.70-0.75	pct
Silicon	0.30-0.35	pct
Sulphur	0.02-0.03	pct
Phosphorus	0.03 - 0.04	pct

Deoxidation was accomplished with a 0.1 per cent aluminum addition followed by a 0.15 per cent calcium addition. Pouring temperature ranged between 2940 and 2960 F. Temperature measurements were made with a Pt-Pt Rh immersion thermocouple.

PHASE I

A. Experimental Technique

During the first phase of the work, extending from May 28, 1948 through June 1, 1948, experiments were made with various cylindrical specimens, as shown in Table 1.

TABLE 1-DIMENSIONS OF CYLINDRICAL TEST SPECIMENS

Inside Diameter, in.	Wall Thickness, in.	Height, in.
3	1/4	4
3	1/2	4
3	3/4	4
6	1/4	4
6	1/2	4
6	3/4	4

The test mold with cores in position is shown in Fig. 1. The cope and drag assembly with the arrangement of full ring risers above each cylinder is shown in Fig. 2. Figures 3 and 4 show the runner arrangement before and after closing the mold.

Each casting poured consisted of a nest of six rings, including one each of the rings described in Table 1;

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C. W. Briggs, Steel Founders' Society of America, Cleveland G. W. Johnson, Vanadium Corporation of America, Chicago

G. A. Lillieqvist, American Steel Foundries, East Chicago, Ind. C. H. Lorig, Battelle Memorial Institute, Columbus, Ohio

H. A. Young, Crane Co., Chicago

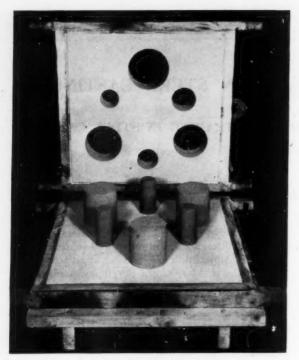


Fig. 1—Cylinder test casting mold showing cores in position.

each ring had the same type core material, so comparative results were obtained for various wall thicknesses in two different ring diameters. In this manner it was hoped to establish an over-all relationship between ring dimensions and tendency toward hot tearing.

Three tests were made with each of six different core mixtures and poured as nearly the same as possible. The six core mixes used are shown in Table 2. After pouring the castings were removed from the mold, sand blasted, and measured on the inside diameter to determine the change in size when made with different core mixes. The castings were then etched in hot hydrochloric acid before inspection for hot tears.

B. Resul's and Conclusions from Phase I Data

Table 2 describes the various core mixtures studied in terms of their compositions, hot strengths, and the degree of tearing experienced with each mix. (Degree of hot tear is expressed as the total length in inches of all the tears in the casting.) High temperature strength determinations show a general trend toward increased hot strength with increasing silica

flour content. However, the inadequacy of any of the cylindrical test specimens to show this effect in terms of increased hot tear susceptibility is shown by an examination of the test data. Some variation in ring diameter occurred with various core mixtures, but no definite correlation was shown between change in diameter and hot tears which resulted. The hot tears were not confined to any specific sand mixtures or casting size.

This inconsistency of results established beyond doubt the need for an improved test specimen having greater sensitivity and improved reproducibility.

PHASE II

A. Experimental Technique

During the second phase of the investigation extending from June 1, 1949 through February 10, 1950, a cylinder having a 6-in. inside diameter and ½-in. wall thickness was used. Figure 5 shows the design of the specimen. Each casting was made in an individual green sand mold which contained the dry sand core. A metal flow-off on each of the four sides of the mold was employed to prevent any overhanging fins from forming around the riser of the casting. The single gate used previously was replaced by a series of eight ½-in. pencil gates spaced uniformly around the circumference of the cylinder.

Since the primary purpose of this series was to determine reproducibility of the test specimen, only two sand mixtures were used and all variables such as core densities, chemical analysis of the metal, and pouring temperature were closely controlled.

All tests run at Armour Research Foundation were duplicated by the Burnside Steel Foundry Co., both organizations following a detailed outline of procedure relating to sand mixing, coremaking, pouring and inspection.

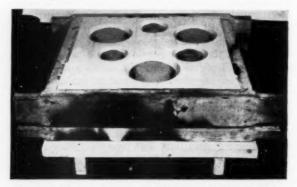


Fig. 2-Cylinder test casting mold showing full ring risers.

TABLE 2-CORE MIXTURES AND DEGREE OF HOT TEARING

		Core Mi	xture, %		Hot Stre	ength, psi	Length of Tear, in.							
Code	No. 60 Sand	Silica Flour	Cereal Binder	Core	12 min 2000 F	12 min 2500 F	3x1/4 Cyl	3x1/2 Cyl	3x3/4 Cyl	6x1/4 Cyl	6x1/2 Cyl	6x3/2		
A	98	0	2	0	1	0	0	0	0	25/8	θ	0		
В	83	15	2	0	2	24	0	0	0	0	0	0		
C	68	30	2	0	25	20	0	0	0	0	. 0	0		
D	97	0	2	1	1	2	2	12	71/2	61/2	0	0		
E	82	15	2	1	4	61	17/8	0	5/8	51/2	0	0		
F	67	30	2	1	7	77	0	0	0	0	0	0		

The core sand mixture used for these tests contained:

No. 60 Sand,	07	*	×		 ×		,					83
Silica Flour,	70			*		 ×					*	15
Cereal Binder,	%				*			*				1
Core Oil, % .									*	*		1
Moisture, %												

Each organization produced 24 cores, half of which were exchanged, so that four comparative sets of data were obtained:

- 1) Burnside cores poured at Burnside.
- 2) Burnside cores poured at Armour Research.
- 3) Armour Research cores poured at Burnside.
- 4) Armour Research cores poured at Armour Research.

B. Results and Conclusions from Phase II Data

A study of the results obtained from these four classes of data showed the following conditions to exist:

- 1) Test castings poured at Burnside Steel, using their own cores, produced a range of tearing from 0 to $9\frac{1}{2}$ in.
- 2) These same cores poured at the Armour Research Foundation produced a range of tearing from 7/8 to 35/8 in.
- 3) Test castings poured at Burnside Steel, using Armour Research Foundation cores, produced a range of tearing from 1 to 7½ in.
- 4) These same cores poured at Armour Research Foundation produced a range of tearing from $3\frac{1}{2}$ to 9 in.

Since considerable data concerning core density and pouring temperature had been accumulated during

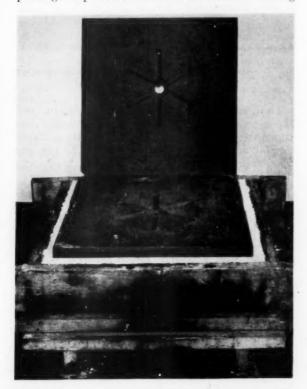


Fig. 3—Cylinder test casting mold showing runner slab before closing.

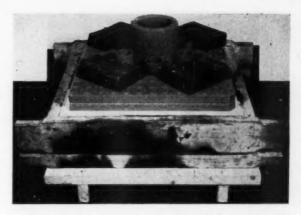


Fig. 4—Cylinder test casting mold ready for pouring showing gate runner system.

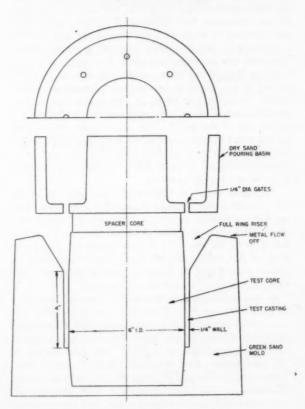


Fig. 5-Sketch of test specimen.

the pouring of these four series of tests, an attempt was made to determine their effect by plotting graphically against degree of hot tearing. A summary of these results showed trends of general interest; for example, as core density was increased the degree of tearing also increased. However, the objective desired—a test specimen capable of yielding highly reproducible hot tears over a large range of cores—was not attained.

PHASE III

A. Experimental Technique

During the third phase of the investigation extend-

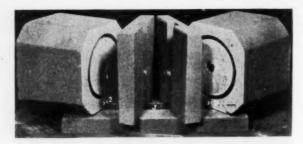


Fig. 6—Exploded view of mold showing position of cylinder cores (1), gates (2), runner system (3), and downgate (4).

ing from February 10, 1950 through August 1, 1951, a cylindrical specimen 6 in. in diameter and with a 3/8-in. wall thickness was used. The length of the cylinder was increased from 4 in. to 6 in., and risering of the specimen was eliminated. These specimens were poured in pairs with the axis of the cylinder in a horizontal, rather than a vertical, position. Each specimen was bottom-gated through a single slot gate 5 in. in length and 1/8 in. wide. By this method of gating it was possible to develop a continuous and highly localized hot band at the low point of the cylinder and diametrically opposite the coldest metal at the top of the mold. The down gate and runner system arrangement consisted of five cores plus a pouring cup shown in exploded view (Fig. 6) and completely assembled (Fig. 7). This assembly was clamped in a fixture before pouring.

The first series of tests with this new specimen was poured at the Armour Research Foundation. The one-piece cylinder core was rammed in a core box containing an insert plate designed so as to form the slot gate in the test core during ramming. Examination of the data collected from these tests shows a rather wide range of hot-tear lengths obtained with each of the three core mixtures studied.

It was subsequently determined during tests run at the American Steel Foundries that this effect was due to variations in slot gate width and that reproducibility of results was dependent to a great extent on the maintenance of accurate slot gate widths in all specimens. The Burnside Steel Foundry tests (described in Phase IV) were all run with slot gates machined to size after baking the core, and as a result a greater accuracy of gate width, and consequently a narrower range of hot tear length, was obtained.

Table 3 describes the three core mixtures used in the Phase III series of tests.

· Forty-five cores were tested: three different batches of five from each of the three core mixes. Eight different heats of metal were involved in the pouring of these specimens.

B. Results and Conclusions from Phase III Data

Table 4 describes the various tears obtained with each of the three types of core sand tested.

TABLE 3—CORE MIXTURES USED IN PHASE III
TEST SERIES

-	Core Mix A	Core Mix B	Core Mix C
Sand, No. 60, %	98	98.5	98
Cereal Binder, %	1	1	1
Core Oil, %	1	0.5	0.5
Kerosene, %	0.25	0.25	0.25
Moisture, %	4	4	4.5
Wood Flour, %			0.5

TABLE 4—CORE MIXTURES AND HOT TEARS

Core Mix A	Core Mix B	Core Mix C
1—6 in. hot tear 2—5 $\frac{1}{2}$ in. hot tears 8—5 in. hot tears 3—4 $\frac{1}{2}$ in. hot tears 1—4 in. hot tear	$4-4$ in. hot tears $7-3\frac{1}{2}$ in. hot tears	8-21/2 in. hot tears

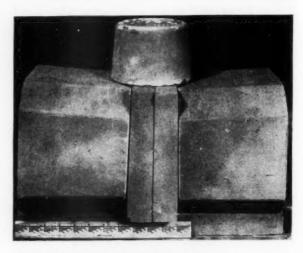


Fig. 7—Assembled test casting mold ready for pouring casting.

Thus, average tear lengths of 51/4 in., 33/4 in., and 23/4 in. were obtained on the three core mixtures containing respectively 1 per cent core oil, 0.5 per cent core oil, and 0.5 per cent core oil with a 0.5 per cent wood flour addition.

PHASE IV

During the fourth and final phase of the investigation extending from August 1, 1951 through February 1, 1952, a final check on reproducibility and sensitivity of the new test specimen was undertaken by Burnside Steel Foundry. Eleven different types of core sand were studied and several hundred cylinders poured in a very thorough investigation of the specimen's characteristics. The effect on hot tearing of various width slot gates was studied, each sand mixture being evaluated for its hot tearing properties when gated with a ½-in., ¾-in., 1¼-in. and 3½-in. slot.

PART II-EVALUATION OF HOT TEAR TEST CASTING

By

G. A. Lillieqvist and J. H. Fuqua**

ABSTRACT

The casting used by the Armour Research Foundation was investigated at the American Steel Foundries Research Laboratory to determine its sensitivity and reproducibility of results. It was found that the sensitivity of the casting was dependent upon the width of the gate which varied from core to core due to the manner of molding the gate and sagging of the relatively weak cores. The sagging can be overcome only by redesigning the core box. A change in the manner of molding the gate was tried at the Research Laboratory without success.

The inherent reproducibility of the casting was difficult to determine due to the variation introduced by the fluctuations in gate sizes. Included in the investigation were castings made with the same three core sand mixes that Arm ur Research used in evaluating the casting. While the Armour Research Foundation's results were not reproduced within what might be termed "reasonable error," the general trend for the strongest m xes to cause the largest tears was substantiated in our work as well as theirs.

Recommendations have been made that changes be made in the pattern equipment and manner of producing the cores to overcome the variables mentioned.

Object of the Investigation

The object of this investigation was to determine whether the results obtained from the Armour Research Foundation hot tear test casting are reproducible, and whether the casting has suitable sensitivity for evaluating the influence of various sand mixes on the occurrence of hot tears in steel. This work and similar studies to be carried on at the Burnside Steel Foundry are made to aid the Research Committee of the Steel Division of the American Foundrymen's Society.

Summary and Conclusions

The casting designed and tested by the Armour Research Foundation for studying the influence of core mixes on hot tear occurrence in steel has been studied and evaluated at the Research Laboratory. The following results were obtained and conclusions drawn:

1. Sensitivity

To measure the sensitivity of the A.R.F. casting, core mixes were selected which when used with the A.S.F. Research Laboratory casting had evidenced varying degrees of influence on hot tear occurrence. These mixes were the regular No. 3 linoil mix, the No. 3 mix plus iron oxide, and the superset mix. The superset and iron oxide mixes were shown to have a hot tear rating number of 1 using the Research Laboratory's RCT-4 ½-in. wall thickness hot tear test casting. The No. 3 mix, our standard for comparison, has an average rating of about 3, considerably

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more severe than the two other mixes. Results obtained using these mixes with the A.R.F. test casting indicated that the casting was exceedingly sensitive. Some castings from superset and iron oxide cores had tears as severe as some of the largest tears in castings made in No. 3 mix cores. Other castings from cores of each of the three mixes had only slight tears. These variations indicated that some variable other than the sand mix influenced the sensitivity.

2. Pattern Equipment

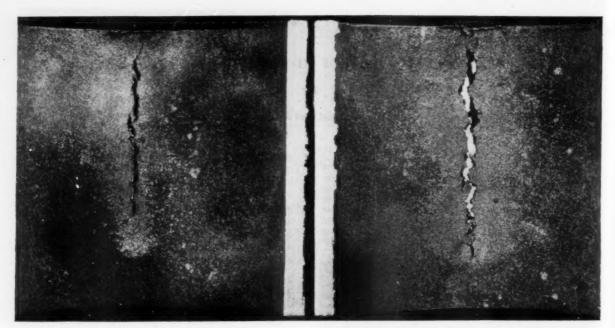
Considerable difficulty was encountered in remov-

TABLE 5—RESULTS OF HOT TEAR CASTINGS MADE USING CORE MIXES A, B, AND C RECOMMENDED BY ARMOUR RESEARCH FOUNDATION

Heat No.	Core Mix			Average Wi	dth Size of Tear, in.
7234	A	2	1	0.101	1/32×11/2
	A		5	0.103	1/32x2
	A	2	3	0.082	1/04×2
	A		7	0.105	1/16×3
	A	2	2	0.094	1/64×21/4
	A		4	0.125	3/32×4
	A	2	6	0.092	Slight indication
	A	2	8	0.107	1/32x21/4
	A	2	9	0.115	1/8×4
	A		10	0.097	1/32×3
7235	B1	2	1	0.083	1/04×11/2
	B1		2	0.083	1/32×11/2
	BI	2	3	0.088	1/32×5
	B1		4	0.052	Misrun casting
	BI	2	6	0.043	Misrun casting
	Bl		7	0.052	1/64×1/2
7252	B2	2	1	0.105	1/64×4
	B2		2	0.060	Indication
	B2	2	3	0.107	1/64×21/2
	B2	-	5	0.114	1/32×21/4
	B2	2	4	0.082	1/64×1
	B2		6	0.072	No tear
7238	C	2	1	0.068	No tear
	C		6	0.070	Misrun
	C	2	2	0.099	1/04x1/6 Erosion
	C		8	0.113	No tear - Erosion
	C	2	4	0.080	No tear - Erosion
	C		5	0.077	No tear - Erosion
	C	2	7	0.069	No tear - Erosion
	C	_	9	0.066	Misrun

TABLE 6-CHEMICAL ANALYSES OF HEATS

Heat No.	С	Mn	Si	P	S	Special	Deoxidizers,	lb/ton
7234	0.25	0.65	0.41	0.014	0.031	3 lb	Ca-Mn-Si+2)	6 lb Al.
7235	0.26	0.71	0.43	0.013	0.032			-
7238	0.26	0.67	0.42	0.015	0.031			
7252	0.23	0.63	0.44	0.014	0.030			



ing the ccres for the cylindrical portion of the casting from the core box. Because of the thin outside wall of the core, those cores which were drawn successfully sagged at the gates. This was especially true of the weaker core mixes. Painting the pattern equipment and venting the box to prevent any partial vacuum at the center of the cylindrical core box remedied the drawing difficulties somewhat. Before using cores for investigation of reproducibility, the gate width of each core was measured with a feeler gauge to determine the effect of variations in gate size on the results.

3. Reproducibility

To determine whether the results obtained with

Fig. 8—Drag side of castings from No. 3 mix cores. Both castings were poured in the same gating system. Note difference in size of hot tears.

this casting could be reproduced, three mixes (A, B and C) used by the Armour Research Foundation were used for making the cylinder cores. A.R.F. reported tears of an average size of $\frac{3}{16} \times 5$ in. with "A" mix castings, $\frac{1}{8} \times 4$ in. with "B" mix castings, and $\frac{1}{16} \times 2$ in. with "C" mix castings. Little varia-

Fig. 9—Drag side of castings from No. 3 mix cores made from the same batch of sand as those in Fig. 8. Castings were poured from same heat as in Fig. 8. Tears are all of different dimensions.



TABLE 7—SAND MIXES USED FOR ARMOUR RESEARCH HOT TEAR STUDY

Mix	A	Bl a	C	
Cereal Binder, %	1	1	l	1
Core Oil, %	1	(0.5	0.5
Kerosene, %	0.25	(0.25	0.25
Wood Flour, %	-	_	_	0.5
AFS-GFN 85 Ottawa washed				
and graded sand, %	98	98	3.5	98
Mechanical	Propert	ies		
Moisture, %	4.0	4.2	4.7	4.3
Green Permeability				
Green Compressive Strength, psi	0.98	0.87	1.0	0.91
Baked Compressive Strength, psi	1660	842	1256	1674
Baked Tensile Strength, psi	282.5	180.5	262.9	274.4
Core Hardness Number	98.4	89	100	99
Compressive Strength, 2500 F				
1-min heating interval	19.3	13.3	7.5	8
2-min heating interval	17.5	7	4	3.5

tion was reported within castings from similar core mixes. Our investigation produced hot tears in castings from "A" mix cores varying from minute cracks to tears $\frac{3}{32} \times 5$ in. No two castings had tears of identical size whereas Armour reported 8 of 15 castings had tears $\frac{3}{16} \times 5$ in, with this mix. "B" mix castings varied from slight indications of tears to $\frac{1}{32} \times 5$ in., while the majority of castings from "C" mix cores had no tears. The same trend for the A, B and C mixes to produce progressively fewer tears was indicated from our results. The magnitude in the case of each mix was much smaller. Therefore, a correct indication of the effect of core mixes was obtained. Satisfactory reproducibility from castings made in cores of similar composition was not obtained.

4. Effect of Gate Thickness on Sensitivity

Variation in gate thickness due to design and sagging of the cores was found to have a considerable influence on the severity of the hot tears. The gate size of the pattern was 0.0934 in. in width. (This necessitated metal temperatures above 2900 F to avert

Fig. 10—Cylinder cores showing location of gate and variation in dimensions of gate. In considering gate size, thicknesses at three noted locations were measured with a feeler gage and averaged. Because of insufficient draft, the inside core was always raised somewhat about the outside walls, and had to be ground flush with outer walls.

TABLE 8—COMPARISON BETWEEN A.R.F. RESULTS WITH A, B, AND C SAND MIXES AND THOSE OBTAINED AT A.S.F. RESEARCH LABORATORY

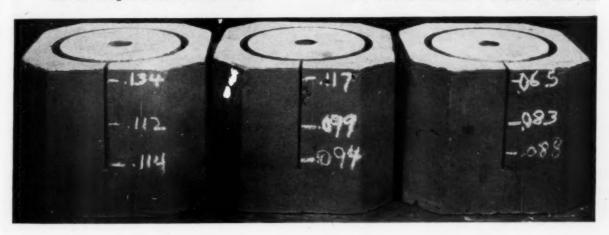
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l	**	1/4	**	51/9		0		1	**	1/8	*		,,	
1	**	3/16	7.5	51/				1	**	1/16	,	3	**	
8	312	3/16	**	. 5				1	**	1/32			29.	
3	2.5	3/16	99	41/2				1	**	162		21/	4 "	
1	2.2	3/16	2.0	4	**			1	**	1/32			* **	
		/10						1	**	1432			6 "	
								1	**	1/64		21/		
								1	12	1/64				
								1.	M	inut				
							Core	M				44.0 (24		
1	tear	3/4	in v	vide	414						in	wide	5 in. l	one
2	**	1/8		"	41/9			î	11		in.	11	21/4 in.	ong.
4	**	1/8		21	4			1	**		in.	246	11/2 in.	**
7	22	1/8		**	31/2			î	**		in.	**	4 in.	2.5
ì	2.5	1/16		**		in.		1	9.5		in.	22	21/2 in.	**
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l	22	1/16	in.	20	2	in.								
1		1/32	in.	29	21/2	in.	11							
-	-No	tea	r.											

freezing off before completely filling the mold.) Because of oscillation of the gate pattern upon removal and sagging of green cores, the gates varied from 0.043 to 0.125 in. in width. Figure 7 indicates quite clearly that the area of the hot tear increased as the width of the gate increased. It also explains why the weaker mixes in some cases produced tears as large as the strong mix.

Recommendations

The results of this investigation suggest that the casting is potentially suitable for hot tear investigation pending changes in pattern equipment and molding methods. The following recommendations are offered:

1. It is recommended that an entire new core box



be constructed for the cylinder core. Walls of sufficient thickness to overcome sagging should be incorporated into the core design. The draft on the core box should be increased to permit easier removal of the core. The core box should be made of a sturdier material to make it more rigid during molding and removal of cores.

2. Since the gate has a controlling influence on the degree of hot tear resulting in the casting, a more suitable method than that used at present is recommended for producing the gate in the core. It is also recommended that this gate be widened somewhat to allow castings to be poured at a lower temperature than the approximate 2925 F mold entering temperature used in this investigation. It is doubtful that such high temperature metal will be available in case it is desired to pour this casting from a production heat.

- 3. A standard procedure for ramming cores should be established to overcome possible variations in density from core to core.
- 4. Core prints should be incorporated into the gate and runner system to assure accurate alignment of the various parts of the mold.

Details of the Investigation

The pattern equipment for making the cores used in this investigation was obtained from the Armour Research Foundation. The cylindrical cores were rammed with a jolt-rollover-draw machine. Because of the difficulty encountered in drawing the cores, the maximum number of jolts used was severely limited. By sanding and repainting the core box, it was possible to jolt the cores 20 times and draw them somewhat satisfactorily. A standard method was set up of covering the core with a 4-in. head of sand, placing a 30-lb weight on top and jolting 20 times. Some

Fig. 11—Castings made from Mix A showing effect of gate thickness on the size of hot tear. Gate thickness is marked on outer wall of casting.

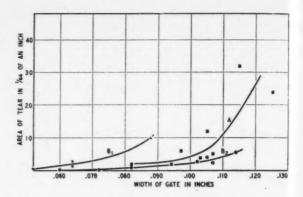


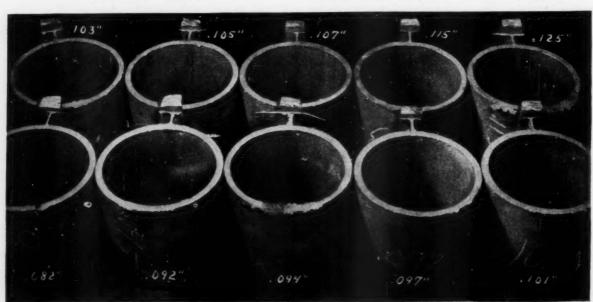
Fig. 12—Effect of gate thickness of Armour Research test casting on hot tear sensitivity of casting using Mixes A and B.

of the variation between our results and those of Armour Research may be attributed to differences in ramming techniques.

Table 5 lists the results of hot tear test castings made using the Armour Research core mixes. The widths of the gates were measured on the baked cores. Most of the castings having average gate widths less than 0.070 in. were misrun.

All heats were poured from Grade "B" steel. Table 6 lists the analyses of those heats made to check the results of the Armour Research sand mixes. All heats were tapped from the furnace at a temperature of 3030 F in order to give an expected mold entering temperature of 2925-2950 F. Castings poured from a lower temperature heat were misrun due to freezing off of the narrow gate.

Table 7 lists the compositions of the three Armour Research core sand mixes and their mechanical properties. Two "B" mixes were made since the gates of the first mix were so small many of the castings were misrun. Although the mechanical properties differed considerably due to high moisture in the



second mix, there was no noticeable difference between the spread of hot tears in the castings from the two mixes. The casting which did not misrun in mix "B" had hot tears varying from $\frac{1}{64} \times \frac{1}{2}$ in. to $\frac{1}{32} \times 5$ in. Those in mix "B2" varied from "no tear" to $\frac{1}{32} \times 2\frac{1}{4}$ in. A comparison of our results with those of Armour Research is listed in Table 8.

Figures 6 and 7 are exploded and assembled views of the cores used for this mold.

Figures 8 and 9 show hot tears in two castings from the same heat using the No. 3 mix for cylindrical core. The four hot tears shown from these two castings are representative of the variable results obtained from the 25 castings made at the beginning of the investigation using No. 3 mix cores, superset bonded cores, and No. 3 mix cores plus an iron oxide addition. As mentioned, the reason for this lack of correlation was found to be the variation in gate size.

Fig. 10 shows three cores with measurements of the gates marked alongside.

Figure 11 shows the entire batch of castings poured from the center cores made with the "A" mix, and the variation in size of hot tears caused by gate variation.

Figure 12 is a graphical representation of the effect of gate variation on the area of the resulting hot tear. The area is only approximate since the width of a tear varies.

PART III- HOT TEAR INVESTIGATION

By

C. H. Wyman,* C. A. Faist** and Geo. Di Sylvestro†

Introduction

The program of investigation carried out by Burnside Steel Foundry Co., Chicago, was initiated in September, 1951, under sponsorship of the Research Committee of the Steel Division of the American Foundrymen's Society. It was desired to show the reproducibility and sensitivity of the test casting developed by the Armour Research Foundation.

It was felt that the test casting produced under production conditions would show more clearly not only the value of this particular design as to reproducibility of results, but also the latitude of sensitivity which might be expected from it.

Therefore it was proposed by the Committee that test castings should be poured at Armour Research Foundation, then at American Steel Foundry Co. (Indiana Harbor, Ind.), and lastly at Burnside Steel Foundry Co. (Chicago). Pattern equipment was furnished by Armour for use at American Steel Foundry Co. and was later sent to Burnside Steel Foundry Co.

Investigation at Burnside Steel Foundry Co.

Following receipt of pattern equipment from American Steel Foundry Co. work was begun at Burnside Steel Foundry in September, 1951. Following recommendations made by American Steel, new pattern equipment was made at Burnside in order to produce cores of greater dimensional accuracy, as well as greater ease in coremaking (Fig. 13). To make the slot gates more uniform the gates were slotted in with abrasive wheel rather than attempting to mold the slot gate in the green cores.

Scope

Following study of the work done at Armour Research Foundation and American Steel Foundry Co. as outlined in Parts I and II, a tentative program of

procedure for the Burnside investigation was drawn up as follows:

A. To investigate the reproducibility and sensitivity of the Armour test casting design.

(I) Pour a series of test castings in triplicate using production methods.

B. To investigate the effects of variation in core and mixes having variable properties (strength, collapsibility, etc.).

(II) Pour a series of castings in triplicate having center cores made from ten different type core sand mixes as well as one series having a center core of green facing sand (1/2 and 1/2).

C. To investigate the effects of variation in the degree of hot spot concentration.

(I) To incorporate in work listed under B a series of castings having variable width slot gates (approximately 1/8, 3/16, 1/4, and 3/8-in. widths and 5-in. length).

D. Investigate effects of metallurgical variables.
 (I) Variation in carbon content of the steel

(Grade B Steel).

(II) Variation in sulphur content (Grade B Steel).



Fig. 13-Pattern equipment for hot tear test specimen

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TABLE 9-CORE MIXES AND PROPERTIES USED IN BURNSIDE INVESTIGATION

"A" Mix	Dry Resin-No	Silica Flou		"D" Mix Dry R	Resin—With Woo	d and Silica	a Flour
Material	Weight, lb	Volume	%	Material	Weight, lb	Volume	%
New Sand	300	25 gal	93.	New Sand	300	25 gal	83.
Truscor	11/4	1/2 gal	0.4	Truscor	11/4	1/2 gal	.1
Super Set	18	2 gal		Super Set	18	2 gal	4.
Release Agent	11/2	l qt		Silica Flour	30	3 gal	8.
		11/2 pt		Release Agent	11/2	l qt	.4
Tycor	31/4	1/2 Pt	0.0	Tycor	31/4		
				Wood Flour		11/2 pt	1.5
					51/4	21/2 gal	1.3
Prop	erties		Core Wght 26 lb, 2 oz	Proj	perties		Core Wght 25 lb, 8 oz
Moisture, %			3.2	Moisture, %			4.6
Permeability			100	Permeability			50
Green Strength	. psi		2.5	Green Strength	n, psi		4.0
Density, lb/cu			100	Density, lb/cu			100
Dry Tensile Str			95	Dry Tensile, p			35
Scratch Hardne			45	Scratch Hardn			80
			0.0250	Deformation,			0.0250
Deformation, in	at 2000 F, 12 Mi	n nsi	14.0		at 2000 F, 12 Mi	n nsi	30
			* *10				
	"B" Mix—1% C				y Resin—0.25%		
Material	Weight, lb	Volume	%	Material .	Weight, lb	Volume	%
New Sand	300	25 gal	81.	New Sand	300	25 gal	85.
Truscor	33/4	11/2 gal	1.	Truscor	11/4	1/2 gal	.1
Super Set	9	1 gal	2.	Super Set	18	2 gal	4.
Core Oil	4	1/2 gal	1.	Silica Flour	30	3 gal	8.5
Silica Flour	50	5 gal		Release Agent	11/2	1 qt	0.4
No Vein	21/2	l qt		Tycor	31/4	11/2 pt	0.9
NO VEIII	-72	. de	0	Coralux (Coarse)	574	172 PC	0.2
Prop	erti e s		ore Wght	Prop	perties		ore Wght
36-1-1			4.0	Maistura 01			4.0
Moisture, %			75	Moisture, % Permeability			65
Permeability			2.3		. mat		3.1
Green Strength,				Green Strength			
Density, lb/cu i			100	Density, lb/cu			100
Dry Tensile, ps			70	Dry Tensile, ps			35
Scratch Hardne	SS		85	Scratch Hardne			80
Deformation, in	./in.		0.0200	Deformation, i			0.0250
Hot Strength a	t 2000 F, 12 Min	, psi	38.0	Hot Strength a	t 2000 F, 12 Min,	psi	110
66	C" Mix—1/2% C	Dil		"F" Mix Liquid C	atalin No. 8818 I	Resin—Silie	ca Flour
Material	Weight, lb	Volume	%	Material	Weight, lb	Volume	%
New Sand	300	25 gal		New Sand	300	25 gal	87.5
Truscor	5	2 gal		Truscor	21/2	l gal	0.8
Core Oil	2	1 qt		Liquid Resin No. 8818	21/4	1 qt	1.
No Vein	21/2	1 qt	0.8	Silica Flour	30	3 gal	8.
				Tycor	31/4	11/2 pt	0.8
				Release Agent	11/2	1 qt	0.4
Prope	erties		ore Wght	Properties	-	C	ore Wght
		20	6 lb, 1 oz				26 lb
Moisture, %			3.0	Moisture, %			4.2
Permeability			20	Permeability			90
Green Strength,	psi	T	oo Weak	Green Strength	, psi		1.8
Density, lb/cu f	t		To Run	Density, lb/cu	ft	1	00
			Green	Dry Tensile, ps	i	1	70
Dry Tensile, ps	i	1	80	Scratch Hardne	SS		85
			55	Deformation, in			0.0200
Scratch Hardner							
Scratch Hardner Deformation, in				Het Strength at	2000 F. 12 Min r	281	34
Deformation, in		nei .	20.5	Hot Strength at	2000 F, 12 Min, I	OS1	34

(a) Pour series in duplicate desulphurized to different sulphur levels with CaSi alloy.

(III) Variation in composition of steel (Grade B vs. Mn-Cr).

E. To investigate the effect of variation in wall thickness of test casting.

(I) Pour series of castings in triplicate from each of three different core mixes, the castings having constant gate width but variable cyl-

TABLE 9 (CONTINUED)

"G" Mix-Dry Ca	talin No. 9616	Resin—Silica F	"I" Mix—Liquid No. 8818 Resin—No Silica Flour						
Material	Weight, lb	Volume	%	Material	Weight, lb	Volume	%		
New Sand	300	25 gal	88.5	New Sand	300	25 gal	97.		
Truscor	21/2	1 gal		Truscor	21/2	1 gal	0.8		
Dry Resin No. 9616	21/4	1 qt		Liquid Resin No. 8818	11/8	1 pt			
Silica Flour	30	3 gal	8.0	Release Agent	11/2	1 qt	0.4		
Tycor	31/4	11/2 pt	0.8	Tycor	31/4	11/2 pt	0.8		
Release Agent	11/2	1 qt	0.4		,-				
Prope	rties	Core 2	Wght lb, 6 oz	Proper	Core Wght 27 lb, 4 oz				

Properties	Core Wght 26 lb, 6 oz	Properties	Core Wght 27 lb, 4 oz
Moisture, %	4.0	Moisture, %	3.2
Permeability	105	Permeability	120
Green Strength, psi	1.9	Green Strength, psi	1.2
Density, lb/cu ft	100	Density, lb/cu ft	100
Dry Tensile, psi	60	Dry Tensile, psi	52.5
Scratch Hardness	60	Scratch Hardness	40
Deformation, in./in.	0.0170	Deformation, in./in.	0.0300
Hot Strength at 2000 F, 12 Min, psi	12	Hot Strength at 2000 F, 12 Min, psi	4

"H" Mix-[4.0% Dry-0.5% Liquid] Resins + 1.5% Wood Flour

"J" Mix-1% Oil-1.5% Wood Flour

Material	Weight, lb	Volume	%	Material	Weight, lb	Volume	%
New Sand	300	25 gal	84.	New Sand	300	25 gal	80.
Truscor	1/9	3/4 gal	0.1	Truscor	33/4	11/2 gal	1.
Super Set	18	2 gal	4.	Super Set	9	1 gal	2.
Silica Flour	30	3 gal	8.0	Core Oil	4	1/2 gal	1.
Liquid Resin No. 8818	11/6	1 pt	0.5	Silica Flour	50	5 gal	13.5
Release Agent	11/2	1 qt	0.4	No Vein	21/2	1 qt	0.7
Wood Flour	51/4	21/2 gal	1.5	Wood Flour	51/2	23/4 gal	1.5

Properties	Core Wght 26 lb, 8 oz	Properties	Core Wght 24 lb, 8 oz
Moisture, %	4.0	Moisture, %	4.2
Permeability	80	Permeability	30
Green Strength, psi	3.3	Green Strength, psi	3.9
Density, lb/cu ft	100	Density, lb/cu ft	100
Dry Tensile, psi	45	Dry Tensile, psi	70
Scratch Hardness	60	Scratch Hardness	65
Deformation, in./in.	0.0300	Deformation, in./in.	0.0300
Hot Strength at 2000 F, 12 Min, psi	3	Hot Strength at 2000 F, 12 Min, psi	111

Blended Facing Sand [Green]

				Properties						
Material	Weight, lb	Volume	%		Minimum	Maximum				
New Sand	1000	78 gal	48	Moisture, %	2.9	3.4				
Old Sand	1000	78 gal	48	Density, lb/cu ft	97	100				
Bentonite	62	81/2 gal	3	Permeability	150	180				
Truscor	18	71/4 gal	0.8	Green Strength, psi	7.5	8.0				
N. V. X.	21/2	3/4 gal	0.1	Deformation, in./in.	0.280	0.0300				

inder wall thickness (1/2 in. and 5/8 in.).

F. To investigate the hot properties of the various type core mixes with respect to hot tear produced.

G. To study, correlate if possible, and to report such other data and information which could be associated with hot tear phenomena.

Procedure

The procedure for investigating steps A through C was such that these phases could be investigated simultaneously. A total of ten core sand mixes were incorporated. A list of these mixes will be found in Table 9 with their compositions and properties.

After baking these center cores were slotted with abrasive wheels to give slots of variable width $\frac{1}{8}$, $\frac{3}{16}$, $\frac{1}{4}$, and $\frac{3}{8}$ in., all 5 in. in length (Figs. 14, 15).

The cores were then assembled as shown in Figs. 6 and 7 and the core mold then locked into pouring jig shown in Fig. 16. This lock-up arrangement was found very satisfactory in expediting the pouring and shakeout of experimental castings by reducing shifts, run-outs, and cracked cores in handling of assembled molds. A series of such lock-ups is shown ready for pouring in Fig. 17.

All castings were poured in series and at the same position in the pouring cycle of the heat. They were hand-shanked over the lip of a 300-lb capacity monorail ladle. No temperature recordings were made, but previous pouring cycle temperature measurements have disclosed that practically all shank heats poured in this portion of the pouring cycle had a pouring



Fig. 14—(Left) and Fig. 15 (Right) showing slotting operation with abrasive wheels.

temperature in the range of 2840 to 2910 F. Previous work on hot tears done by other investigators had indicated that on simple design castings of small weight pouring temperature had little bearing on hot-tear incidence.

Specifically, each mix was divided into three pouring series, each series being poured off on a different Grade B Steel heat. Each series contained within it four gate width variations (1/8, 3/16, 1/4, and 3/8 in.). Therefore, a total of 12 castings (24 cylinders) was poured for each type of core sand mix.

After being poured and shaken out (pour to shakeout time held to 5-10 min) the castings were hand shotblasted and then magnetic particle inspected. Inspection consisted of measuring the actual gate length, cylinder length, tear length, gate width, core weight, casting weight, etc. These data were then recorded. Next a duplication of the actual tear obtained in each cylinder was made by picking off the magnetic particle indication on a strip of Scotch tape and recording these tapes with the data. In this manner fairly accurate tear recordings could be made for future work, once the castings had been scrapped.

Figures 18 and 19 show the castings after shotblast and inspection.

In addition to the ten core mixes a series of castings was made using a green sand core (core box alterations are shown in Fig. 20).

Fig. 16—Pouring jig.

After completion of the first phase of the investigation a general meeting of the Research Committee of the Steel Division of A. F. S. was held at Burnside Steel Foundry Co. for the purpose of reviewing the work done at Burnside. It was generally agreed by committee members after hearing Burnside's report that a second phase be run jointly by Armour, American Steel Foundry Co., and Burnside to further substantiate the validity of the Burnside results and data.

For purpose of clarification this report therefore refers to Phase I as that phase which included work done prior to the committee meeting at Burnside and to Phase II as that phase which includes all work subsequent to the meeting.

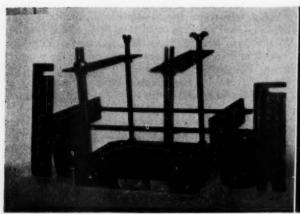
Phase II therefore was begun at Burnside for the primary purpose of establishing the validity of Phase I data, i.e., duplication of Phase I results.

It was believed that this could be done most economically by limiting Phase II investigations to duplication of one particular gate width and thus reduce the total number of castings. It was further decided that all ten core sand mixes would be run in duplicate series rather than triplicate series, still further reducing the number of castings to be poured.

All series of castings of Phase II were poured and processed in the same manner as Phase I series.

Following the duplication series it was decided to investigate the effect of metal composition on hot-

Fig. 17-Series of lock-ups ready for pouring.





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tear incidence. It was planned to run a duplicate series from a medium manganese-chromium steel for comparison to the previously poured Grade B Steel castings.

Next a series was run to determine the effect of sulphur level on hot-tear incidence. For this purpose two different heats of Grade B Steel were used. A series of ten core mixes was run on each heat. The steel going into these castings was desulphurized using a special Ca-Si alloy in addition to the regular standard deoxidation practice (Al, 21/2 lb/ton and Ca-MnSi, 3 lb/ton).

Finally, three series were run in duplicate to determine the effect of cylinder wall thickness variation on hot-tear incidence. For this portion of the investigation three different resin core sand mixes, and three oil sand core mixes were used. Each mix was poured in a series of castings have three different cylinder wall thicknesses (3/8, 1/2 and 5/8 in.).

After pouring, inspection and recording of data, representative castings from each of the three different wall thicknesses were sectioned so that soundness and macrostructure of the castings could be observed (Figs. 21-23).

Results Obtained

During the initial portion of Phase I of the investigation it became apparent almost immediately that, except for isolated, and minor cases, the experimental casting as designed by Armour Research Foundation was an instrument for producing, under a specific set of conditions, a hot tear which was both

reproducible and sensitive. Furthermore, it was heartening to observe that this reproducibility was valid under the more severe test conditions of actual foundry practice.

Survey of the data accumulated on the series of castings poured in triplicate in Phase 1 showed that hot-tear incidence was reproduced with very gratifying results. Figure 24 shows the actual tear reproduction obtained during Phase 1, and shows that hottear incidence varies with ingate cross-sectional area increase (all gates approximately 5 in. long by varying widths).

It should be noted particularly that the increase in gate cross-sectional area can be presumed to be to a degree an increase in the degree of hot-spot concentration. It is interesting to note that for all core mixes (oil and resin) having no wood flour additions, as the gate area increased so the tearing increased, showing a steady resistance to the solidifying metal. In the case of the three mixes with wood flour (H, I, and D) the hot tear increases normally with the increase in gate area, but then decreases with further increase, indicating that these cores collapsed at intermediate hot-spot concentration (see H, J, and D, Fig. 24, and Fig. 25).

The Phase II series data showed that the trend obtained for hot-tear incidence corresponded to Phase I, although the actual values of tear differed somewhat (Fig. 26). Examination of data from both phases showed that the only marked difference between the two was the difference in the core densities (due to difference of moisture content of mixes resulting in

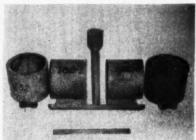


Figure 18



Figure 19

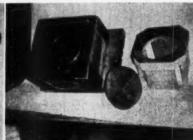


Figure 20



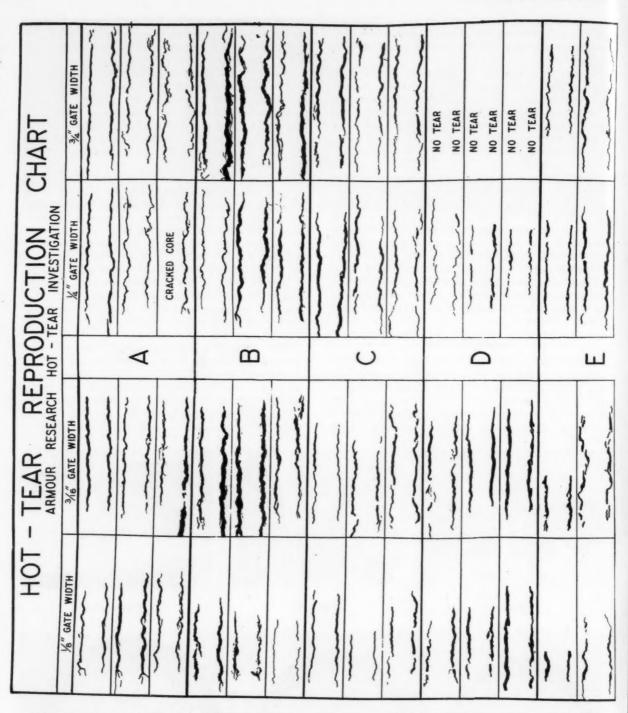
Figure 21







Figure 23



variable ramming properties). When differential densities were plotted against differential hot tear a very good relationship was found, indicating that the density variations of the cores used in the two phases were to a large extent responsible for the discrepancy in the hot-tear incidence (Fig. 27).

When Phase I and Phase II were corrected for core density variation it was found that tear incidence of both phases were practically identical. This is shown on Fig. 28.

It was gratifying to find that core density variation

between cores of the same type of mix appeared to be the most important single factor in determining hot tearing incidence when a number of castings were poured using a specific type core mix.

Since the core density correction factor worked out so well with Burnside data, it was decided to check outside data for purpose of substantiating the validity of this correction factor.

Various data on hot tear length and core density was requested from Armour Research Foundation from data obtained during their investigation.

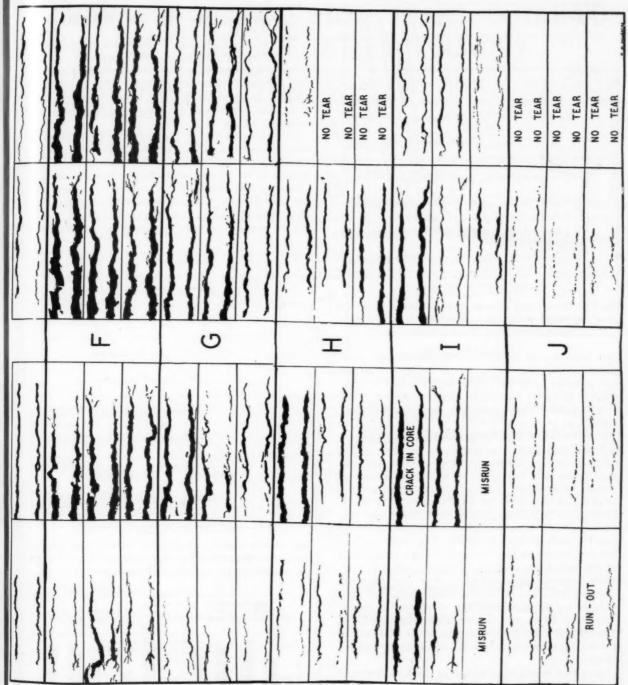


Fig. 24—Hot tear reproduction chart reduced about half in printing.

When the Burnside correction factor was applied to their data it was found that the difference between maximum and minimum hot tear lengths for any particular gate area, and any specific core sand mix was nullified when corrected for density variation. That is to say, for example, any minimum hot tear value corrected for the difference between minimum and maximum core densities took on the value of the maximum hot tear length, i.e., accounted for hot tear length variations.

o g e

It was therefore assumed that the density correc-

tion factor was valid.

Following the pouring of the series of castings to check the validity of Phase 1 results as previously reported to the Research Committee, work was begun to determine, if possible, whether metal composition had any effect on hot-tear incidence.

For this portion of the investigation it was decided to pour a series of castings in duplicate from a medium manganese-chromium steel to compare with hot-tear incidence of Grade B Steel.

Only one size of slot gate was used, although cast-

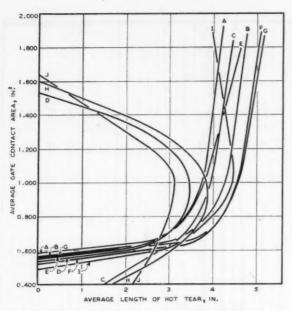


Fig. 25-Relationship between cross-sectional gate contact area and length of hot tear for various type core sand mixes investigated. Grade B Steel used.

ings were poured using the ten different core sand mixes. A chart showing the type and extent of hot tear obtained with each core mix is shown in comparison to Phase I Grade B Steel tears on Fig. 29. It will be noted that, although the tears are reproducible as to type of steel, there is no comparative correlation to be found between the two grades.

It was next decided to investigate the effect of the sulphur level on hot-tear incidence. Again ten core sand mixes were used and two series of test castings poured in duplicate. Each of the two series was desulphurized to a different sulphur level. The desulphurization was accomplished through the use of a calcium-silicon alloy addition to the steel prior to pouring each of the two series. The results of this test can be seen in Fig. 30 which shows a comparison of the two series, each at a different sulphur level. Aside from the reduction of sulphur level all other elements remained at approximately the same percentage weight level with the exception of silicon. The base metal was raised approximately 0.27 Si (weight per cent) by the calcium-silicon addition to a level of approximately 0.74 per cent (by weight) silicon.

As a final test it was decided to check the effect of varying the cylinder wall thickness of the test casting on hot-tear incidence. For this phase of the investigation it was concluded that the gate size should be held constant. Furthermore, instead of pouring ten core sand mixes it was decided to use three oil sand core mixes of varying strengths and three resin core sand mixes of varying strengths. The variations in these mixes is listed as follows:

Oil Sand Cores:

- 1. 1/2 pct oil (same as Phase 1 "C" mix)
- 2. 1/2 pct oil plus 1/8 pct fine Perlite
- 3. 1/2 pct oil plus 11/2 pct Wood Flour

Resin Sand Cores:

- 1. Same as Phase 1 "I" mix plus 51/2 pct Dry
- 2. Same as Phase 1 "I" mix plus 51/2 pct Dry
- Binder plus 1/8 pct fine Perlite
 3. Same as Phase 1 "I" mix plus 51/2 pct Dry Binder plus 11/2 pct Wood Flour

Figure 31 shows the hot tear reproductions of the 1/2 in. and 5/8 in. wall thickness cylinder results as compared to the 3/8-in. cylinders.

It can be seen that the hot-tear incidence becomes lesser in degree with increase in cylinder wall thickness (constant slot gate cross-sectional area). This appears to be in line with the theory that, as the volume of the casting in the immediate vicinity of the concentrated hot spot area increases (due to progressive increase in wall thickness from 3/2 to 5/2 in.) the heat concentration is spread out or widened.

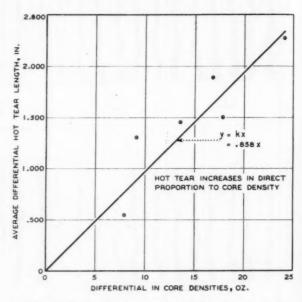


Fig. 27-Relationship which exists between core density variation and resultant hot tear variation. Valid for all mixes except mix which contained "coralux."

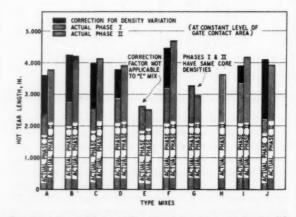
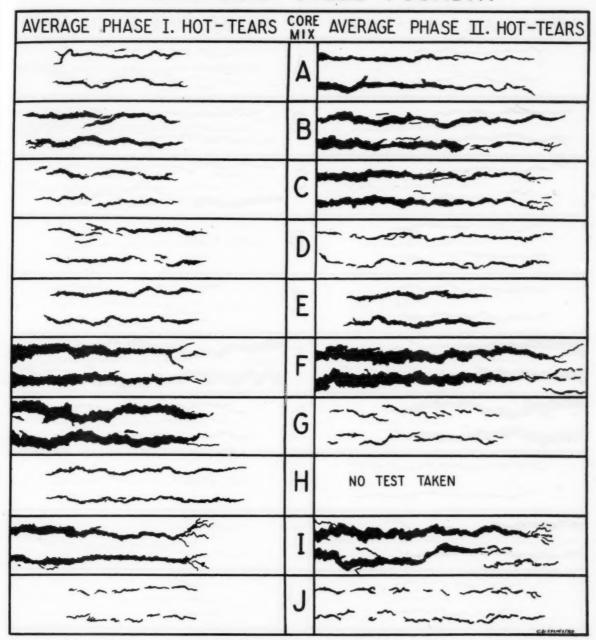


Fig. 28-Comparison of length of hot tear obtained with various type mixes of Phase I and Phase II at constant level of gate contact area.

CHART OF ACTUAL HOT-TEARS OBTAINED AT BURNSIDE STEEL FOUNDRY



GRADE B STEEL

AVERAGE SLOT-GATE APPROX. 3/16 IN. X 5 IN. WITH 3/8 IN. CYLINDER

Fig. 26—Chart of actual hot tears obtained at Burnside Steel Foundry. Chart was reduced about one-fourth in reproduction.

CHART OF ACTUAL HOT-TEARS OBTAINED AT BURNSIDE STEEL FOUNDRY

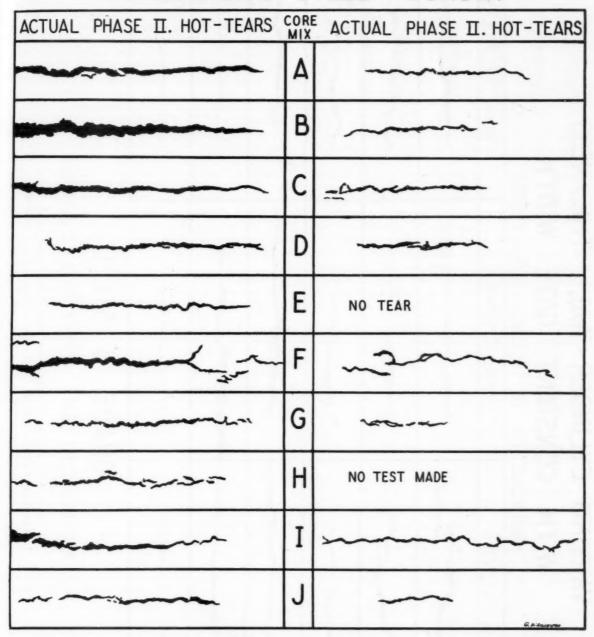
AVERAGE PHASE I. HOT-TEARS	CORI	AVERAGE PHASE II. HOT-TEARS
y min	Δ	はなるると
	В	- management
	С	
	D	NO TEAR
	Ε	-2~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~
	F	いい
	G	~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~
	Н	NO TEAR
	I	
~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~~	J	~ 6.818741817.00

GRADE B STEEL MN-CR STEEL

AVERAGE SLOT-GATE APPROX. 3/16 IN. x 5 IN. WITH 3/8 IN. CYLINDER

Fig. 29-Chart of actual hot tears obtained at Burnside Steel Foundry. Chart reduced about one-fourth in reproduction.

## CHART OF ACTUAL HOT-TEARS OBTAINED AT BURNSIDE STEEL FOUNDRY



SULPHUR LEVEL .045 WGT.

SULPHUR LEVEL .039 WGT.

## GRADE B STEEL

AVERAGE SLOT-GATE APPROX. 3/16 IN. X 5 IN. WITH 3/8 IN. CYLINDER

Fig. 30—Chart of actual hot tears obtained at Burnside Steel Foundry. Chart reduced about one-fourth in reproduction.

# VARYING CYLINDER WALL THICKNESS WITH CONSTANT GATE WIDTH REPRODUCTION OF ACTUAL HOT - TEARS

5/8 IN. CYLINDER	Valle: CICINDEN	- Acide	- James				
1/2 IN CYLINDER	12 IN. CILINDEN		- Anna Contraction of the Contra	- manual.	- market		·
3/s IN CYLINDER	% IN. CILINDEN	The state of the s	- manufacture of the same	- Astronomy and serve		wow of the state o	
XIX	V W	RESIN	RESIN	ВЕЗІИ	4	0IF	9

# GRADE B. STEEL

AVERAGE SLOT - GATE APPROX. 3/16 IN. X 5 IN. LONG

Fig. 31-Varying cylinder wall thickness with constant gate width. Reproduction of actual hot tears. Chart reduced about one-third in printing. This would cause the hot spot area to remain in the critical hot tearing temperature range for shorter intervals and therefore produce a lower frequency to tear.

The chart (Fig. 31) shows secondarily that as either of the oil or resin sand core series decreases in core strength (strongest to weakest mix) the hot tear incidence tends to decrease. This of course was found to be so in previous tests run during the investigation.

Figure 32 shows a comparison of average hot tear length values obtained for the 3/8, 1/2 and 5/8 in. cylinder wall thicknesses.

Table 10 represents the six basic core sands used for the tests on variable wall thickness cylinders. Three resin mixes and three oil mixes are represented.

For the purpose of studying the behavior of core sand hot properties on hot-tear incidence a series of tests of each core sand mix used during the investigation was made. These sample cores were made up at Burnside Steel Foundry and sent to the Harry Dietert Co, Detroit, in sealed containers for hot property testing. The hot properties run included hot strength, collapsibility and hot deformation. The test results were then sent back to Burnside Steel Foundry Co. for analysis.

Analysis of the hot property results showed that there was not particular relationship between hot strength (12 min at 2000 F), collapsibility (85 psi

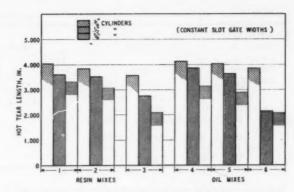


Fig. 32—Comparison of hot tear lengths of  $\frac{1}{2}$ ,  $\frac{1}{2}$  and  $\frac{1}{2}$  in. cylinder wall thicknesses. Constant slot gate widths.

at 2000 F) and hot deformation at 2000 F and the average tear incidence produced by any particular core mix. It appears that these properties individually could not be used as a criterion for determining hot tear behavior, or incidence.

Finally the various core mixes were arranged in ordered arrangement under each of the various hot properties, i.e., the various mixes were arranged separately under three headings (hot collapsibility, hot strength and hot deformation). The arrangement under each heading was done in such a way that

TABLE 10-Core MIXES FOR 1/2 IN. AND 5/8 IN. CYLINDERS

		l. Resin			. Resin			3. Resir			3. Oil o Silic	a		4. Oil ne Perli	ite	W	6. Oil ood Flo	our
Material	Wt.	Vol.	%	Wt.	Vol.	%	Wt.	Vol.	%	Wt.	Vol.	%	Wt.	Vol.	%	Wt.	Vol.	%
New Sand	300	25 gal	93	300	25 gal	92	300	25 gal	92	300	25 gal	96.5	300	25 gal	96	300	25 gal	96
Truscor	11/4	1/2 gal	0.4	11/4	1/2 gal	.4	11/4	1/2 gal	0.4	5	2 gal	1.5	5	2 gal	1.5	5	2 gal	1.5
Super Set	18	2 gal	5.5	18	2 gal	5.5	18	2 gal	5.5									
Liq. Resin No.8818	11/2	11/2 pt	0.5	11/2	11/2 p	.5	11/2	11/2 pt	0.5									
Tycor	31/4	11/2 pt	0.8	31/4	11/2 pt	.8	31/4	11/2 pt	0.8									
Core Oil										2	1 qt	0.6	2	1 qt	0.6	2	1 qt	0.6
No Vein										21/2	lqt	0.8	21/2	1 qt	8.0	21/2	lqt	0.8
Fine Perlite				4	8 gal	.0012					•		37/8	31/2 gal	.0012			
Wood Flour							5	21/2 gal	1.5					7				
Release Agent	11/2	1 qt	0.4	11/2	1 qt	.4	11/2	1 qt	0.4							5	21/2 gal	1.5
Water			3			3.5			4.5			3			3.5		0	4.5

TABLE 11-METHOD OF CORRELATING HOT TEAR TO HOT PROPERTIES

4			of Decreasing Hot I	
Mixes Arranged in Order of Decreasing Average Tear Lengths		Collapsibility (Sec. @ 2000°F) Order	Hot Strength (psi @ 2000°F) Order	
Type Mix	Tear Length, in.	No.	No.	No.
F	4.550	1. F <u>=</u> 66	1. B=38	1. F=.024
В	4.220	2. I=57	2. F=34	2. B=.018
I	4.150	3. B=17	3. A=14	3. A=.018
C	4.000	4. A=12	4. I= 4	4. C=
A	3.750	5. C= 2	5. $C = 0$	5. I=-
Method of Arranging			nparison of Average	
to Arrive at Overall ( Indicating Decreasing Ho		WILL	Arranged Average of Hot Properties	
Individual Ord	er Overall Order	Hot Tear		Hot Property
Mîx Number	Number	Order		Order
F = 1 + 2 +	1 = 4	F = 4.550 =	= 1st	F = 1st
B = 3 + 1 +		B = 4.220 =	= 2nd	B = 2nd
I = 2 + 4 +	4 = 10	I = 4.150 =	= 3rd	I = 3rd
	3 = 10	C = 4.000 =	= 4th	A = 4th
C = 5 + 5 +	4 = 14		= 5th	C = 5th

TABLE 12

Mixes Ar- ranged in		ge Tear Lo n Cylinders		O	Order of Mixes for Hot I'roperties						
order of Increasing tear length	Wall	Thickness,	in. 5/8	Total Deform in./in.	Rate of Deform in./in.	Collap- sibil- ity	Hot Strength	of properties 1.	Density 2.		
A	3.575	2.750	2.100	A	A	В	С	A	A		
Resin C)	3.850	3.530	3.080	C	В	A	A	C)	B)		
BĴ	4.050	3.600	3.320	В	C	C	В	B)	c}		
D	3.850	2.150	2.080	D	F	D	D	D	D		
Oil F)	4.040	3.650	2.900	E	D	E	E	E)	F		
E	4.130	3.880	3.150	F	E	F	F	F)	E		

the mix with the particular property valuation indicating highest tendency toward hot tear was at the top, and the least tendency toward production of hot tear at the bottom. Five mixes were so arranged and assigned order numbers, 1 (highest tear tendency) through 5 (lowest tear tendency). This was done for all three hot properties. Next the order numbers were totaled for all three properties. This total represented the average overall order, or position for the various mixes with respect to the combined influence of individual hot properties on hot-tear incidence.

Table 11 shows the overall averages used above to correlate the hot-tear incidence with the average weighted hot properties of the various mixes.

Further attempt was made to correlate the hot properties of various mixes investigated, to the resultant hot tear lengths obtained when these mixes were applied in casting of the Armour hot tear test casting.

Hot property specimens of the same mixes were again made up and submitted to the Harry W. Dietert Co. in sealed containers. Tests conducted on these specimens included hot strength, collapsibility, total

hot deformation and rate of hot deformation. Each type of test was conducted at 600, 800, 1000, 1200, 1400 and 1600 F. Tests previously submitted had been tested at 2000 F.

After receipt of the above test data from the Harry W. Dietert Co., the data were plotted at Burnside Steel Foundry Co. in an attempt to correlate the various data to hot tear lengths obtained during the Burnside investigation of hot tear.

All attempts to correlate hot properties to hot tear results were very disconcerting. In some cases, value relationships appeared, but nothing absolute could be evolved.

Finally all data were arranged in order according to an assumed tendency toward hot tear influence. This is explained by examination of the Table 12. It will be noted that the table is divided into two sections, one for resin core mixes and one for oil core mixes. The mixes are the same as those reported in Burnside hot tear report to A.F.S. In the first column are the mixes arranged in order of increasing order of tear length. The second, third and fourth columns indicate the average tear lengths obtained with each core mix during the Burnside tests for \( \frac{1}{8} \).

TABLE 13—METHOD OF ARRIVING AT ORDER NUMBERS
Only Hot Strength Taken As Example. Other Hot Properties Done Same Way.

Hot	Streng	th Data:	rious Test	Temp & 12-	Min Scak (	psi)		E	I o. Tir ach M Appea	lix	II *No. of Times in position	III Sum of
1	Mix	600 F	800 F	1000 F	1200 F	1400 F	1600 F		t 2d		× Pos. Mult.	ducts
	A	(1) 473	(3) 520	(1) 227	(2) 104	(2) 100	(3) 139	A 2	2	2	$2 \times 1 + 2 \times 2 + 2 \times 3 =$	-12
	B	(2) 440	(2) 340	(3) 202	(1) 102	(1) 87	(1) 87	A S B S	2 2	1	$3\times1+2\times2+1\times3=$	10
	C	(3) 355	(1) 370	(2) 225	(3) 115	(3) 132	(2) 122	C			$1 \times 1 + 2 \times 2 + 3 \times 3 =$	14
	D	(3) 60	(3) 61	(3) 28	(3) 8		(2) 0	D (	0	5	$0 + 0 + 6 \times 3 =$	18
	E	(1) 122	(2) 130	(2) 36	(2) 16	(3) 2 (1) 9	(3) 0	D C	4	0	$2 \times 1 + 4 \times 2 + 0 =$	10
	F	(2) 100	(1) 170	(1) 65	(1) 20	(2) 8	(1) 19	F 4	1	1	$4\times1+1\times2+1\times3=$	9
Posi fixe of E	s Arra Decreas lue of easing	(2) Interm (3) Lowest anged in (5) sing Nume (5) Column I	ned. value - t values - Order erical	-Intermed.   -Lowest Ho	Hot Tear T	endency —	Multipliers Multipliers Multipliers	= 2	٠			
	A- B-	_12 _10	OFF. 1				NOTE: 0					
	E-	_10 Th	OTE: 1 is column i			1					12 are then treated by a	
	-1	stre	reported in ngth colum ole 12.			i					in column 1 of Table in order of increasing 1	

1/2, and 5/8 in. cylinder wall thicknesses. The next four columns represent the order in which the various mixes fell (in order of assumed tendency toward hot tear length increase) when the 600 through 1600 F rest data were averaged as per subsequent explanation.

The last column represents the overall order in which each of the various mixes fell, when averaged

out for combined hot property value.

Table 13 shows the method used for obtaining the order of the mixes under any particular hot property in Table 12. This method was used for all order columns and also for obtaining the overall order of mixes in column 1, Table 12.

It can be seen that this method appears to give some rather definite correlation between the various core mixes and their individual hot tear length values.

The relationship of core density to hot tear tendency shown in column 2 of Table 12 is in good agreement with results previously reported in the Burnside hot tear report; that is to say, hot tear tendency is greater with increase in core density (independent of core mix ingredients).

On the basis of experimental data obtained and through use of the method of arranging mixes in order as presented herein, the following conclusions

are:

1. The combined influence of hot tear properties when taken collectively, correlate with the degree of hot tear obtained during test period as shown below:

Mixes Ar in incre order of tear le	asing f hot	Mixes Arranged by averaging method in order of increas- ing hot tear length	Mixes Arranged in order of increasing core density (relative)
	Λ	A	A
Resin	C	C)	В
	В	B) Same	C
	D	D	D
Oil	F	E	F
	E	F	E

2. The density of the core apparently plays a large part in determining the degree of hot tear, the tendency being for an increase in tear with increase in core density, regardless of the mix constituents.

In conclusion, it should be noted that there are very small increments of tear length difference between mixes C and B, and, F and E. This is reflected in columns 1 and 2 of Table 12 where the order of C and B are the same, and where E and F are nearly the same value. Note brackets denoting this fact,

During the overall investigation a comprehensive data sheet was kept on all castings poured. These data later enabled various factors to be studied which were thought to have possible bearing on hot-tear incidence.

It must be kept in mind, however, that the program of investigation carried out at Burnside Steel Foundry Co. had as its primary aim (1) to show reproducibility and sensitivity of the test casting in question and (2) to show the effect of various core sand mixes having a wide variation in properties.

Therefore any data which were accumulated aside from the above primary scope of the program were used only to show relationships of a secondary nature

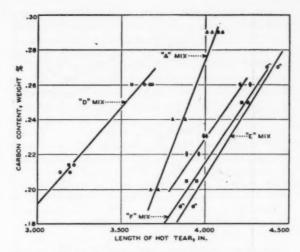


Fig. 33—Trend chart showing apparent influence of carbon level on degree of hot tear.

insofar as factors having to do with hot-tear incidence were concerned.

The factors listed are those which could be checked with actual accumulated data and are metallurgical in nature. Certain reservation must be applied to the validity since as stated previously, the metallurgical factors were of secondary importance to this investigation. The charts pertaining to these factors are submitted only for the purpose of showing that certain of these factors seem to show trends in influencing hot-tear incidence. To prove or disprove the influence on hot tears of any of these factors would undoubtedly involve a substantial and complete investigation entirely outside the scope of the work done at Burnside.

Interesting trends, however, are suggested on Fig. 33 showing the effect of carbon level on hot tear incidence; Fig. 34 showing the influence of sulphur level on hot tear incidence, Fig. 35 showing effect of quality rating ("P" value) on hot tear incidence, and lastly the lack of relationship between total residual aluminum content (in range of 0.06 to 0.075 weight per cent) on hot tear incidence (see Fig. 36).

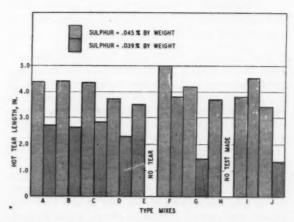


Fig. 34—Effect of sulphur level on length of hot tear.

Grade B Steel.

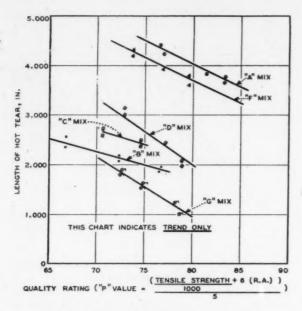


Fig. 35—Chart showing relationship between hot tear length and quality rating for various type mixes of core sand. This chart indicates trend only.

These relationships are presented in this report to stimulate possible future work in evaluating these factors more completely.

## Analysis of Results

The investigation, as it was carried out at Burnside Steel Foundry Co., showed that under conditions of actual production the test casting in question gave results which were both reproducible and sensitive. In the beginning of the program it was found that various conditions had to be controlled closely to give repetitive results. The more important of these were:

- 1. Well made pattern equipment (to prevent box distortion in ramming).
- 2. Well matched cores, when core mold was assembled.
  - 3. Uniformity in width and length of slot gate.
  - 4. Uniformity in ramming of cores (center cores).
  - 5. Uniformity in choke control during pouring.
  - 6. Uniformity of metal analysis.
  - 7. Uniformity of core sand mix.

The statement "well made pattern equipment" refers to equipment which will produce cores free from distortion and with free drawing properties. It was found that cores had to be well matched in assembly, since unmatched or ill-fitting cores produced fins or run-outs in pouring. Fins particularly affected the reproducibility in that a fin caused localized increase in cooling rate of the casting at that point and resulted in a variation of tear incidence. Since the length and width of the slot gate was directly a measure of the hot spot concentration it is readily understood that the degree of hot tear would vary with ingate dimensional tolerances. From Phase I results it was apparent that unless a specific degree of choke was maintained, the hot tear reproductibility would

be adversely affected, i.e. the greater the variance from a choked system the more variable became the reproducibility. It was noted also that variation in moisture content of a core sand mix changed the rammability of the cores produced and subsequently caused variation in core density. This was found to greatly affect hot-tear incidence. Finally, variation in metal composition, primarily in sulphur level and carbon content, seemed to affect hot tear reproducibility.

The above statements are made in an effort to show some of the more important factors which must be accounted for in pouring of reproducibility tests.

The results of tests made in Phase II of this report show that the Phase I results as reported to the Research Committee of the Steel Division of A.F.S. are valid. It should be remembered that variation in length of hot tear between the two phases can be accounted for when variation in core densities is considered.

It was interesting to note that hot-tear incidence increased as the effective degree of heat concentration increased. Since the effective cross-sectional gate contact area is the prime factor in determining the hot spot concentration, it has become habit to speak of these two terms synonymously. It must be considered, however, that unless a degree of choke is provided in the gating system, the relationship of effective ingate area as a measure of hot spot concentration will diverge. This is amply shown by the greater divergence in reproducibility when comparing reproducibility of each of the various slot gate width results in Phase I.

## Hot Tear Incidence

The larger the slot gate width the less accurately is reproduced the hot tear incidence. This was also shown in Fig. 25 where it is seen that as the ratio of choke becomes less the slope of the curves for the various mixes changes, the most abrupt change occurring at the point where the ratio of choke becomes less than unity, indicating an open system. At this point and beyond the degree of tear incidence increases at decreased rate, again indicating that the hot spot concentration in the immediate vicinity of the contact area is somewhat lessened. Another way of looking at this would be to consider that the more open the system the more intermittent the continuity of flow of the incoming metal and therefore the greater the variance in degree of heat concentration.

It may further be pointed out, on Fig. 31, that the hot tear begins further to the right (away from the entrance of gate). This is accountable for in view of the increased internal volume of the mold cavity of the ½ in. and ½ in. wall thickness cylinders. Since the incoming metal enters a larger mold cavity (than in the case ¾ in. cylinders) the metal is less restricted in flow (inside mold) and therefore the volume of incoming metal has a natural tendency to flow more from the farthest end of the gate. Therefore, for constant gate width the hot tears in the cylinders would be expected to originate nearer and nearer the end of the gate most remote from the sprue as the cylinders increased respectively in wall thickness. In

checking over the charts of hot tear reproduction for the 3/8, 1/2 and 5/8 in. cylinders (Fig. 31) this appears to be the case. Finally, note that the general tendency for the tears to begin farther away from the sprue end of the gate indicates a shift of the hot spot with increase in cylinder wall thickness.

Survey of accumulated data showed that rate of pouring of the test casting had no apparent influence on hot-tear incidence, even though pouring times varied all the way from 4 sec to 15 sec (corresponding to pouring rates of 10.7 lb/sec to 2.8 lb/sec). It was believed that the large total surface area, of this particular casting design, compared to the relatively small casting volume produced a casting with

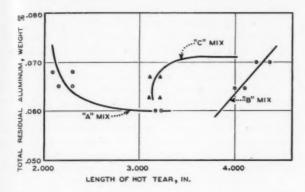


Fig. 36—Trend chart showing relationship between degree of hot tear and total residual aluminum in steel. Aluminum determination made by spectrographic method. There is no apparent correlation.

too high a resultant cooling factor and subsequently a casting comparatively insensitive to pouring temperature variation.

Throughout the progression of the investigation the reproducibility of results was very heartening. It was interesting to note, however, that whenever new core sand mixes were made, slight discrepancies occurred when the cylinders made with these mixes were compared with cylinders made from previous similar mixes. Until a study of data showed reasons for these discrepancies they proved slightly incomprehensible. Study of data showed the reason for variation in hot-tear incidence was caused by variation of core density. This variation was further traced to original moisture content of the core sand mixes which manifested itself in variable ramming characteristics of the core sand. It was finally shown that a definite relationship existed between core density and hot-tear incidence as shown on Fig. 27. As little as a few ounces variation in core weight in an approximately 24 to 27-lb core had an effect on the cumulative hot tear to be found in a cylinder. This fact substantiates the extreme sensitivity of the test

After this factor was applied as a correction to any tear in question, it was found that practically all tears obtained for a specific set of conditions were reproducible to nearly an identical degree. A review of Fig. 28 shows this. Note how the applied correction brings the two phases together. This is further shown on Fig. 27 where the dotted line portion shows that the tear would have been theoretically had the core densities of Phases I and II been identical.

The last remaining degree of variation which might be called "unaccountable variation" for purpose of discussion was considered to be a multiplicity of such other factors as metal composition, core mix variation, and the human errors involved. These reasons were considered applicable since all castings were produced under actual shop production conditions.

Figures 33, 34, 35 and 36 are presented to stimulate further interest in the field of research into the effect of metallurgical factors as they may affect hot tear incidence.

The data obtained at Burnside Steel Foundry was not sufficient to prove the validity of these charts. However, it is felt that the information so contained on these charts does indicate interesting trends which bear more complete and comprehensive study by

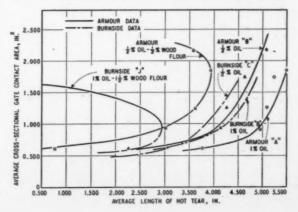
An interesting comparison of hot tear data is shown on Fig. 37 whereon is plotted the length of hot tear us average cross-sectional ingate contact area for the three basic core mixes used by Armour Research Foundation in its investigation and three of the basic mixes used by Burnside Steel Foundry in its investigation. Note that for the ½ per cent and 1 per cent oil mixes both investigators report fairly close results (considering variation in basic core mix raw materials). Comparison of the core mix with wood flour added, likewise is reasonable on the basis of the Armour mix maving ½ per cent less oil and 1 per cent less wood flour than did the Burnside mix.

Comparison of these curves, however, indicate the validity of the trend on charts previously shown in this report.

## Conclusions

The study of hot tear reproducibility of the Armour Research Foundation test casting at Burnside Steel Foundry Co. was begun in September, 1951, and ended February, 1952. During that period 414 cylinders having a net weight of 5,417 lb were poured. This amounted to 8,000 lb gross weight.

Fig. 37—Comparison between hot tear results obtained by Armour Research Foundation and Burns: de Steel Foundry.



The primary purposes of the investigation were to (1) show whether or not the test casting as designed by Armour Research Foundation would lend itself to reproducibility under actual production conditions, and (2) show whether or not the test casting in question was sensitive enough to permit reproducible results under varying conditions.

As a result of the tests made during this investigation under actual production conditions, the following summary of conclusions is made:

## **Primary Conclusions**

- 1. The Armour Research Foundation test casting furnishes reproducible results.
- 2. The test castings is sensitive to variation in conditions.

## Secondary Conclusions

- 3. Hot tear incidence is directly related to the degree of hot spot concentration.
- (a) Hot spot concentration is related to cross-sectional gate contact area.
- (b) Hot spot concentration is related to degree of choke of gate system.
- (c) Hot spot concentration is related to relative volume of casting section immediately adjacent to ingate contact.
- 4. Variation of hot tear incidence between two cylinders made from cores of similar core mix is directly related to differential density of cores which metal surrounds.
- (a) Moisture of core mix is principal factor determining rammability and therefore the density.
- 5. Hot tear incidence is apparently affected by various metallurgical factors in the following manner:
- (a) Hot tear incidence increases with increase in carbon content.

- (b) Hot tear incidence increases with increase in sulphur content,
  - (c) Hot tear incidence decreases with increase in

- 6. Hot tear incidence is directly related to the properties of the individual core mixes.
- (a) This is shown by variation in tear incidence with change in basic core mix constituents.
- (b) Although no definite correlation is presented between hot properties and tear incidence, there appears to be some intangible relationship which links combined hot property effects to hot tear incidence.
- 7. Pouring rate variation had no effect on hot tear incidence for this particular design test casting (pouring rate varied from 2.8 to 10.7 lb/sec). It was believed that the very high cooling factor of this particular test casting was responsible for this being so.

These conclusions are based entirely upon test data taken for the Armour Research test casting design. Again it should be stressed that different conditions would be presented by a casting of different shape, weight or design. However, it has been shown that certain basic factors concerning various core mixtures appear to influence the degree of hot tear incidence.

It appears that these basic factors, when fully interpreted, can be applied in production toward the goal of hot tear reduction, or elimination in steel castings.

The investigation submitted in Part III of this report was conducted by the Research Department under direction and supervision of the Metallurgy Department of Burnside Steel Foundry Co. and is respectfully submitted to the Research Committee of the Steel Division of the American Foundrymen's Society.

## HEAT FLOW IN MOIST SAND

By

## Victor Paschkis*

## I. Introduction

In the 1951¹ report of the A.F.S. Heat Transfer Committee, experiments for the determination of heat flow in moist sand were described. In all cases dealt with in this former report, the initial moisture content of the sand, expressed in per cent weight of the dry sand, was 4 per cent.

The present report is an extension of this work, to cover the case of an initial moisture content of 6 per cent. The sand was the same as that used in previous work, designated as sand 20-30, according to the work

of C. F. Lucks and co-workers.2

As work with the computing circuit described in the former report¹ progressed, difficulties became more and more noticeable. For example, in order to get good operational facilities the change in moisture content was carried out in four steps only, whereas a much larger number of changes would be desirable. Similarly, the thickness of the individual section in the computation experiments was made ½ in., but it would be desirable to decrease that thickness by ½ or ½ in order to improve smoothness of the curves. Those and other improvements were not carried out because it was felt that the present experiments are still preliminary in nature, and it was decided to avoid complications.

Since the properties of the sand may be of considerable influence on the results, they are listed here even though they are identical with those given in the former report. Merely the conductivity in the moist range is different, because of the changed initial value. The computation of this "moist conductivity".

is the same as previously.

Thermal conductivity of dry sand is compiled in Table 1; that of moist sand for low temperatures in Table 2.

The sand has a dry density of 108 lb/cu ft. The specific heat values for the various temperatures were

found by interpolating between the following values:

70	F							0.19	Btu/lb,F
1100	F							0.23	Btu/lb,F
2200	F							0.28	Btu/lb,F

The conductivity at full moisture content is not very different for 4 and 6 per cent-0.597 against 0.672.

This is understandable from Fig. 12 of the former report¹ which shows that the conductivity-moisture curve flattens out with increasing moisture content. Consequently, the differences in conductivity between 6 and 4 per cent full moisture content increases when the sand is partially dried. For example, when the moisture content has dropped to ½ of the original value, the conductivities for 4 and 6 per cent initial moisture content are 0.223 and 0.387, respectively,

TABLE 1—THERMAL CONDUCTIVITY OF DRY SAND 20-30 EXTRAPOLATED AFTER LUCKS AND CO-WORKERS

Temperatures, F		Conductivities Btu/ft, hr, F
2200		0.704
2000		0.600
1750		0.480
1500	15	0.417
1250		0.350
1000		0.300
750		0.250
600	1111	0.215
500		0.191
400		0.169
300		0.149
212		0.133

TABLE 2—THERMAL CONDUCTIVITIES FOR AN INITIAL

MOISTURE CONTENT OF 6 PER CENT

(Referred to the Weight of the Dry Sand)

Moisture Content in % of Original Moisture Content	Thermal Conductivit (Btu/ft, hr, F)				
100	0.672				
75	0.620				
50	0.530				
25	0.387				
0	0.133				

Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York City.

This is a report on a Research Project sponsored by the A.F.S. Heat Transfer Committee at Columbia University. Members of this Committee are as follows: H. A. Schwartz, Chairman, J. B. Caine, V. Paschkis, A. E. Schuh, C. E. Sims, H. F. Taylor and E. C. Troy.

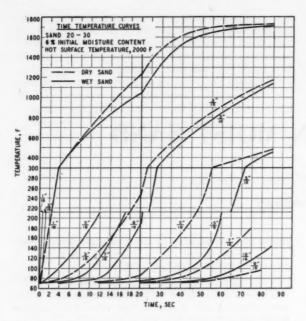


Fig. 1—Time-temperature relationships—6 per cent initial moisture content, 2000 F hot surface temperature.

which is a ratio of almost 1.75. Of course, ½ of the original moisture content means 1 per cent actual moisture content in the case of the 4 per cent original value, and 1.5 per cent of the 6 per cent original value.

## 11. Experimental Results

1. Temperature-Time Relationships are shown in Figs. 1 and 2 for hot surface temperatures of 2000 and 2600 F, respectively. The higher surface temperature (2600 F) results of course at any given time (e.g., 80 sec) and at a given point in the sand (e.g.,  $\frac{1}{16}$  in away from the surface) in a higher temperature (2400 F) than the lower surface temperature (2000 F), which results at  $\frac{1}{16}$  in. below the surface after 80 sec in a temperature of 1700 F. The latter temperature is lower than might be expected from a proportionality of temperatures:

$$(2000-70)/(2600-70) = 0.76,$$
  
but  $(1700-70)/(2400-70) = 0.70$ 

In order to have the same ratio, the temperature with 2600 F at the surface would have to be 1840 F, instead of the measured 1700 F. From a comparison of such ratios it becomes obvious that:

(a) The lack of proportionality is more pronounced at greater depths than at smaller ones;

(b) The lack of proportionality is in part caused by the decrease of dry sand conductivity at lower temperatures, and in part by the moisture content of the sand.

In the graphs the curves for dry sand have been repeated from the respective graphs of the former report.¹ In view of the shorter time in some of the earlier experiments the dry sand curves do not extend all the way to 80 sec.

2. Temperature-Space Relationships are shown in Figs. 3 and 4, the former holding for 2000 F, the lat-

ter for 2600 F. In these figures distances from the interface are plotted as abscissas and temperatures as ordinates. Each line holds for a different value of time of exposure to the constant hot surface temperature. The initial average temperature gradient over the first 1000-degree temperature drop is almost twice as steep in case of the higher pouring temperature than in the case of the lower one, as may be seen by comparing the curves for 2 sec.

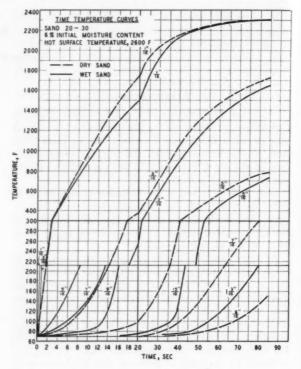


Fig. 2—Curves showing time-temperature relationship—6 per cent initial moisture, 2600 F hot surface temperature.

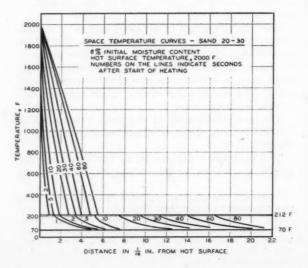


Fig. 3—Space-temperature relationships—6 per cent moisture content, 2000 F hot surface temperature.

A drop of 1000 F (from 2600 F to 1600 F) occurs over a distance of 0.01 in.; but for the surface temperature of 2000 F an equal drop of 1000 F (from 2000 to 1000 F) occurs over a distance of 0.02 in. But considering an average gradient for the drop to 212 F, the difference has almost disappeared: the gradient in case of the 2000 F surface is 19800 F/in., whereas in the case of the 2600 F the gradient is 19900 F/in.

3. Time-Space Relationships. These are shown in Fig. 5. Distances from the surface are plotted as abscissas, square roots of times of exposure to the constant hot surface temperature as ordinates. Two pairs of curves are shown, one to dry completely, the other to reach a temperature of 212 F. In each pair one curve holds for an initial temperature of 2600 F, the other for 2000 F. The lines are almost straight indicating an approximate square law relationship.

In plotting this curve, and incidentally the curve 2 of the former report, the influence of coarse lumping mentioned in Part I became particularly noticeable.

## III. Conclusions

1. Rate of Heat Extraction. As explained in the earlier report, conclusions may be drawn from the experiments regarding the heat withdrawn from the casting, even if the present investigations assume that the "hot surface" which represents the casting-mold interface is held at a constant temperature; in actual casting practice the temperature of the interface will not be constant. The amount of heat withdrawal from the hot surface (from the casting) can be expressed in Btu/hr, sq ft and compared with the amount of heat withdrawal which would obtain with completely dry sand. These comparisons are carried out in Figs. 6 and 7 for surface temperatures of 2000 F and 2600 F, respectively.

In each graph a third curve represents the ratio of

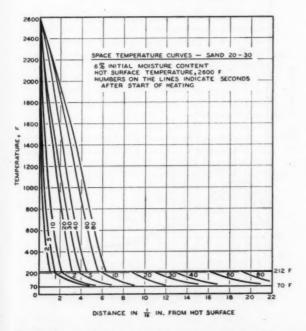


Fig. 4—Curves showing space-temperature relationships— 6 per cent moisture, 2600 F hot surface temperature.

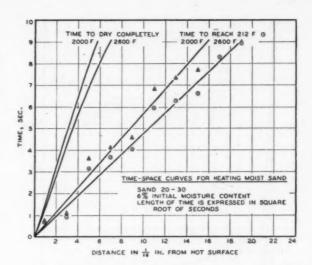


Fig. 5-Time-space relationship-heating moist sand.

the rates of heat extraction. All curves are plotted against times as abscissas. As in the case of the former report, the differences between wet and dry sand vanish after quite short times, indicating that at later times, i.e., for the solidification of thick castings, the difference between green and dry sands should be negligible. Under the condition of the experiment with the interface suddenly raised to a finite temperature, the rate of heat extraction either with dry sand or with moist sand is theoretically infinite.

However, the ratio of heat extraction by moist sand to that of dry sand is still finite, and is much lower than obtained from the analyzer experiments; 2.24 in case of 4 per cent moisture, and 2.45 in case of the 6 per cent moisture. Moreover, the rates of heat extraction drop exceedingly quickly from their initial theoretical infinite value. Exact computation is not possible in view of the fact that at least an exceedingly thin layer reaches immediately temperatures beyond 212 F, with the ensuing change of properties. But assuming that the properties do not change until sometime after the start of the experiment (this is the assumption underlying the analyzer experiments), Table 3 may be compiled.

The figures for 1 sec compare reasonably well with those observed for the moist sands on the analyzer; they are added as last line in Table 3 for the sake of comparison.

2. Influence of Moisture Content. With figures for 0, 4 and 6 per cent moisture content (by weight, with the weight of dry sand as reference point) available,

TABLE 3—COMPUTED RATES OF HEAT FLOW AT VERY
EARLY TIMES
(2000 F surface temperature, 104 Btu/sq ft, hr)

		1	Moisture Content							
		0% (dry)	4%	6%						
0.01	sec	95.6	215	234						
0.1	sec	30.3	68.2	74.2						
1	sec	9.6	21.6	23.4						
1	sec observed	4.8	21.4	18.2						

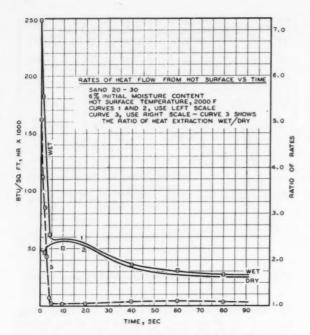


Fig. 6—Curves showing rates of heat flow vs. time. Hot surface temperature, 2000 F.

one can compare, at least in a preliminary way, the influence of moisture content on performance.

(a) Comparison of Heat Extraction

The heat extraction continues over long times, theoretically over infinitely long time. In actual foundry practice it continues up to the moment when the casting is shaken out. However, Figs. 6 and 7 of the present report, and Figs. 8, 9 and 10 of the former report¹ indicate that after relatively short time there is practically no difference between the heat extraction, by the wet sand compared with that by dry sand.

Hence it appeared reasonable to compare the total heat extracted from the casting during the first 60 sec. This comparison is shown in Table 4. The heat extracted is expressed in Btu/sq ft. The table shows the heat extraction for a 2000 F and 2600 F surface temperature and for dry sand, 4 per cent and 6 per cent moisture. It may be seen that at 2000 F the dry sand extracts least heat, the 4 per cent somewhat more and the 6 per cent still more. At 2600 F surface temperature the order is reversed and the 4 per cent moisture sand seems to extract more heat than the 6 per cent sand.

In all cases the differences are relatively minor. The obvious irregularities at 2600 F are explainable by the exigencies of the lumping procedure as used to date, and are part of the difficulties referred to in Section I. The procedure could be refined at great expense in time; the refined procedure would probably eliminate such irregularities.

At 2000 F the ratios of heat extracted are as follows: Dry: "4%": "6%" = 1:1.16:1.21

At 2600 F the ratios are as follows: Dry: "4%": "6%" = 1:1.19:1.11

It should be repeated here that, quite in general

and without reference to the irregularity at 2600 F, heat extraction in actual casting will be somewhat different because the interface temperature will not remain constant as was assumed here. But, in order to convey some ideas on order of magnitudes and the change in distribution which may be affected in a casting due to the changed heat extraction, the following figures are given.

With a metal having a specific heat of 0.2 Btu/lb, a density of 450 lb/cu ft, and a heat of fusion of 126 Btu/lb, the amount of 1000 Btu/sq ft represents a drop in superheat of 132 F over a thickness of 1 in., or a drop of 264 F over a thickness of ½ in.; if there were no superheat to be withdrawn (i.e., if the liquid metal would reach the mold at the liquidus temperature), then the amount of 1000 Btu/sq ft would suffice to extract the heat of solidification over a thickness of 0.21 in.

(b) Comparison of Times

Table 3 in the preceding section shows that the amount of heat extracted is not greatly influenced by the moisture content. This at first appeared surprising, because the heat which has to be supplied to a given volume of sand before it can exceed 212 F is more than doubled if the moisture content (referred to dry sand) is 4 per cent: 7892 Btu/cu ft as compared with 3720 Btu/cu ft for dry sand (see Table 3, reference 1).

One explanation for the surprisingly small influence of moisture is to be seen in the small part which the heat absorption of sand up to 212 F plays. Take

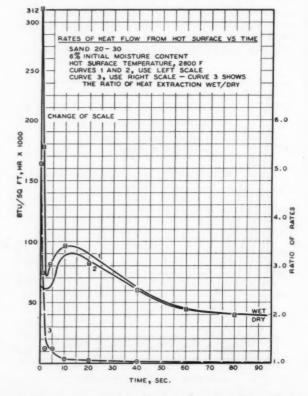


Fig. 7—Curves for rates of heat flow vs. time. Hot surface temperature, 2600 F.

for example, the case of 2000 F surface temperature and 6 per cent moisture.

After 20 sec a total of 404 Btu/sq ft have been withdrawn (see Fig. 6). Of this total 242 Btu/sq ft have been absorbed by the first 2.7/16 in. of sand, in their rise above 212 F, and the balance of 162 Btu/sq ft is stored in the first 7.5/16 in. to a varying degree, but all below 212 F. Figure 8 shows the approximate temperature distribution at this time.

Another way of looking at the small change of heat extraction is by observing the times to reach 212 F and to leave it. In case of wet sand there is a dwell time at 212 F, necessary to supply the heat of evaporation of moisture; hence there is for any point one value of time to reach 212 F and another to leave it

(evaporation completed).

For dry sand there is only one time to reach 212 F. Now, since the apparent conductivity of moist sand is higher than that of dry sand, the temperature of 212 F will be reached sooner in case of moist sand than in case of dry sand. This tends to increase the amount of heat extraction, since the temperature difference between constant surface temperature and the local temperature is decreased. But then the dry sand has, so to say, the opportunity to catch up and even to overtake the wet sand: the times at any point in the mold to leave 212 F for wet sand is longer than the time for dry sand to pass through 212 F.

This reversal counteracts the original increase in heat extraction from moist sand as compared with dry sand. Thus the difference in total heat extraction is relatively small. This is shown in Table 5, in which the times to reach 212 F and to leave it (complete dryness) are shown for various distances from the sur-

face and for the two surface temperatures.

The times to reach 212 F in case of 4 per cent moisture are longer than those in case of 6 per cent

Table 4—Total Heat Extraction (Btu/sq ft)
During First Minute

Surface Temp., F	Heat	Extraction by Sand w as Noted	ith Moisture
	0% dry	4%	6%
2000	688	801	830
2600	1118	1328	1231

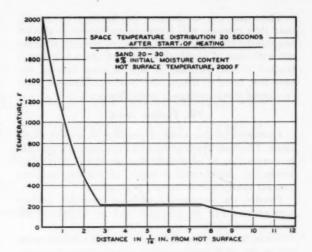


Fig. 8—Curve showing space-temperature distribution 20 sec after start of heating. Hot surface temperature, 2000 F.

moisture. The further one moves from the interface, the greater is the distance through which the heat has to travel through the sand. A large part of this travel is under the influence of a temporarily constant head of 212 F (obtaining from the evaporating parts).

This head finds, in case of the 4 per cent moisture sand, a medium with lower thermal conductivity than in the case of the 6 per cent moisture sand. Consequently, it takes longer to reach the 212 F in sand with 4 per cent moisture compared with that of 6 per cent. The greater amount of water to be evaporated causes, however, the 6 per cent sand to dwell longer at 212 F than does the 4 per cent sand and, therefore, the times to "leave 212 F" are longer in case of the 6 per cent sand than in the case of the 4 per cent sand.

3. Practical Application. In view of the small ratio of heat extraction values of moist sand compared with dry sand, it may be stated in general terms that the moisture content will have no great influence beyond the first few minutes of solidification.

However, thin castings or thin parts of castings which freeze in a short time and which have a relatively small amount of heat to give off to the mold, may be greatly influenced by moisture content.

TABLE 5-TIMES FOR TEMPERATURES TO REACH AND LEAVE A TEMPERATURE OF 212 F

		Surface	at 2000 F							
Distance from surface, in.	4% Sand		6% Sand		Dry	4% Sand		6% Sand		Dry
	Reach,	Leave,	Reach,	Leave,	Sand	Reach,	Leave,	Reach,	Leave,	Sand
2/16		9.0	***************************************	11.5	11.4		7.5		9.0	7.0
3/16	3.0	22.8	1.2	24.0	17.8	2.0	17.4	0.87	18.6	14.3
4/16		39.0		40.5	41.6		29.0		31.5	24.2
5/16	15.0	58.8	13.2	64.3	47.5	11.0	42.8	9.8	48.4	35.8
7/16	18.8		17.0			14.0		13.5		63.3
9/16			21.0			17.0		16.3		
11/16	51.5		47.0			37.0		35.4		
13/16			54.3			41.5		39.5		
15/16			60.0			67.0		43.7		
17/16								69.0		
19/16								80.3		
21/16								87.2		

Thus one can summarize the findings by stating that thin castings will freeze in considerably shorter time under the influence of moisture content, whereas the influence of moisture in case of thick castings is limited to the increase of solidification rate of the very first layer to freeze; the total solidification time in case of thick castings is not influenced.

## IV. Further Work

Extension of this work is possible in several directions. Through application of a finer network smoother curves can be obtained, and the irregularities investigated. Work with a lower, or higher, moisture content than covered to date is possible.

Finally, the assumption of a constant hot surface temperature can be dropped, and an actual casting can be simulated.

Since, however, the difference in heat extraction between dry and moist sand is certainly small during the first minute and almost negligible thereafter, the extension of work in any of the three directions is reasonable only if the high heat extraction in the first few seconds is metallurgically significant from the viewpoint of crystal growth, etc.

Of course, there is considerable theoretical interest in the problem. But there is some question how much practical significance an extension of the work would have outside the immediate chilling effect.

## Bibliography

- 1. Victor Paschkis, "Heat Flow in Moist Sand," A.F.S. TRANSACTIONS, vol. 59, pp. 381-391 (1951).
- 2. C. F. Lucks, O. L. Linebrink, and K. L. Johnson, "Thermal Conductivity of a Sand Mixture," Transactions, A.F.S., vol. 56, p. 363 (1948).

## DISCUSSION

- Chairman: H. A. Schwartz, National Malleable and Steel Castings Co., Cleveland.
- Co-Chairman: E. C. Troy, Foundry Engineer, Palmyra, N. J. Recorder: C. E. Sims, Battelle Memorial Institute, Columbus.

J. B. CAINE (Written Discussion): 1 Dr. Paschkis' work has some interesting applications in regard to sand collapsibility and hot tearing. Recent work on hot tearing by Bishop, Ackerlind and Pellini* shows that hot tearing in steel sections under 2 in occurs 2 to 3 min after pouring and with contractions measured in thousandths of an inch. This work emphasizes the need for information concerning the rate of heat input into the first 1/16 in. or so of the sand during the first few minutes after pouring Data in these space, time intervals are difficult to obtain experimentally.

Bishop, Brandt and Pellini (A.F.S. Transactions, vol. 59, p. 442, 1951) have published sand temperatures ½ in. back from the sand-metal interface at an interface temperature of 2600 F, the same as used in Fig. 2 of this paper. However, the sand temperatures determined experimentally are of a different order of magnitude than those given in this paper. Figure 2 of this paper shows the sand ½ in. back from the sand-metal interface reaches a temperature of 400 F in 0.35 min, 800 F in 1.0 min. The experimental temperatures are 400 F in 2 min, 800 F in about 3.4 min.

Could these discrepancies be due to sand grain size? Reference 2 of this paper shows quite a difference in heat conductivities under equilibrium conditions due to grain size. The sand temperatures were determined experimentally with an A.F.S. 80 sand, the sand used in this work is much coarser A.F.S. 20-30 grain size.

DR. PASCHKIS (Written Reply to Mr. Caine): There are two possible explanations for the difference between the times indicated in the present paper and in the paper by Bishop, Brandt and Pellini (A.F.S. TRANSACTIONS, vol. 59, p. 442, 1951).

First, as Mr. Caine indicates himself, the sand in the two investigations was not identical. The lower conductivity used by Bishop and coworkers should result in a slower temperature rise in the sand.

The second possible explanation is the difficulty in accurately locating the thermocouple in sand and of measuring temperatures in sand altogether.

The temperature gradient in the sand is very steep and a small error in location of the couple, perfectly reasonable for normal experimentation, may result in considerable variations in time to reach a certain temperature. Moreover the temperatures in the sand are hard to measure because of possible poor contact of the couple with the sand. This latter difficulty is, however, somewhat controversial.

¹ Foundry Consultant, Cincinnati 15, Ohio,

^{* &}quot;Metallurgy and Mechanics of Hot Tearing," A.F.S. Transactions, vol. 60 (1952).

# EFFECTS OF MELTING FURNACE ATMOSPHERE ON FLUIDITY, HOT TEARING TENDENCY, AND OTHER PROPERTIES OF MALLEABLE IRON

By

E. A. Lange* and R. W. Heine**

## Introduction

The properties of white cast irons are known to vary in a way that is related to factors operating during the practice of foundry melting. Some of the properties suspected of being affected by melting conditions are fluidity, tendency toward hot tearing, mottling tendency, response to malleablizing, and the mechanical strength of the malleablized white iron. The aforementioned properties could be favorably or unfavorably altered by a number of melting variables.

Among these are the materials involved such as the metal charge, fuels, and refractories; slags and gases which may interact with the melt; the cycle of temperature and composition changes which occurs; the type of melting furnace; the pouring practice; and many other factors. A few of the foregoing melting conditions which influence metal properties have been studied experimentally, but many remain uninvestigated. Included in the unevaluated factors are the effects of melting furnace atmospheres. The furnace atmosphere presents a melting condition of considerable significance in the malleable iron foundry in view of the extensive use of reverberatory type furnaces where reactions between the molten metal and the atmosphere above it occur over an extended period in the melting cycle. The influence of these melting furnace gases on the properties of malleable iron is being investigated as a research project of the A.F.S. Malleable Division Research Committee. The present paper reports some of the progress made in the investigation. Since the research work is far from completed, this paper should be viewed as offering results only in their present state of development.

Nitrogen, carbon monoxide, carbon dioxide, oxy-

gen, hydrogen, water vapor, and possibly hydrocarbons may exist under various conditions in melting furnace atmospheres. Of these gases very small percentages of nitrogen, oxygen, and hydrogen have been proven to exist in some form in cast irons. The melting furnace atmosphere, then, may act as a source of these gases and alter the amounts present in the iron. Once present in the iron, fundamental properties of the metal may be influenced. In addition to direct effects of gases in iron, indirect ones may exist because of chemical reactions between the melt and furnace gases. An atmosphere could be oxidizing, neutral, or reducing and thus promote or restrict chemical re-

Fig. 1—Photograph showing the physical set-up for melting under atmospheres and pouring fluidity and hot tear castings.



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^{**} Asst. Prof. of Metallurgical Engineering, Dept. of Mining and Metallurgy, University of Wisconsin.

This report covers a research project initiated and sponsored by the Research Committee of the Malleable Division of the American Foundrymen's Society. Members of this Committee are as follows: C. F. Joseph, Chairman, W. D. McMillan, Vice-Chairman, H. Bornstein, W. A. Kennedy, J. H. Lansing, R. P. Schauss, R. Schneidewind and Milton Tilley.

actions. Whatever the direct or indirect effects of the gases may be, a study of the problem is made difficult by the fact that the reacting metal bath and furnace gases in foundry melting are complex in chemical composition and thermal history. The molten iron contains carbon, silicon, manganese, sulphur, phosphorus, and other elements while the furnace atmosphere consists of a mixture of the previously listed gases varying in concentration according to the combustion conditions prevailing. In this situation, complicated by rising and falling temperatures and the presence of some slag, the nature and existence of equilibrium conditions between molten metal and atmosphere is difficult to visualize. The complexity and variation of the melting furnace atmosphere particularly makes a direct study of the influence of this atmosphere on metal properties inadvisable. It was therefore decided to simplify the problem, at least in the beginning, by investigating the effects produced by atmospheres developed from introducing pure gases of only one constituent over the molten metal. This method of attack was expected to yield the principal and significant effects of individual gases. The effects of nitrogen, carbon monoxide, carbon dioxide, hydrogen, water vapor, and argon on hot tearing tendency, fluidity, chemical composition changes, mottling, and pinholing were studied by the procedure described in the following section.

## **Experimental Procedure**

Since the investigation is concerned with the effects of a particular phase of melting practice on metal properties, details of the steps in the entire melting cycle are of the utmost importance. The steps followed in the melting process are therefore listed below together with explanatory information where necessary:

(A) Nine pounds of duplex process white iron sprue are melted down in a covered 17-lb magnesia crucible by induction heating. The melting stock* is the same material used in earlier work.¹ When the metal temperature reaches 2500 F, slag is removed from its surface. Time required for this step is 55 to 60 min.

(B) The melt is then heated to 2600 F in about 5 min. At this time the composition is adjusted. Higher carbon contents are obtained by adding pulverized electrode graphite. Carbon contents below the melt-down of about 2.30 per cent are obtained by oxidation with the crucible cover removed. Ten points of manganese in the form of low carbon 85 per cent ferro-manganese are added to all heats except a few in which the effect of higher manganese percentages were studied.

(C) The melt is then heated to 2700 F, allowing about 10 min for solution of additions.

(D) A cover specially prepared for maintaining the desired gas atmosphere is placed on the crucible. The gas flow is started and continued at 2 liters/min over the molten metal. The physical set-up of gas tank, meter, purifying train, thermocouple pyrometer, furnace, and crucible cover is shown in Fig. 1. The

crucible cover permits the insertion of a silica protection tube for the platinum-platinum rhodium thermocouple and the introduction of the gas atmosphere through a silica tube extending downward to within 3/8-in. of the metal surface. A hole 5/8-in. in diameter in the center of the cover permits the exit of the furnace gases at virtually one atmosphere pressure and allows the withdrawal of metal samples.

(E) The crucible is flushed with the 2 liter/min gas flow for 15 min while maintaining the temperature at 2700 F. Samples of metal and gas are taken at this time, called the zero-time of a 60-min holding period. The metal samples are taken by sucking them into a pyrex tube which is subsequently quenched in water. Gas samples are taken by displacement of mercury in a gas sampling bottle.

(F) The molten metal is held at 2700 F for 60 min, maintaining gas flow at 2 liter/min. Metal and gas samples are taken at 30- and 60-min time periods. Temperature variation during this period is  $\pm$  10 F.

(G) At the end of the 60-min holding period fluidity and hot tear castings are poured. About 10 sec or less elapse from the moment the furnace power is cut off to the end of pouring. Data are immediately taken on the hot tear casting by the procedure described in Reference 1.

Data obtained by following the practice described

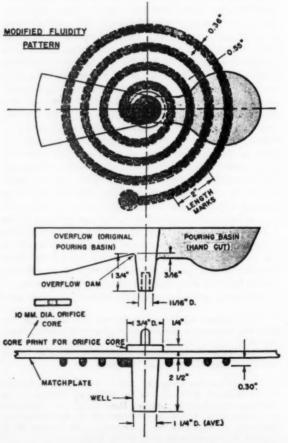


Fig. 2—The modified spiral casting employed for making fluidity measurements (Reference 2).

Average analysis of 2.54 per cent C; 1.30 Si; 0.41 Mn; 0.147
 S; 0.11 per cent P.

bove established the relationships existing between melting furnace atmosphere and metal properties. Fluidity effects were measured by casting a fluidity spiral from each heat. The spiral employed was deeloped by E. A. Lange, L. F. Porter, and P. C. Rosenthal, and is a modification of the Saeger and Krynitsky fluidity test. Figure 2 shows the spiral casting and its use in fluidity measurements is described in detail by Porter and Rosenthal.2 Also poured from each heat was the hot tear test casting developed especially for this research project and described in Reference 1. The two castings were poured from a number of heats melted under each atmosphere but varying in carbon percentage so that the relationship between carbon content and these two properties could be determined, assuming, of course, that other elements in the iron are held virtually constant. The samples of metal and gas withdrawn from the furnace during the 60-min holding period made it possible to determine the existence and nature of chemical reactions between melt and atmosphere. Observations of mottling tendency were made on fractures of the lower one-third of the well of the fluidity casting (Fig. 2). The fluidity well also provided visual evidence of any tendency for metal porosity due to gas evolution during pouring and cooling of the casting.

The practice of holding the metal at 2700 F in contact with the furnace atmosphere for a 60-min period was found necessary since shorter periods did not

cause an apparent equilibrium to be reached. Some heats made under a nitrogen atmosphere but held only 30 min at temperature definitely proved that an equilibrium condition with respect to the metal properties of fluidity and hot tearing was not reached during the shorter holding period. The difficulty of reaching some kind of equilibrium is a problem in all gas-metal studies, and was encountered several times during this work in spite of the 1-hr holding period.

## **Chemical Composition**

The composition of the furnace gases and their effects on the chemical composition of the metal during the 60-min holding period was determined for each of the atmospheres studied. Data obtained in heats where nitrogen, argon, carbon monoxide, hydrogen, water vapor and carbon dioxide were introduced as the furnace atmosphere are presented in Tables 1 through 6. These results may be interpreted according to the chemical nature of the atmosphere, i.e., neutral, reducing, or oxidizing. The composition changes of the metal followed established principles reported in other papers by the authors 3.4 On the basis of these principles it was anticipated that metal composition would be influenced by the melting procedure as follows:

1. The principal losses of manganese and silicon should occur during melting down and when metal temperatures are under 2600 F.

TABLE 1-DATA FOR HEATS MADE AT 2700 F UNDER CO ATMOSPHERE

Heat		Met	al Composit	ion, %		Gas Comp	osition, %2	Fluidity,	Hot Tear Spring
No.	C1	Si	Mn	P	S	CO,4	co	in.	Deflection, 0.0001 in
K-1	2.53	F	1			1.14	_	23.0	48
K-2	2.60				0.157			23.7	49
0 min	2.60	1.31	0.44	0.107		1.23	_		
30 min	-	1.32	0.42			0.82	_		
60 min	2.58	1.32	0.43	0.112		0.92	98.90		
K-3	2.20	_						19.4	_a
0 min	2.20	1.34	0.42			0.87	99.10		
30 min		1.31	0.42			0.94	_		
60 min	2.18	1.31	0.43			0.95	_		
4	2.18					0.81		18.6	57
						0.63	_		
						0.56	_		
5	2.80					1.18	_	25.1	52
						0.75			
						0.76	_		
6	2.88					0.85	-	24.8	57
						0.75	_		
						0.76	_		
7	2.92					1.36	-	27.6	68
						0.80	****		
						0.72	-		
8	2.08		0.43			0.82	_	17.1	51
						0.73	-		
						0.78	-		
9	2.08		0.44			0.99	_	18.3	81
						0.88			
						0.83			
10	2.45					0.93	_	20.3	50
						0.66	-		
						0.62			
11	2.02					0.70		15.5	95

¹ Percentage carbon determined on twinnings taken from hot tear casting, other analyses on samples sucked from melt.

⁸ Hot tear casting did not fill out at stud end.

² Gas analyses were performed with a Bureau of Mincs type precision gas analyzer.

⁴ Where three values are presented they represent samples taken at the 0-, 30- and 60-min periods.

TABLE 2-DATA FOR HEATS AT 2700 F UNDER HYDROGEN ATMOSPHERE

Heat		etal sition, %	Gas Co	omposi	tion, %	Fluidity,	Hot Tear,	
No.	C	S	CO	H ₃	CH,	in.	0.0001 in.	
			12.2	86.8	1.59			
M-1	2.27		11.4	84.8		11.5	85	
			10.0	86.9	1.11			
M-2	2.58		9.0	67.4	4.72	19.7	71	
M-3-0	2.88	0.155	13.1	83.4	3.00			
M-3-30	2.80							
M-3-60	2.74		13.1	85.0	2.5	22.8	72	
M-4-0	2.50		18.6	77.9	1.9			
M-4-30	2.45							
M-4-60	2.40		15.5	81.3	2.3	18.9	86	
			12.1	71.2				
M-5	2.01		14.0	60.5		12.2	62	
M-6	2.29		15.3	82.4	2.3	16.1	92	
			14.8	83.0	2.2			
M-7	2.04	0.156	14.6	83.7	1.7	13.2	93	

2. No losses of silicon or manganese should occurduring a holding period at temperatures above 2670 F under neutral or reducing gases. Silicon or manganese losses would not be anticipated under oxidizing gases unless iron oxide as a solid or in a slag is developed by those gases.

3. Carbon losses would begin under oxidizing atmospheres as soon as the dry slag cover disappears on heating above about 2500 F.

4. No change in sulphur or phosphorus would be anticipated.

The data presented on Tables 1 through 7 support these principles completely. A summary of metal analysis changes during the 60-min holding period is

TABLE 3-DATA FOR HEATS MADE AT 2700 F UNDER

Heat		Gas Comp	osition, %1	Fluidity,	Hot Tear	
No.	C, %	CO ₃	co	in.	0.0001 in	
L-1*	2.36	3.79	0.34	15.7	77	
		0.45	0.17			
		0.29	0.09			
L-2	2.39	0.33	1.17	17.4	58	
		0.13	0.37			
		0.10	0.30			
L-3 2.79	2.79	0.42	1.26	23.7	51	
		0.23	0.76			
		0.21	0.57			
L-4	2.13	0.42	0.97	16.0	72	
		0.30	0.45			
		0.16	0.48			
L-5	2.32	0.73	0.44	16.9	75**	
		0.40	0.67			
		0.16	0.29			
L-6	2.60	0.53	0.79	19.5	-**	
		0.33	0.50			
		0.25	0.30			
L-7	2.13			17.1	54	
L-8	1.85			14.1	130+	

• Per cent Mn, 0.43

** Fin at hot spot.

+ No break-bad crack at sprue.

¹ The three values given are 0-, 30-, and 60-min samples.

given in Table 7. No loss of silicon or manganese occurred during the holding period at 2700 F regardless of the atmosphere employed. There was, however, a difference in the melt-down loss of silicon and manganese in the heats made under the oxidizing atmospheres produced by carbon dioxide and water vapor. The neutral gases produced a final metal analysis of about 1.30 per cent Si, whereas the oxidiz-

Table 4—Data for Heats Made at 2700 F Under  $N_2$  Atmosphere

Heat Metal	Metal	Compos	ition, %	Atmosp	here, %1	Fluidity,	Hot Tear Spring	Holding	
No.	C	Si	Mn	CO,	CO	in.	Deflection, 0.0001 in.	Time, hr	
P-1	2.04					13.7	99	1/2	
P-2	2.40					20.0	64	1/2	
P-3	2.24					14.0	52	1/2 1/2 1/2 1/2	
P-4	2.52			6.75	8.61	21.5	55	1/6	
				0.64	0.21			/-	
P-5	2.39			(60)					
	4.00			1.55	0.52	17.0	52	1	
P-6	2.68					23.5	37	1/2	
P-7	2.52			1.30	0.51		48	î	
	0.7			0.72	0.18	_			
P-8	2.13			2.15	0.50	_	53	1	
				0.98	0.33	_			
				0.862	0.29	-			
P-9	2.49			0.34	0.17	_	54	1	
				0.31	0.15				
P-10	2.59	1.31	_				55	1	
P-11-0	2.58	1.29	.41						
P-11-60	2.53	1.28	.40			17.8	60	1	
P-12	2.39			0.33	0.17	16.2	55	1	
				0	0				
				0	0				
P-13	2.53			0.34	0.34	18.8	40	1	
				0.28	0.14				
				0.19	0.19				
P-14	2.07					11.9	101	1	
P-15	2.92					25.9	25	1	
P-16	2.86	1.29				21.2	61	1	
P-17	2.34					14.8	58	1	

ring gases yielded one of about 1.20 per cent. This effect is thought to be due to the fact that more iron-oxide-bearing slag was formed, and also carried over on the crucible walls from heat to heat during experiments with the latter gases. However, it should be recognized that the oxidation losses occurred during melt-down or raising of the temperature and not upon holding under the oxidizing atmosphere once the 2700 F temperature was reached.

The principle metal analysis variable was carbon. It can be seen from Table 7 that virtually no carbon losses occurred from CO and N2 atmospheres. The slight loss evident, 2 to 5 points/hr, is believed due to iron oxide remaining on the crucib'e walls from the melting-down step. Hydrogen caused a small loss of 10 to 14 points/hr. The composition of the furnace gases, Table 2, shows the presence of methane and appreciable carbon monoxide, and this explains the carbon drop. A more rapid carbon loss resulted from the oxidizing gases, carbon dioxide and water vapor. The latter gases caused carbon to be lost at the rate of 55 to 69 points/hr. The oxidation rate in air with the metal at 2700 F was about 1 point/min, so that similar rates prevailed in the three oxidizing gases studied. The aforementioned rates were dependable enough so that carbon contents could be controlled and predicted with reasonable accuracy.

The composition of the furnace atmosphere generated by each gas is of interest. All the tank gases were treated by a purifying train before entering the furnace chamber. Water was removed by a calcium chloride tower series, and CO₂ was removed by concentrated KOH and ascarite. Melting furnace atmospheres produced by the gases argon and nitrogen,

Table 5—Data for Heats at 2700 F Under Carbon Dioxide Atmosphere

Heat		Meta positi	on, %		as ition, %		Hot Tear,
No.	C	Si	Mn	CO ₂	CO	in.	0.0001 in.
Q-1	2.13			23.0		12.6	76
				35.9	63.8		
			_	32.7	67.1		
Q-2	2.46			37.8		17.4	73
				41.7			
Q-3	2.50			48.3		16.8	56
				48.0			
Q-4	2:60					21.1	47
Q-5	2.16			40.5		13.7	78
				40.6			
				41.9			
Q-6	2.49			44.9		17.7	47
-				43.3			
				39.3			
Q-7	3.42	1.22	0.39	39.7			
(0-min)							
0-7	2.77	1.22	0.42	34.4		24.3	55
(60-min)						-	
0.8	2.98	1.22	0.35	43.0			
(0-min)		-					
0-8	2.29	1.21	0.37	39.6		17.3	56
(60-min)			-				-
Q-9	2.04			37.8		12.6	80
-				35.7			
Q-10	2.12			37.6		10.9	63
2 .0				35.3			
				36.6			

TABLE 6-DATA FOR HEATS AT 2700 F UNDER WATER VAPOR ATMOSPHERE

Heat		Metal oositio	n, %	C		Gas sition, ⁶	%• Flu	idity,	Hot Tear,
No.	C	Si	Mn	Co ₂	H ₂	CO	CH,	in.	0.0001 in.
R-1	2.44	1.15						19	45
R-2	2.41			6.82	42.65	45.20	5.80	18.9	51
R-3	1.80			3.37	45.30	44.80	6.00	6.5	No break
R-4	2.63	1.17	.35						
(0-min)									
R-4	2.08	1.18	.35					11.3	76
(60-min)									
R-5	2.53			5.53	45.50	39.70	9.90	19	40
R-6	3.36	1.19	.39						
(0-min)									
R-6	2.70	1.21	.40	5.88	46.20	0 40.15	7.70	23.3	33
(60-min)									
* 60 M	linute	Samp	oles,	satui	rated	basis			

Tables 3 and 4, showed small percentages of both carbon oxides. The carbon oxides are believed to originate by the reaction of carbon in the melt with the slight amount of iron oxide adhering to the crucible walls from the melt-down operation and carrying over from heat to heat. The same iron oxide is thought to be responsible for the CO in the atmosphere of hydrogen treated heats (Table 2). Both the hydrogen and water vapor atmospheres developed appreciable percentages of methane.

Interpretation of furnace atmosphere compositions developed by water vapor and carbon dioxide, Tables 5 and 6, is questionable because of the nature of the sample. Samples were taken near the point of effluence of the gases from the furnace. However, the active effect of the gas exists where it contacts the metal and there its composition is rapidly changing due to the chemical reactions involved. The gas analysis presented in Tables 5 and 6 are therefore probably qualitative rather than quantitative in their relationship to the reactions actually progressing at the metal-gas interface.

## Atmosphere Effects on Fluidity

Fluidity spiral castings were poured from each controlled atmosphere heat. The fluidity in inches spiral length is given for each heat in Tables 1 through 6. These data are plotted in Figs. 3 and 4. Figure 3 shows the data revealing the effects of introducing carbon monoxide, argon, and hydrogen over the melt when the spiral casting is poured from 2700 F at various carbon levels. Also shown in Fig. 3 is the comparison with standard fluidity results obtained when an atmosphere formed by merely covering the melting crucible is in contact with the metal. The curves shown for the air standard were plotted from the following formulas:

Fluidity, in.  $\equiv 3.81 \text{ F} + 0.055\text{T} - 161.5$ where T  $\equiv$  temperature, °F

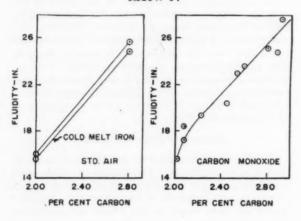
F = a chemical composition factor obtained from the equation  $F = 3.4 C + 0.26 Si + 0.57 C \times P$  in which C is the percentage carbon, Si the silicon, and P the phosphorus percentage in the iron.

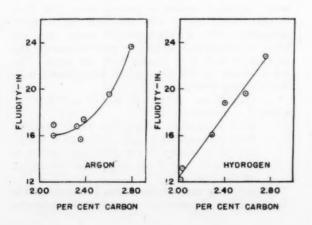
These formulae were obtained from Reference 2. The fluidity curve obtained for irons melted under the

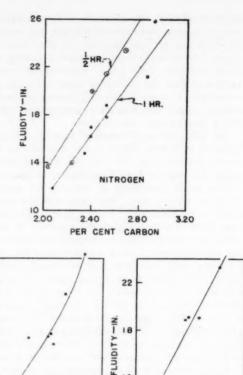
CO atmosphere closely follows the standard air fluidity curve and substantiates it. This might be expected since metal in a covered crucible, i.e., the condition referred to herein as standard air, would have an atmosphere above it high in CO content. A notable effect shown in Fig. 3 is the decrease in fluidity caused by hydrogen. The same effect is produced by argon at high but not at low carbon contents. Lowered fluidity was also obtained when nitrogen, carbon dioxide, and water vapor atmospheres were employed. Data for the aforementioned gases are plotted in Fig. 4.

For comparison purposes, idealized fluidity curves produced by all the atmospheres studied are given in Figs. 5 and 6. From the latter figures several facts are evident. Of the gases studied, a carbon monoxide atmosphere produces maximum fluidity. Fluidity is lowered by the neutral gases argon and nitrogen, by the oxidizing gases carbon dioxide and water vapor, and the reducing gas hydrogen. Nitrogen appears to be the most effective in lowering fluidity over the

Fig. 3—Graphs showing the relationship between percentage carbon and fluidity in inches when the duplex iron is poured from 2700 F after a 1-hr holding period at temperature in contact with the special gas atmosphere. The two curves on the standard air atmosphere fluidity graph are for two different irons, the upper curve being for the duplex iron having about 1.31% Si and 0.11% P, and the lower curve one for a cold melt iron of 1.07% Si and 0.126% P.







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10 -

Fig. 4—Graphs showing the relationship between percentage carbon and fluidity in inches when the duplex iron is poured after holding 1 hr at 2700 F in contact with atmospheres developed by  $N_2$ ,  $CO_2$ , and water vapor. In the case of  $N_2$ , a curve for a  $\frac{1}{2}$ -hr holding period is also shown.

2.00

2.40

PER CENT CARBON

2.80

CARBON

2.40

PER CENT CARBON

DIOXIDE

2.80

TABLE 7—EFFECT OF GASES ON CHEMICAL COMPOSITION
DURING 60-MIN HOLDING PERIOD

Heat No.*	Gas at 2 liter/min		ΔC pts/hr		ΔSi ts/hr	Mn, %	ΔMn pts/hr
K-2-0	co	2.60		1.31		0.44	
K-2-60		2.58	_2	1.32	+1	0.43	-1
K-3-0		2.20		1.34		0.42	
K-3-60		2.18	_2	1.31	_3	0.43	+1
M-3-0	H,	2.88					
M-3-60		2.74	-14				
M-4-0		2.50					
M-4-60		2.40	-10				
P-11-0	N ₂	2.58		1.29		0.41	
P-11-60		2.53	_5	1.28	-1	0.40	-1
Q-7-0	CO ₂	3.42		1.22		0.39	
Q-7-60		2.77	65	1.22	0	0.42	+3
Q-8-0		2.98		1.22		0.35	
Q-8-60		2.29	69	1.21	-1	0.37	+2
R-4-0	$H_2O$	2.63		1.17		0.35	
R-4-60		2.08	_55	1.18	+1	0.35	0
R-6-0		3.36		1.19		0.39	
R-6-60		2.70	-66	1.21	+2	0.40	+1

* 0 and 60 indicate the initial and final analyses during the 60-min holding period.

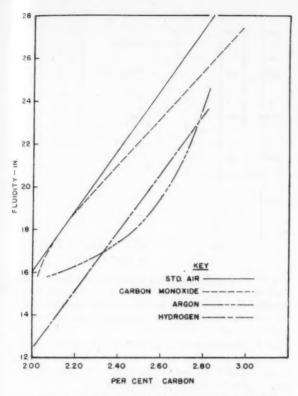


Fig. 5—Idealized curves of fluidity vs. carbon percentage in the duplex iron poured from 2700 F after a 1-hr holding period.

entire carbon range. Figs. 5 and 6 also show that there is tendency at the higher carbon percentages for fluidity to increase under all atmospheres toward the higher level promoted by a carbon monoxide-rich atmosphere. The nature of the data suggest the possibility that the decreased fluidity promoted by all atmospheres except carbon monoxide is due to an absence of CO in the melt rather than the presence of some other gas. The actual explanation of the fluidity effects observed, of course, requires additional work.

## Atmosphere Effects on Hot Tearing Tendency

A hot tear test casting was poured from each heat melted under a special atmosphere. The hot tear test described in Reference 1 was utilized and yielded the data presented in Tables 1 through 6. The data obtained from heats melted and held at 2700 F under carbon monoxide and hydrogen atmospheres, Tables 1 and 2, are shown plotted as curves in Fig. 7. This figure includes a graph of the relation of percentage carbon to resistance to hot tearing as set forth in Reference 1 for melting under an air atmosphere.

The upswing above 2.60 per cent carbon in the curve for metal melted under CO is accompanied by micro-mottling, as in the case of air atmospheres, Reference 1. The effects revealed in Fig. 7 have special significance in terms of the range of values encountered. The curve for hydrogen appears to be an upper limit, while that for carbon monoxide a lower limit of

hot tear resistance at carbon percentages below about 2.50 per cent when the metal is poured from 2700 F. On the other hand, the hot tear resistance curve for metal melted under air is intermediate, being between the curves of the two former gases except at about 2.60 per cent carbon where it falls to a minimum.

Atmosphere effects on hot tearing similar to those of air were produced by nitrogen and water vapor. Data from Tables 4 and 6 for the effects of these gases are shown plotted in Fig. 8. The data for water vapor atmosphere heats follow the curve for air values closely. The nitrogen data show considerable scatter although they remain well within the limits of the two extremes encountered, CO and H₂, and follow in a general trend the curve of the air data. It may be noted in Fig. 8 that a minimum resistance to hot tearing was encountered at high carbon content under a nitrogen atmosphere.

The hot tear data for carbon dioxide atmosphere heats, Table 5, are shown plotted in Fig. 9. Some values from these data are seen in Fig. 9 to fall directly on the curve for metal melted under a CO atmosphere. Other values are displaced from the CO curve and suggest a curve of the same general shape but shifted to the right. The displacement may be associated with some oxidation factor since air and water vapor show similar effects.

The increased resistance to hot tearing above 2.60 per cent carbon evident in Figs. 7 through 9 is related to the micro-mottling phenomenon reported in Refer-

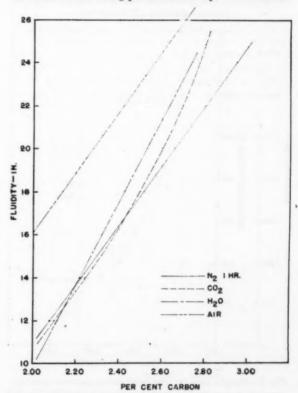


Fig. 6—Idealized curves of fluidity vs. carbon percentage in the duplex iron poured from 2700 F after a 1-hr holding period.

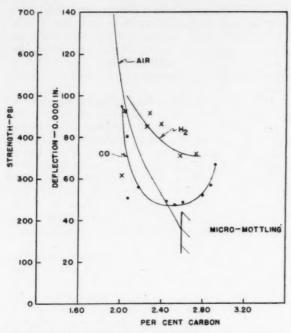


Fig. 7—Graphs showing the relationship between percentage carbon and resistance to hot tearing when the duplex iron is poured from 2700 F after a 1-hr holding period under atmospheres produced by carbon monoxide and hydrogen. The curve for the air values is reproduced from Reference 1. Micro-mottling indicates the presence of primary graphite (Reference 1).

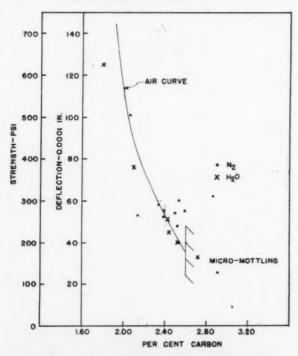


Fig. 8—Graphs showing the relationship between percentage carbon and resistance to hot tearing when the duplex iron is held for 1 hr at 2700 F under atmospheres produced by nitrogen and water vapor. The curve for the air values is reproduced from Reference 1.

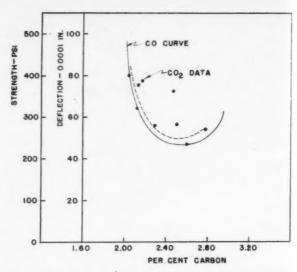


Fig. 9—Plot of hot tear data obtained from duplex metal held for 1 hr at 2700 F under an atmosphere developed by CO₂. The CO curve is the same as that shown in Fig. 7 for data from CO atmosphere heats.

ence 1. The effect of atmospheres on mottling can be observed microscopically in the hot tear castings, but is more easily studied in the well of the fluidity casting, as pointed out in a later section.

Effect of Temperature on Hot Tearing: As a part of the study of the influence of atmospheres on hot tearing, the effect of pouring temperature was determined. Since temperature greatly influences the solubility of gases in metals, it was thought necessary to evaluate the temperature effect. In this study, 3-lb heats of duplex iron were melted by induction heating. Three hot tear castings were poured from each heat, each casting at a different carbon level. The procedure followed is given in detail in Reference 1, and is essentially the same as that described earlier in this paper except that the melting furnace atmosphere was air. Castings were poured from 2700, 2800, and 2900 F. Data for castings poured from 2800 and 2900 F are presented in Table 8. Data for castings poured from 2700 F are presented in Reference 1.

Figures 10 and 11 are graphs of the hot tear data in Table 8 for metal poured from 2800 and 2900 F, respectively. It may be noted in Figs. 10 and 11 that high hot tear values occasionally occur in the range of 2.30 to 2.40 per cent carbon. This is believed to be an effect of carbon monoxide or hydrogen at the higher temperatures since it does not occur at the lower temperature of 2700 F.

For comparison purposes, idealized curves at the three pouring temperatures and for metal melted under CO are presented in Fig. 12. The latter figure shows that the carbon monoxide atmosphere raises the minimum from the hot tear curve for 2700 F in the vicinity of 2.60 per cent C. Furthermore, the curve of the CO data is similar in shape to the curve of hot tear values for metal melted under air at 2900 F. The similarity of curves should be associated with the increased solubility of CO at the higher temperatures. Figure 12 also reveals that less resistance to

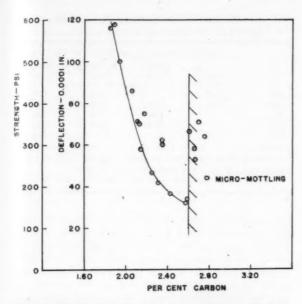


Fig. 10-—Graph of resistance to hot tearing as a function of carbon percentage in the duplex iron melted under an air atmosphere and poured from 2800 F. Micro-mottling indicates the presence of primary graphite.

TABLE 8-HOT TEAR DATA FOR METAL MELTED UNDER AIR ATMOSPHERE

Heat	Pouring	Meta	Hot Tea Spring Deflection		
No.	Temp, F	C	Si	Mn	0.0001 in
G 37		2.78			44
38		2.42			37
40		2.57	1.32	0.42	32
41		2.25			47
42		1.89			118
43		2.59			34
44		2.31			42
45		1.94			100
46		2.67			53
47		2.06			86
48	2800	1.85			116
49		2.66			64
50		2.15			58
67		2.66			58
68		2.35			62
69		2.13			70
70		2.66			71
71		2.35			- 60
72		2.18			75
76		2.61			68
77	*	2.11			71
52 1		2.60			50
53		2.24			72
54		1.97			77
58		2.64	1.25	0.44	44
59	2900	2.42			42
60		2.25	1.29	0.41	47
64	- 1	2.67			59
65		2.30			58
82		2.71			73
83		2.42			49
84		1.90			94

TABLE 9-DATA FOR HEATS AT 2700 F UNDER AIR
WITH VARYING MANGANESE CONTENT

Heat	. 1	Hot Tear,			
No.	C	Si	Mn	S	0.0001 in.
0-7-1	2.69	1.23	0.33	0.157	86
0-7-2	2.45		0.33		42
0-7-3	2.31		0.33		73
0-8-1	2.54	1.31	0.55		75
0-8-2	2.30		0.55		48
0-8-3	1.88		0.55		118

hot tearing exists when the iron is poured from higher temperatures under these melting conditions. This experimental evidence of the effect of temperature has often been experienced in the foundry under operating conditions.

Effect of Manganese-Sulphur Ratio on Hot Tearing: No detailed investigation of the influence of manganese-sulphur ratio on hot tearing was planned as a part of this project. However, in order to detect any major effect of this factor operating within the analysis range of the heats, a few trials were made. Two 3-lb heats were made under air atmosphere, one at a low manganese level, 0.33 per cent Mn, and one at a higher manganese level, 0.55 per cent Mn. The intermediate manganese level of about 0.43 per cent is the one existing in the majority of this work so that data for this level already existed. Data for six castings, three each from the heats at the low and the high manganese levels, are given in Table 9. Although they are not plotted, the hot tear values for the low manganese level, 0.33 per cent Mn, fall on the curve for 2700 F, Fig. 12. The hot tear values for the higher manganese level, however, are displaced on the left somewhat. Manganese-sulphur ratio has an effect, then, which is not significant at the low and intermediate levels but may be at higher levels. The comparisons in these experiments are

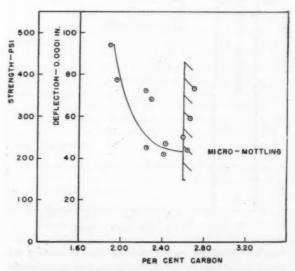


Fig. 11—Graph of resistance to hot tearing as a function of carbon percentage in the duplex iron melted under an air atmosphere and poured from 2900 F. Micro-mottling indicates the presence of primary graphite.

made at the intermediate level of about 0.41-0.45 per cent manganese, the effort being to hold the manganese content reasonably constant thus singling out the effect of the atmosphere being studied. Variation in the manganese content of the metal in all heats of this research project is thus within limits where the Mn-S ratio is not a factor in the atmosphere effects being studied.

## Furnace Atmosphere Effects on Pinholing

The fluidity casting provided some data on the tendency of the various atmospheres to cause pinholing. All heats made under hydrogen and water vapor atmospheres were offenders with respect to pinholing. The well of the fluidity casting (Fig. 2) revealed this defect whenever it occured. A photograph of pinholing in the fluidity well casting is shown in Fig. 13. The other gases, CO, CO₂, N₂, and A were not observed to cause this defect.

## Atmosphere Effects on Mottling

The effects of melting furnace atmosphere on tendency toward mottling are set forth in Table 10. These data are based on observations of the fracture of the well of the fluidity casting (Fig. 2) at a location one third of the distance up from the bottom of the well. Atmospheres produced by hydrogen and water vapor are shown in Table 10 to exert a profound carbide stabilizing effect. The other gases apparently did

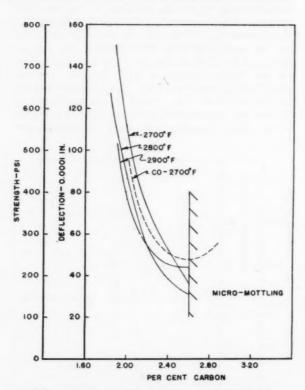


Fig. 12—Idealized curves showing the relationship between percentage carbon in the duplex iron melted under an air atmosphere and poured from 2700 F, 2800 F, and 2900 F. A curve for the effect of CO atmospheres at the 2700 F is included for comparison. Micro-mottling indicates the presence of primary graphite.

TABLE 10-EFFECT OF FURNACE ATMOSPHERE ON MOTTLING IN WELL OF FLUIDITY CASTING*

	Carbon Content % and Extent of Mottling								
Gas	Gray	Highly Mottled	Scattered Mottle	None, White Fracture					
CO	2.80		2.53	2.45					
CO ₂	2.77	2.60	2.50	2.46					
N ₂		2.86	2.59	2.53					
N ₂ A**		2.79	2.60	2.40					
H ₂				2.74					
H ₂ O				2.70					

* After holding for 60 min at 2700 F and with 1.27 to 1.32% silicon and 0.41 to 0.45% manganese in the iron.

** No castings available between 2.40-2.60% carbon.

not appreciably alter the maximum carbon percentage permissible for a fully white structure, 2.45 to 2.53 per cent carbon. This limiting percentage of carbon to prevent mottling under the latter gases might be primarily related to the silicon percentage in the iron, about 1.30 per cent on the average. It may be noted that the range of 2.45 to 2.53 per cent carbon maximum for mottling in the fluidity well agrees nicely with the observation that microscopic mottling is evident at about 2.60 per cent carbon in the hot tear casting. The latter casting is one which cools much faster and would favor suppression of mottling to the higher carbon level of 2.60 per cent.

## Discussion

The melting atmosphere has been revealed as capable of markedly altering the fluidity and hot tearing properties of the iron. It is difficult to explain the effects of the gases in changing these properties. At this stage of the project it is best to consider only the data at hand and refrain from theorizing. However, one factor is of major importance in both properties of fluidity and hot tearing, namely, the solidification process. Since this is true, it might be expected that a relationship between fluidity and hot tearing should exist. Such a relationship does exist, as shown in Fig. 14. The data for fluidity and hot tearing tendency obtained from carbon monoxide atmosphere heats presents a very nice correlation of the two properties as shown in Fig. 14. A satisfactory relationship also exists at a lower level for N2 and CO2 atmosphere heats, although the data for the latter shift to the curve for CO at high fluidity (higher carbon percentages). This tendency is also evident in the individual graphs of fluidity and hot tearing for CO2 heats, Figs. 4 and 9, respectively. The two curves in Fig. 14 are separated by a definite zone wherein the data fall sporadically. A possible reason for this is that the upper curve may represent conditions for metal saturated with carbon monoxide at 2700 F, while the lower curve represents metal in which the CO has been lowered to a limit attainable by holding 1 hr at 2700 F under the respective atmospheres. Thus, two distinctly different relationships between fluidity and hot tearing tendency appear to exist, depending on the melting furnace atmosphere. This would suggest the possibility of two different freezing mechanisms influenced largely by a smaller or larger amount of CO in the metal.

The data for hydrogen atmosphere heats are not plotted in Fig. 14 since they present an entirely different picture which has not yet been evaluated. Since hydrogen may have a definite alloying effect in iron-base alloys there may be good reasons for this difference in behavior as compared with the other gases.

It may be observed in Fig. 14 that occasionally data for other atmospheres fall on the curve for CO heats. This observation is especially true at high carbon contents. Saturation of the melt with CO is more likely at higher carbon contents, and might be occasionally encountered at lower carbon contents with the other atmospheres when equilibrium is not reached because of some melting factor.

Obviously, the influence of furnace atmospheres in causing changes in fluidity and hot tearing tendencies may be related to many factors other than freezing mechanism. For instance, the gases may produce surface films on the liquid metal, or the

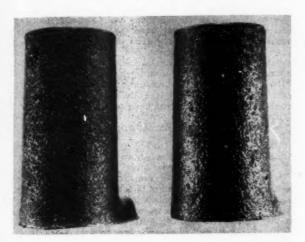


Fig. 13—Pinhole defects in the well of the fluidity casting typical of all heats made under hydrogen and water vapor. These defects were absent when melted under other gases, CO for example, as shown in the sample on the right.

formation and distribution of sulphide inclusions might be markedly altered, thereby changing metal properties. At this point, however, it is not possible to theorize beyond the limits of the data available at the time of writing this report.

## Summary

The results presented in this paper pertain to metal properties produced by holding a molten iron at 2700 F for 1 hr under atmospheres developed by carbon monoxide, carbon dioxide, nitrogen, water vapor, hydrogen, and argon. The principal conclusions which are possible may be classified according to their nature in the following groups:

## 1. Chemical Composition

(A) No losses of silicon or manganese occur when the melt is held at 2700 F whether the furnace atmosphere is oxidizing, neutral, or reducing.

(B) Carbon losses occur rapidly in oxidizing at-

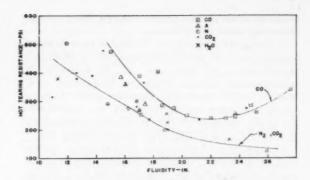


Fig. 14—Graph showing the relationship between hot tear and fluidity values obtained for each heat under CO,  $CO_2$ ,  $N_2$ , A, and water vapor atmospheres.

mospheres CO₂, H₂O, and air, and slowly in the reducing atmosphere, H₂.

## 2. Fluidity

(A) The relationship of fluidity and percentage carbon for metal melted under a CO atmosphere follows closely the fluidity data published in Reference 2 and established for a furnace atmosphere condition called standard air, i.e., a covered crucible.

(B) The atmospheres developed by carbon dioxide, water vapor, nitrogen, hydrogen, and argon all caused a decrease in metal fluidity, nitrogen being the most effective.

(C) A ½-hr period of holding the melt at 2700 F under a nitrogen atmosphere did not decrease fluidity as much as a 1-hr holding period.

## 3. Hot Tearing Tendencies

(A) Hot tearing resistance decreases with increasing carbon content in metal melted under all the conditions studied. The foregoing statement applies only as long as micro-mottling does not occur at the higher carbon levels.

(B) A carbon monoxide atmosphere over the molten metal raises the minimum hot tearing resistance encountered at 2.50-2.60 per cent carbon.

(C) A hydrogen atmosphere introduced over the metal markedly increases resistance to hot tearing.

(D) Hot tearing resistance of metal melted under atmospheres produced by water vapor and nitrogen is like that produced by air atmosphere melting.

(E) Increasing the melting temperature under air atmospheres lowers the resistance to hot tearing.

## 4. Pinholing

Metal melted under hydrogen and water vapor atmospheres resulted in extensive pinholing in the well of the fluidity casting.

## 5. Mottling

The atmospheres produced by hydrogen and water vapor markedly increased the maximum percentage carbon permissible without the occurrence of mottling.

## Acknowledgments

The authors wish to express their indebtedness to many persons for their assistance in carrying out the work reported in this paper. Among those offering

their help were the following:

1. Members of the A.F.S. Malleable Iron Research Committee—C. F. Joseph, chairman; W. D. McMillan, vice-chairman; H. Bornstein; W. A. Kennedy; R. Schneidewind; Milton Tilley; R. P. Schauss; S. C. Massari, A.F.S. Technical Director; J. H. Lansing. White iron sprue for melting stock was supplied by the group through W. D. McMillan, International Harvester Co., and The Pratt Manufacturing Co.

2. University of Wisconsin administrative leaders, Prof. K. F. Wendt, Associate Director, the Engineering Experiment Station, and Professor G. J. Barker, chairman of the Department of Mining and Metal-

lurgy.

3. Colleagues of the Department of Mining and Metallurgy—Prof. P. C. Rosenthal; Research Engineer Lew F. Porter; and Mechanician Ray C. Wiggen, who constructed the hot tear testing apparatus.

## **Bibliography**

1. E. A. Lange and R. W. Heine, "A Test for Hot Tearing Tendency," A.F.S. PREPRINT 52-14.

2. L. F. Porter and P. C. Rosenthal, "Factors Affecting the Fluidity of Cast Iron," A.F.S. PREPRINT 52-2.

3. R. W. Heine, "Oxidation-Reduction Principles Controlling the Chemical Composition of Molten Cast Irons," A.F.S. Transactions, vol. 59, p. 12 (1951).

4. E. A. Lange and R. W. Heine, "Some Effects of Temperature and Melting Variables on the Chemical Composition and Structure of Gray Irons," A.F.S. Transactions, vol. 59, p. 472 (1951).

## DISCUSSION

Chairman: W. K. Bock, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: F. B. ROTE, Albion Malleable Iron Co., Albion, Mich.

J. E. REHDER (Written Discussion): The subject of this paper is of major interest and is a field in which practically no careful research has been done, and the authors and the A.F.S. Malleable Iron Research Committee have made an excellent start. Much of course remains to be done, and it is therefore essential that the initial work be subjected to constructive criticism so that the whole work be of maximum effectiveness. The following remarks are in such spirit.

It is considered by the present writer that the holding temperature of 2700F is not well chosen, since in all melting and refining furnaces used in the malleable industry, bath temperatures of at least 2800F and more usually 2850F are used. This will affect both equilibrium data and the rate at which it is attained, and it is obviously desirable to have experimental data apply wherever possible to commercial conditions.

The method of applying and sampling the atmospheres used in the present work is believed open to question in its relationship to practical application. With relatively inert gases such as nitrogen and argon, the action on continu us passage of gas is essentially one of sweeping or flushing out of other gases, and the results obtained would then depend on the rate of flow of gas and the total time involved. At high rates of flow there would of course be a chilling action on the surface of the melt unless the gas was preheated. With other gases such as hydrogen, air, and carbon dioxide, which are active with respect to the iron, it is considered a misnomer to talk about "equilibrium" values with the techniques used since the results would appear to depend entirely on the experimental conditions. This is not to say that the results obtained are without value, but simply to suggest that they are relative.

The graphs presented, which show the effects of various gases

on fluidity and hot tear resistance, are open to subjective interpretation when considering where to draw lines among experimental points, and it is considered that the experimental errors involved do not justify such fine subdivisions as shown in certain of the curves in Figs. 5 and 6. The text of the paper, however, is more conservative in this respect and the summary as written would appear to be justified. One exception is the statement in section 3B of the summary, where the present writer does not agree that the data has shown that a carbon monoxide atmosphere increases the hot tearing resistance at 2.50-2.60 per cent carbon.

In determining the effect of iron temperature on hot tear resistance, the iron was melted to various temperatures and apparently poured at those temperatures. This confuses the matter since more than one variable is changed. It would seem more advisable to melt to and hold at various temperatures but to pour at a constant temperature. The results would have still greater practical value if the pouring temperature used were one commonly used in practice, for example 2600F. Such procedure would probably modify the results shown in Figs. 10, 11, and 12, but would be of more practical use.

The results noted under the headings of pinholing and mottling tendency are interesting and useful, the former being a good experimental demonstration of the relationship of hydrogen and pinholing, the latter being not surprising but at least quantitative for the conditions used. Malleable founders who might consider injecting water vapor into their furnaces, however, are cautioned that in an air furnace considerable water vapor is already present, and also that the effect of hydrogen on porosity and on hard iron cracks in commercial castings must be contended with.

To summarize, the above remarks are believed to be pertinent and constructive, and it is the writer's hope that this excellent and worthwhile program is to be continued.

## DISCUSSION BY MILTON TILLEY

MILTON TILLEY (Written Discussion) 2: Professors Heine and Lange have done an excellent job on this preliminary survey of this part of the field under investigation.

It is gratifying to have a demonstration of the effect of temperature on the oxidation of carbon, silicon and manganese and show why carbon is removed from the bath during the refining period while silicon and manganese remain unaffected during the period when the bath temperature exceeds 2500F.

The data on the effect of atmosphere on fluidity and atmosphere and temperature on hot tearing shows significant changes and indicates further investigation is in order. For instance, tabulation of the effect of atmosphere on fluidity at 2.4 per cent carbon and 2700F metal temperature shows—

## RESULTANT FURNACE ATMOSPHERE

Gas Used	Fluidity	02 C	$O_2$ CO	$N_2$	$H_2O$	$\mathbf{H}_2$	CH ₄	Argon
Air	22.20							
CO	21.00	0.8	98.2					
H,	17.55		13.0			85.0	2.0	
Argon	17.10	0.5	0.5					99.0
H ₂ O	17.00	5.0	42.0		0.0	45.0	8.0	
CÔ ₂	16.20	40.0	60.0					
No	16.20	0.5	0.5	99.0				

No analysis of the furnace atmosphere is presented when using air but it will probably contain CO and CO $_2$  at about 2 to 1 ratio,  $N_2$  and some  $O_2$ ,  $H_2$  and  $CH_4$ . Evaluation of effects becomes complicated with the effect of the action of the introduced gas and the effect of the resultant gases. For instance, CO gas gave a high fluidity, CO $_2$  gas gave a low fluidity and air and CO $_2$  gases result in an atmosphere containing both CO and CO $_2$ . Air gave a high fluidity and  $N_2$  a low, yet air contains  $N_2$ .

Looking at the effect of atmosphere on hot tear strength, again at the average 2.40 per cent and 2700F bath temperature we find.—

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² Metallurgist, National Malleable & Steel Castings Co., Cleveland.

#### RESULTANT FURNACE ATMOSPHERE

Gas Used	PSI	O ₂	CO2	СО	N _z	H ₂ O	H ₂	CH ₄	Argon
60	238		0.8	98.2					
CO ₂	256		40.0	60.0					
Air H ₂ O	269		5.0	42.0			45.0	8.0	
1			0.5	0.5	99.0				
H,	390		13.0				85.0	2.0	

Using a mean of 262 psi for normal operating conditions where CO₂, Air, H₂O, N₂ are present a high CO content results in about a 10 per cent reduction and a high H₂ content gives a 10 per cent increase. This may be significant to the foundryman.

Comparing the effect of temperature at 2.40 per cent C with the introduction of air we have—

Temp.	PSI
2700	262
2800	188
2900	225

The interesting point is do we have the least resistance to hot tearing at 2800 F, the temperature at which many of us tap.

Turning to the effect of carbon content at the three temperatures we find the following...

2700 F	PSI
2.20% C	356
2.60% C	$181 \pm 49\%$ reduction
2800 F	
2.20% C	262
2.60% C	162 = 38% reduction
2900 F	
2.20% C	275
2.60% C:	220 - 20% reduction

Here we find that the reduction of psi from 2.20 per cent C to 2.60 per cent C is less the higher the temperature, which confirms the past observations that higher temperatures reduces tendency to mottle. Here again we see the least resistance to hot tearing at 2800 F.

It is incumbent upon the steering committee to guide the future investigation along lines that will produce information that the foundryman can use to improve operating practices. With regard to the effect of atmospheres I believe that the effect of oxidation and reduction and water vapor should be followed up as being the things the operator can control. Also that the melting down conditions should be investigated in addition to the refining period, because the effects can be more pronounced if not different.

With regard to the effect of temperature of bath on hot tear strength, it would be in order to find out if the condition depicted at 2800 F is real and the cause.

The authors are to be congratulated on the work done so far. MESSRS. HEINE AND LANGE (Written Reply): Mr. Rehder's discussion will be considered first. Since the results presented in the paper represent research which is still under way and therefore incomplete, this paper should be viewed as summarizing the status of the work at the time the paper was written. Many of the questions which were unanswerable at that time will be clarified as the work continues. In answer to the temperature question, holding temperatures of 2700 and 2900 F are being studied. The selection of gases used in the laboratory study is such as to show the direction of any equilibria which may be attained by commercial melting. In other words, it is hoped that the extremes of changes of metal properties which might be caused by gases will be discovered. It would not be advisable in the beginning to melt and hold at various temperatures and then pour at another temperature. Since the chemistry of the melt changes with temperature, the authors have chosen the method of holding and pouring at the same temperature in order to eliminate this variable. No comments are believed necessary on the balance of Mr. Rehder's discussion. The authors wish to thank him for his thoughtful discussion.

The authors have enjoyed Mr. Tilley's analysis of the data in terms of its foundry application. Evaluation of this type will assist in the ultimate practical use of the information. It is believed that more satisfying analyses of the experimental data will be possible when more information is available. It is our hope that every effort will be spent in making the information of practical use when the principles are well known.

## A TEST FOR HOT TEARING TENDENCY

By

E. A. Lange* and R. W. Heine**

#### ABSTRACT

Hot tearing is a casting defect developed at high temperatures during cooling of the casting from the pouring temperature. Rupture of the casting occurs locally due to over-stressing

of the cooling and contracting metal.

Because hot tear defects are caused by a number of complex inter-related variables, evaluation of the principal factors is difficult. A survey of the literature on this subject reveals the problems encountered by numerous investigators. I.A.A. Principal among the problems has been the development of a test procedure for measuring differences in hot tearing tendency; especially a test which is able to separate factors that are peculiar to the metal or casting from factors that are peculiar to the mold in which it is contained. A testing procedure designed specifically for evaluating factors in hot tearing which are peculiar to the metal is described in the following pages. The hot tear test and some results obtained with this test as applied to white cast irons are presented in this report.

#### Factors in Hot Tearing

The factors which influence hot tearing may be classified in two principal groups:

1. Inherent characteristics of the metal.

2. External factors not properties of the metal. Considering the first group, hot tearing factors which are inherent properties of the metal are the following:

(a) Strength and ductility of the metal in the hot

tearing temperature range.

(b) Contraction of three types, namely, liquid, liquid-solid, and solid contraction with decreasing

temperature.

Hot tears occur only when stresses are produced in the metal which exceed its strength and ability to deform in the hot tearing temperature range. Stresses which cause a hot tear arise from restraint of the normal contraction propensities of the metal. In order that stresses may be developed the metal must have attained a certain amount of strength during cooling and some condition of restraint must be imposed. It is at this point that the second group of factors producing hot tearing comes into play, namely, those external to the metal. External factors include the following:

(a) Restraint of metal contraction caused by the

mold or cores.

(b) Restraint of contraction caused by the casting itself or its gating system.

(c) Temperature gradients or "hot spots" operating in conjunction with restraint of contraction.

The above three factors are functions mainly of the mold materials and construction, design of the casting and its gating system, and the temperature gradients which may exist. They are all factors separate from the inherent characteristics of the metal. Since this research was aimed principally at studying inherent differences in the metal, the second or external group of hot tearing factors (item 2, above) will not be dealt with. Rather, the hot tear test was designed so that the external factors as far as possible are held constant.

#### Mechanics of Hot Tearing

The influence of properties inherent in the metal are revealed by a more detailed consideration of the generally postulated mechanism of hot tearing. Molten metal in a mold cools and contracts in the liquid state as the temperature drops. As the temperature falls, freezing begins and is accompanied by freezing contraction. As freezing progresses, contraction continues. If the metal freezes over a temperature range, contraction of the liquid, solidification contraction and solid state contraction continue simultaneously until the metal has completely frozen. Early in freezing, liquid and liquid to solid state contraction predominate, but late in the freezing process solid state contraction becomes of greater significance.

During the later stages sufficient solid metal exists so that coherence of the solidified metal is developed. Coherence occurs when the solidying metal develops a measurable strength. When coherence or strength has been established during cooling it is possible for external conditions to produce stresses, by restraint of contraction, which may exceed this strength and cause rupture or hot tearing. Thus, it is seen that the high temperature strength and ductility of the

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Note: This is a progress report on an A.F.S. sponsored Research Project.

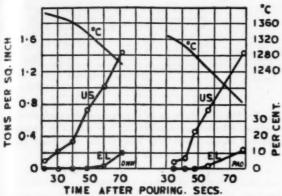


Fig. 1—Graphs reproduced from an article¹ showing the influence of temperature on the ultimate strength (U.S.), and elongation (El.), of cast steel tensile bars pulled while cooling from the casting temperature.

metal after coherence is established are extremely

Available literature suggests that hot tears develop at temperatures close to, just above or below, the end of the freezing range where strength and ductility are very low and that fracture occurs by brittle failure. No information is available which gives the strength or ductility of white cast irons in the hot tearing temperature range, but reference may be made to data on steels to illustrate the usual behavior of metals. Figure 1 shows data on high temperature strength and ductility of cast steel obtained by H. F. Hall¹ by casting steel tensile test bars which were

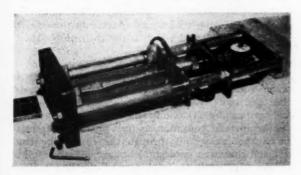


Fig. 2—Hot tear testing machine and casting with drag half of core sand mold in place.

pulled after various time intervals of cooling from the pouring temperature.

Strength is revealed in Fig. 1 as increasing much more rapidly than ductility as temperature drops, thus accounting for the apparently brittle failure of the hot tear defect. Although the metal may have considerable strength, stress in excess of this strength causes brittle failure since the metal has little ability to deform. While the data of Hall are not directly applicable to white cast iron, the principles appear to be illustrated. Once the metal has cooled in the solid state to temperatures where good strength and ductility are developed, hot tearing is no longer likely since the metal can resist the usual conditions imposed by a sand mold.

Hot tearing, then, is primarily a defect produced at high temperatures, i.e., at temperatures near or in the eutectic freezing range of white cast irons. It is for this reason that temperature gradients are so effective in producing hot tears because they serve to focus the effects of restraint of contraction at "hot spots" where the metal is weakest and least ductile.

#### Hot Tear Test

In view of the mechanism of hot tearing, any variables which influence the high temperature strength, ductility, and shrinkage characteristics of white cast iron will change its hot tearing tendencies. The major possibilities evident are listed below:

1. Chemical composition including all elements

present in the white iron.

2. Freezing temperature range; probably a major effect of carbon and silicon.

3. Possible evolution of gases during freezing.

Freezing mechanism, primary and/or eutectic freezing.

Grain boundary effects, non-metallic inclusions, for example.

6. Melting variables which may influence any of

It may be seen that the factors listed above are indirect ones in that it is their influence on the properties of high temperature strength, ductility and contraction that produces differences in hot tearing tendency. Thus, it would appear that a method of testing incorporating the effects of the high temperature strength, ductility and contraction properties of the metal, or their combined effects, may be a good measurement of hot tearing tendency. Furthermore, to successfully use the test it should have sufficient sen-

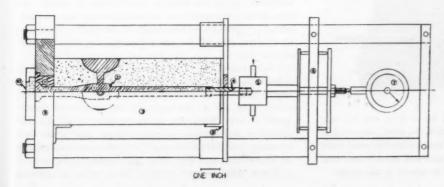


Fig. 3—Schematic diagram of hot tear testing device: (1) hot tear casting; (2) point of fracture; (3) core sand mold; (4) steel stud; (5) water cooler; (6) differential beam-type spring; (7) indicator gage, 0.0001 in. per division; (8) alignment frame; (9) base block; (10) end core; (11) chill plug.

sitivity so that changes in the effects of factors listed above are measurable. Development of such a test

is considered in the following pages.

The method of measuring hot tearing characteristics developed in the metallurgical laboratories of the University of Wisconsin comprises a testing machine and a test casting. The machine is designed to test the combined effects of hot strength, ductility and contraction in producing hot tearing in a standard casting. Figure 2 is a photograph of the testing machine. A schematic diagram of its construction is shown in Fig. 3. Detailed drawings of individual members of the hot tear testing device are given in the appendix.

The standard casting is cast in a core sand mold which fits into the testing machine as shown in Figs. 2 and 3. The core sand mold is made with two halves parted horizontally and a small third piece, (10) Fig. 3, which forms one end of the casting. The mold pieces and the core boxes are shown in Fig. 4.

#### Sand Mixture Characteristics

The core sand used for the mold consisted of 75 A.F.S. fineness number silica sand mixed with 1.20 per cent cereal, 1.80 per cent water, and 1.5 per cent oil based on 100 lb dry sand. The green molds were baked for 3 hr at 450 F. The casting produced in this mold may be seen in Figs. 2 and 3. Detailed dimensions of the standard casting finally adopted are given in Fig. 5. It consists of a straight ½-in. diameter bar 85% in. long with one section 2 in. long enlarged to 13/16-in. diameter to provide a "hot spot."

The joint of the ½-in. and ½-in. sections is made without a fillet to ensure breaking of the bar at the same location for each test, point (2) of Fig. 3. One end of the bar is fixed since it is cast into the rigid base of the machine. The other end is cast around a ¾-in. diameter by a 2½-in. long threaded stud. The stud is attached to a loading device. Contraction of the casting as it cools and solidifies actuates the load-

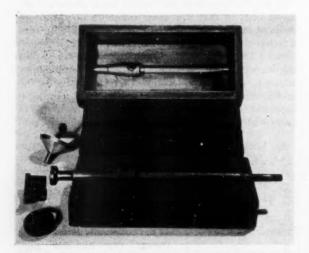


Fig. 4—Corebox and cores used for making the hot tear casting mold. The torn casting, with the sprue removed, is shown resting in the drag half of the mold. The sand mold is completed by the chill plug and end core at one end and the stud at the other end as shown in the photo.

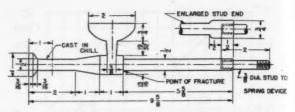


Fig. 5—Detailed drawing of hot tear casting with straight stud-end adopted as standard for the test. The enlarged stud-end shown above was used in the first tests made with the casting.

ing device which produces hot tearing and complete fracture of the test piece at the joint of the small

and enlarged sections.

Construction and functioning of the testing device itself may be readily visualized with the aid of Figs. 2 and 3. From Fig. 3 the structure of the apparatus is seen to consist mainly of the base block (9) and frame. The base block (9) contains a removable, split, chill plug (11) into which the fixed end of the test bar is cast. The open end of the chill plug is covered by an end-core and plate (10). The core sand mold (3) is supported in an alignment frame (8) which lines up the base-end opening of the mold with the chill plug opening and also lines up the stud-end opening of the mold with the connecting stud (4).

When the mold is put into the machine, the drag half is placed in the alignment frame first. Then the 3/8-in. diameter stud is screwed into the water cooler (5) with the free stud-end closing off the mold opening provided for it. Finally, the cope half of the mold is placed over the drag, the cope is weighted, and the mold is ready for pouring. After the casting is poured freezing occurs very quickly in the flanged base-end of the mold which is the chill plug (11).

#### Solidification Progress

Freezing is also rapid around the stud which is water cooled (5). Freezing thus progresses toward the "hot spot" which is the enlarged section of the hot tear casting (1). Contraction of the cooling casting begins and continues to pull on the differential beam-type spring (6) through the connecting stud (4) and the water cooler (5) linkage. Water cooling of the stud is an essential part of the design since expansion of the stud due to heating from the casting would offset the contraction of the casting. Deflection of the spring is followed by means of an indicator gage (7) reading in 0.00010-in. units. As the spring is deflected by the contracting casting, stresses produced by the pull of the spring rise to a level high enough to cause hot tearing and rupture of the bar.

Data taken after the casting is poured consists of deflection and time readings. Initially there is a slight back up of the spring due to sudden heating of the stud up to the temperature permitted by the water cooler. The time from pouring until the dial indicator leaves zero is determined, and then time readings are taken at regular deflection movements of 0.0005 in, and the final maximum deflection.

These data permit plotting of a curve of time vs.

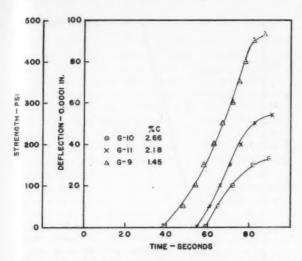


Fig. 6—Graph of spring deflection (casting contraction) as a function of time after pouring. The enlarged stud-end castings were poured from remelted duplex scrap at 2800 F and at three different carbon contents.

deflection or stress values up to the point of hot tear rupture as shown in Fig. 6. Deflection of the spring is related to the force applied to the bar, and therefore stress in the bar as pointed out in the vertical axis of the graph (Fig. 6). Maximum spring deflection or stress in the bar prior to hot tear rupturing is therefore a measure of the ability of the casting to resist hot tearing. Actual data obtained with this test and its interpretation will be considered later.

#### Important Characteristics of the Test

Certain aspects of the testing apparatus and the standard casting have an important bearing on its success.

One very important feature of the testing machine is the differential beam-type spring (6), Fig. 3. The load deflection characteristic of this spring is a critical factor in the testing device, and several springs were tested before a suitable one was found. Spring characteristics investigated are given in Table I. The spring adopted (called spring C) deflects 0.00010 in./lb. Calibration of the spring was performed by dead weight loading.

The importance of the spring characteristic is explained by the following observations. Hot tear rupturing of the casting will occur only if restraint by the spring causes stresses to increase at a greater rate than the strength increases as the casting cools and contracts. Thus, if the spring has too weak a characteristic it may never develop sufficient restraint to produce hot tears or fracture the bar. Briggs² reports this effect for spring-loaded cast steel bars and the present experiments verified it for white cast iron.

The first beam-type spring used in the testing machine, spring "A" as reported in Table 1, failed to rupture the bars because of its lack of stiffness. On the other hand, a spring of too great stiffness would cause hot tearing but would fail to give a sufficient range of readings to make measurements meaningful. In this sense, the spring in the hot tear

machine may be compared with an internal core in a mold. If the core is readily collapsible, hot tearing of the casting does not occur; assuming no other complicating factors.

If the core is extremely hard and rigid, hot tearing may occur in every casting. However, borderline core hardness and rigidity might result in sporadic hot tearing. Thus, the action of the differential spring in the hot tearing machine simulates the effect of restraint of the proper degree such as might be obtained from some cores. Obviously, it has the additional advantage of making quantitative measurements possible. The spring adopted for this research work (spring "C") may conceivably not be the best one for all purposes. However, time limitations imposed by the nature of the project did not permit evaluation of additional spring characteristics, and so the one selected was used as the basis for later investigations of hot tearing.

The shape of the hot tear test casting is another factor of importance which required experimentation. Bars varying from one having a greatly enlarged middle section to one with no enlarged section were tested. These bars are illustrated in Fig.

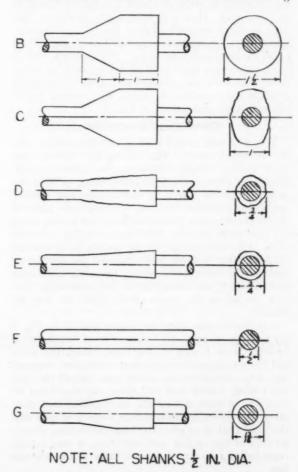


Fig. 7—Various designs of test bars studied with the hot tear testing machine. The rippled lines on the drawing indicate casting surfaces made by hand directly in core sand without a pattern.

TABLE 1-TEST MACHINE SPRING DEFLECTION

Type of Spring	Characteristic	Pre-load	Results
Beam_A	0.0020-in. deflection /lb	5-35 lb	Did not produce hot tear ruptur- ing.
Beam—B	0.00010-in, deflection /lb	20 lb	Zero reading was difficult to repro- duce
Differential Beam—C	0.00010-in, deflection /lb	0	Satisfactory

7. All bars were found to be sound except for shrinkage near the sprue, an area not affecting the point of fracture. It was found possible to produce hot tears in all of the castings, but they differed in their suitability to testing. The perfectly uniform ½-in. diameter bar was found to cool so rapidly as to make data taking difficult. On the other hand, the bar with maximum enlarged section gave rather indeterminate end points.

The most suitable bar tested was bar G in Fig. 5. This bar was used with two different stud ends. An enlarged end such as that shown in Figs. 2 and 5 was used initially in order to insure a complete casting at the stud end. However, later work showed it possible to cast a perfectly straight end, 1/2-in. diameter, around a 3/8-in. diameter threaded stud and obtain perfectly satisfactory bonding. The straight end is more satisfactory since it offers less restraint to movement than does the enlarged end, and consequently yields more reproducible data. The casting finally adopted is the one shown in Fig. 5.

#### Contraction Characteristics

The stud end of the casting involves rather complex thermal effects. The casting as a whole contracts during cooling. However, the stud expands. The latter effect could conceivably offset or alter the total contraction effects of the casting and confuse the results obtained. A study was therefore made of the free contraction characteristics of the casting using the steel studs and also silica rods to replace the steel studs. Free contraction of the casting as a function of time after pouring was observed by simply removing the differential spring from the apparatus and following the movement of the connecting linkages, including the water cooler, with the dial indicator.

Free contraction data are shown plotted in Fig. 8 for castings of equivalent carbon level having silica or steel studs frozen in the ends. Since the silica rod has an insignificant thermal expansion, comparison of its time-deflection curve with that of the casting having a steel stud will show any significant effects due to expansion of the water-cooled steel stud. Since the curves are identical in slope it is seen that the steel stud does not alter the contraction characteristics of the casting and, therefore, it may be disregarded as a source of possible error in the hot tear test.

Also shown in Fig. 8 is a curve for a casting allowed to solidify and contract against the differential spring C. Comparison with the free contraction curves shows that the spring causes the test casting to

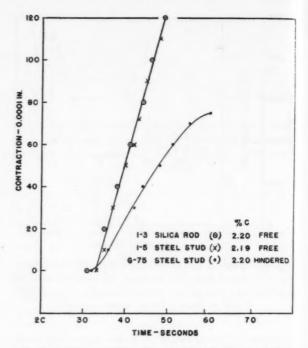


Fig. 8—Free and restrained contraction curves of 2.20 per cent carbon iron castings with straight stud-ends. Castings made of duplex iron scrap remelted and then poured from 2700 F.

elongate due to the stresses imposed up to the point of hot tear rupturing. The slope of the contraction curve, whether free or restrained, is the rate of contraction of the casting; measured in 0.00010 in /sec. As will be shown later, the contraction rate whether the bar is free or restrained is a function of percentage of carbon in white cast irons so that the foregoing comparison must be made at a given carbon level, 2.20 per cent in the case reported.

Experience with the hot tear testing procedure revealed other points of importance. The split chill plug (item 11, Fig. 3) requires a radius on the edge meeting the mold. Occasional flash at the junction of mold and chill plug causes hairline cracks to develop there and influence the hot tear deflection values obtained. This cracking was eliminated by filleting the edge of the plug as shown in the detailed

drawing in the appendix.

Another source of variation was found in the weighting of the cope half of the mold. A rigid clamp was employed and found very unsatisfactory. Hot metal caused the sand mold to expand against the clamp, which in turn caused the cope to bear against the long shank of the casting, thus seriously altering the hot tear data. A 4-lb weight on the cope was found satisfactory, as was also a clamp giving this amount of spring pressure to the cope. Experience with the aforementioned problems has resulted in the development of the test to its present condition. While the needs of the project for which the test was designed are met, it is believed that additional research on the test itself would produce refinements and even greater reproducibility of results.

#### Experimental Results Using the Hot Tear Test

First application of the test was to determine the effect of carbon content of white cast iron on hot tearing properties when other normal composition variables such as Si, Mn, P and S are held constant. Such information was considered essential to pursuit of the main project dealing with the influence of melting furnace atmospheres on malleable iron properties. Since certain furnace atmospheres will cause marked changes in carbon content of the metal, it is necessary to separate the effects of carbon percentage from those of the furnace atmospheres to be studied.

To determine the influence of carbon on hot tearing it was necessary to pour a series of test castings varying over a wide range of carbon percentages. A standard melting procedure was adopted for this purpose. Heats (3½ lb) of duplex iron scrap were melted in a 5-lb capacity induction furnace having a magnesia crucible. The exact steps in the melting cycle are tabulated below:

(a) Melt down 3.50 lb dup!ex scrap in covered crucible. Remove accumulated slag at 2500 F. About 55 min are required for this step.

(b) Heat to 2600 F in 10 min and add graphite and 0.10 per cent Mn as ferro-manganese. Electrode graphite was used to raise the carbon percentage from the melt-down analysis to any desired level.

(c) Increase temperature to the desired level, 2800 F in the initial work, and allow 10 min time for solution of additions with crucible covered.

(d) Remove crucible cover to permit air oxidation of carbon down to the desired percentage in about 10 min. Adjust temperature and hold at desired level during this period. Temperature measurement performed with platinum-10 per cent rhodium, platinum thermocouple.

(e) Cut the power input to the furnace and pour the first test bar. Turn on power to maintain temperature. Ten sec are required for the operation.

(f) Prepare hot tear testing apparatus for the next casting and pour when ready. Since the cover is off, oxidation of the carbon occurs during this period of 10 to 14 min.

(g) Repeat step (f) for the last casting from the heat.

The melting procedure described above provided hot tear data at three carbon levels from each heat. By varying the carbon additions and the oxidation periods it was possible to make castings over the entire carbon range of interest.

The melting stock available for the various heats consisted of one lot of 2 tons of duplex cupola-air furnace white cast iron sprue, and a second lot of coldmelt air furnace sprue; both provided through the efforts of the A.F.S. Malleable Division Research Committee. Pertinent information regarding the sprue is given in Table 2.

In the process of developing the hot tear test and obtaining the first results with the test, 135 castings were poured. Complete chemical analysis of each casting presented a time-consuming task. Accordingly, it was decided to analyze each one for carbon and make periodic checks of silicon and manganese percentages in the iron. Sulphur and phosphorus were checked only occasionally, since no change is anticipated in the latter two elements during induction melting.

On the basis of previous experience, it was known that silicon and manganese could be closely controlled if no undue fluctuations in the furnace charge occured. Chemical analysis of 27 castings of one series is given in Table 3. It is seen in this and subsequent tables that silicon and manganese are controlled in a satisfactory range, as is normally expected in this type of melting. The bars listed in Table 3 were of the G type shown in Figs. 5 and 7, having the enlarged end at the stud. As was pointed out earlier, the bar with the enlarged end was studied first and subsequently modified to the straight-end type.

Maximum spring deflection or stress and percentage carbon for each bar in Table 3 are plotted in Fig. 9. From this graph it can be seen that a greater contraction can occur and a greater force is required to produce hot tear rupturing at the lower carbon contents. The converse is true at higher carbon percentages as long as the bars are fully white. Whenever the carbon content was high enough to produce visual mottling, hot tearing did not occur by this testing method.

The change of deflection with time for individual test bars appears significant. Time-deflection curves typical of bars at three different carbon levels are shown in Fig. 6. Certain differences in the three curves of Fig. 6 are evident. The maximum deflection values decrease with increasing per cent carbon—the data plotted in Fig. 9.

The slope of the curves differs markedly with per

TABLE 2-TEST MELTING STOCK

Material			Analy	sis, %			
	C	Si	Mn	P	S	Cr	Remarks
Duplex sprue							Cupola Charge
Sample No. 1	2.53	1.34	0.47	0.113	0.148	0.03	Sprue, 34.%
No. 2	2.54	1.30	0.38	0.109	0.146	0.02	Malleable Scrap, 4.0%
No. 3	2.54	1.27	0.39	-	name.	-	Steel, 52.0%
							Malleable Pig, 6.0%
					- 1.5		Silvery Pig. 4.0%
							(0.002% boron added)
Cold melt sprue				-			Air Furnace Charge
Final analysis	2.48	1.10	0.37	0.124	0.085	0.007	Sprue, 59.8%
							Pig, 29.6%
							Rail, 10.6%

cent carbon. Slope at any point on the curve equals the rate of contraction at that moment. In the lower carbon material the rate of contraction is quite constant over a major part of the curve, but near the breaking point it decreases indicating yielding or tearing in the casting. On the other hand, the curve for the high carbon iron shows virtually no straightline portion. Its rate of contraction decreases rapidly to the point of fracture, seemingly indicating a mushing or yielding action by the metal.

From Fig. 6 it may also be noted that the time when the indicator pointer leaves zero, hereafter called the zero time, which corresponds to the time of positive coherence in tension, increases with increasing per cent carbon. Average time for zero deflection reading at various carbon levels is given in

The increase in zero-time is held to be due to delay in freezing caused by the greater percentage of eutectic liquid in the higher carbon iron. Thus, slope of the contraction curve appears to be a significant variable of the iron related to carbon content. The slope of the curve for each bar was determined for the initial or straight line part of the curve, and is given in Table 3. A graph of contraction rate (slope) and percentage carbon is shown in Fig. 10.

The curve shows a marked decrease in rate of contraction under these testing conditions as the carbon exceeds about 1.90 per cent, with minimum rates at maximum carbon percentages of 2.60-2.70. Since rate of contraction decreases with increasing

TABLE 3-ENLARGED STUD END BAR G-SERIES-2800 F POURING TEMPERATURE

asting No.*	Composition, 0		0-Time, sec	Max. Rdg., 0.0001 in.	Max. Time,	Initial Slope, 0.0001	
	C	Si	Mn				in./sec
1 :	2.66	1.22	0.356	56	38	86	1.50
2 :	2.38			53	25	79	-
3	1.74			51	69	81	2.67
4	2.63			55	33	85	1.36
5	2.23			54	45	89	2.00
6	1.54			48	96	95	2.35
7 :	2.59			58	44	92	1.76
8 :	2.24			55	45	86	2.10
9	1.45			38	93	87	2.20
10 3	2.66	1.24	0.398	59	33	90	1.31
11 :	2.18			55	54	91	2.10
12	1.61	1.24	0.364	54	91	100	2.67
13	2.71	1.23	0.409	59	34	86	1.36
14 5	2.46			55	39	97	1.74
15 5	2.19	1.23	0.394	54	37	92	_
16 2	2.40			49	44	86	1.66
17 2	2.14				30	87	_
18	1.97			43	48	75	_
19 5	3.02-	-mo	ttled	72	**	_	-
20 2	2.60			51	25	78	1.43
21 2	2.30			45	56	80	2.35
22 2	2.40			43	44	78	1.90
23 2	2.07			50	57	78	2.67
24 1	1.96			43	68	78	2.67
25 2	2.33	1.27	0.416	48	58	86	2.00
	2.13			52	50	75	2.50
	1.87			48	81	79	2.86

* Did not hot tear

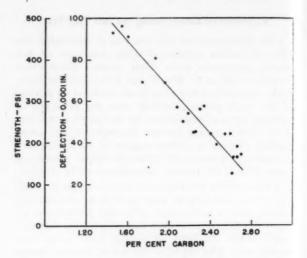


Fig. 9-Graph of maximum spring deflection or stress necessary to produce hot tearing and percentage carbon in the iron (Table 3). Enlarged stud-end type "G" castings were poured from remelted duplex iron held at 2800 F in an air atmosphere.

carbon percentage, it might be expected that total time for the specimen to break would increase. Experimentally this was verified, Table 3, since it was found that in the range of 1.71-2.00 per cent an average of 79 sec were required, while at 2.60-2.70 per cent, 87 sec were necessary.

The maximum contraction rate evidenced in Fig. 10 at about 1.80-2.00 per cent carbon probably is due to a change in freezing mechanism. Below this carbon range, eutectic freezing during final solidification would not be expected. However, above the carbon range of 1.80-2.00, eutectic freezing would play an increasingly important role with increasing carbon content. The aforesaid changes in freezing mechanism are also undoubtedly responsible for the effect of percentage carbon on resistance to hot tear rupturing as presented in Fig. 9.

The scatter of data in the graph (Fig. 9) prompted refinements in the test discussed in earlier sections of the report. However, it should be recognized that the movements of the hot tear casting are measured in increments of 0.00010 in. and, therefore, the scatter of data shown in Fig. 9 actually represents rather small differences in total dimensional movement prior

to hot tearing.

Improvement in the data was obtained with the use of a straight stud casting as shown in Fig. 5, and a split chill plug having a radius at the edge con-

TABLE 4-AVERAGE ZERO DEFLECTION READING TIMES

Carbon, %	Zero Time, sec*	
 1.41-1.70	46	
1.71-2.00	47	
2.01-2.30	51	
2.31-2.60	52	
2.61-2.70	57	

* Enlarged stud end bar G poured with duplex iron remelt from 2800 F.

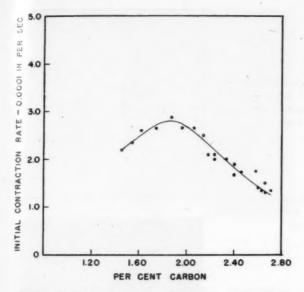


Fig. 10—Graph of slope of restrained contraction curve as a function of percentage carbon in test bar "G" having the enlarged stud-end and contracting against spring "C." Data are given in Table 3.

tacting the mold. The experiments were repeated using the same material and the melting practice outlined previously. However, the temperature of pouring was changed to 2700 F, and the percentage of manganese was raised slightly to obtain a better manganese-sulphur ratio. Chemical analysis and other data for the castings produced are given in Table 5.

The data of Table 5 are plotted in Fig. 11. The same relationship between percentage carbon and resistance to hot tearing exists in the data of Fig. 11 as in the previous case, but less scatter of data occurred. An additional feature may be noted in Fig. 11. A minimum strength or resistance to hot tearing was observed in the vicinity of 2.60 per cent carbon. Above this percentage the stress to cause hot tear rupturing increased and became erratic. Since it had been previously learned that hot tearing did

TABLE 5-HOT TEAR TEST DATA
Straight Stud End-Bar G 2700 F Pouring Temperature,
Duplex Melting Stock

Casting	Cor	mpositio	n, %	Max. Rdg.,	Initial Slope,
No.	C	Si	Mn	0.0001 in.	0.0001 in./sec
G 55	2.49	1.31	0.44	45	2.30
56	2.32			62	3.25
57	1.89	1.27	0.41	150	_
61	2.60			36	2.00
62	2.39			53	2.50
63	2.03			- 80	3.30
73	2.73			55	1.40
74	2.57			49	2.40
75	2.20			75	3.30
79	2.81			53	1.30
80	2.56			40	2.50
81	2.08	1.31	0.43	90	3.80
0-71	2.69			86	
72	2.45			44	
73	2.31			73	_
* 0.145-0	.159% S,	0.11%	P in ten	heats tested.	

not occur when the carbon content of the iron was high enough to cause visual mottling, it appeared likely that the effect noted above 2.60 per cent carbon was related to incipient mottling.

Microscopic examination of the casting in the vicinity of the hot tear fracture revealed the presence of primary graphite in varying degree in different castings containing over 2.60 per cent carbon. Examples of such primary graphite are shown in Fig. 12. This structural condition was given the name "micromottling" since it is not evident to the unaided eye if the white iron casting is fractured. Once micromottling occurs it is evident that a new freezing mechanism as well as a change in contraction characteristics of the casting will influence the data obtained with the hot tear test.

The graphite shown in Fig. 12 occurs in groups of flakes commonly identified as the mottling type of primary graphite, but it also appears as isolated flakes. The flakes are readily identified microscopically with polarized light, and they appear to be the forerunner of extensive visual mottling which occurs at higher carbon levels. Since the extent of micromottling depends on cooling rate and variations in melting practice, it follows that the resistance to hot tearing will fluctuate over wide limits at carbon contents above 2.60 per cent, with no hot tearing occuring when the mottling is visually evident. The foregoing observation is confirmed in Fig. 11.

Modification of the test casting to the straight studend type improved the reproducibility of the data to

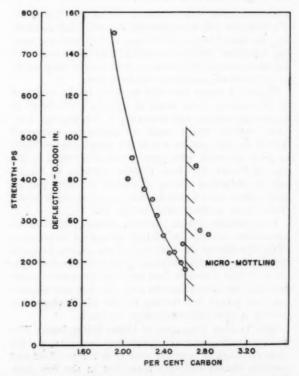


Fig. 11—Graph of stress necessary to cause hot tearing, and percentage carbon in the iron (Table 5). Straight stud-end type "G" castings were poured from remelted duplex iron held at 2700 F in an air atmosphere.



Left—Primary graphite of the cluster type in sample G-73. 150X, unetched. Right—Primary graphite of isolated flake type.



Fig. 12—Micro-mottling in test bars in Table 5 containing over 2.60 per cent carbon.



Left — Primary graphite of the isolated flake type in sample G-73. 750X, unetched.

Right—Same as photomicrograph at left, in polarized light. 750X.



the extent evident in a comparison of Figs. 9 and 11. The straight end also provided new evidence confirming the data relating percentage carbon and hot tearing strength. Stresses resulting from contraction of the iron around the threaded stud caused longitudinal hot tears to develop at the stud end.

Figure 13 shows hot tears occurring in the stud end of the casting. The width of the crack is related to percentage carbon and resistance to hot tearing. Carbon contents which result in deflection values of 0.0040 in. (200 psi) or less in the apparatus are likely to show the wide open type of tear illustrated at the top of Figure 13. Low carbon contents which result in deflection values of 0.0100 in. (500 psi) or more allow only a fine hairline hot tear crack or no crack at all, as illustrated in Fig. 13.

Intermediate carbon contents produced tears of intermediate width. In carbon ranges of micro-mottling, the tear at the stud end of the casting behaved erratically, probably depending on the degree of mottling. Thus, it is seen that very fine agreement exists between the data obtained with the hot tear apparatus and actual hot tearing of the metal when cast around a non-collapsible core, the stud.

Hot Tearing Properties of Other White Irons: The hot tear testing apparatus was next used to study the characteristics of a white iron of a composition and melting history different than that in the first case. The cold-melt sprue listed in Table 2 was used as melting stock for this purpose. Using the melting practice described previously, four heats were made to produce 12 castings which provided the data of

Table 6, which is plotted in Fig. 14. A pouring temperature of 2700 F was employed.

The curve in Fig. 14 reveals that the same relationship of percentage carbon and hot tearing tendency prevails in this iron as did in the first case studied. Some differences in the curves of Figs. 14 and 11 are to be expected because of differences in chemical composition of the iron and its prior thermal history. Chemical analyses of the castings are given in Table 6. Principle chemical differences between this iron and that produced from the duplex sprue is the lower sulphur content, 0.085 as against 0.145 per cent, and lower silicon content, about 1.10 per cent as compared with about 1.30 per cent.

TABLE 6-HOT TEAR TEST DATA

Straight Stud End Bar G-2700 F Pouring Temperature,
Cold-Melt Melting Stock

Casting No.*		Composition, %*		Max. Rdg.,
	C	Si	Mn	0.0001 in.
H-1	2.53	1.07	0.41	51
2	2.20			58
3	1.62			139
4	2.33			55
5	2.11			93
6	1.71			
7	2.74			55
8	2.39			46
9	1.90			_**
10	2.64	1.10	0.40	55
11	2.33			53
12	1.93			132
* 0.087% S, 0	0.126% P.			
Flash at	parting lin	e, no data.		

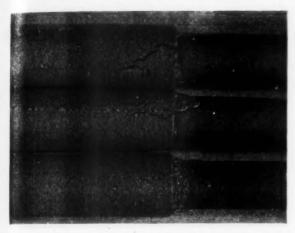


Fig. 13—Hot tear cracks at the straight stud-end in type "G" castings. Wide crack in the upper casting occurs at high carbon content, 2.60 per cent, sample G61. Absence of crack or narrow, hairline crack occurs at low carbon content, 1.89 per cent, sample G-57. Intermediate carbon contents produce cracks such as shown in the middle casting.

In spite of the chemical differences in the irons their resistance to hot tearing is similar, although castings made from the cold-melt iron sprue did not give as low heat tearing strength values as the duplex iron sprue did in the range of 2.50 to 2.60 per cent carbon. A comparison of Figs. 11 and 14 shows that the hot tearing resistance of both irons is almost identical over the range of 1.90 to 2.20 per cent carbon. Differences in the two irons become more pronounced in the range of 2.20 to 2.60 per cent carbon, possibly due to a greater effectiveness of silicon and sulphur at the higher carbon level.

The agreement as well as the differences between the curves of Figs. 11 and 14 are good evidence of the ability of the hot tear testing apparatus to give reproducible results and yet detect variations in the metal when they exist.

Contraction Rate of Straight End Bars: The effect of carbon content on the slope of the contraction-time curve for the test casting having the straight stud end should be the same as for the bar having the enlarged end (Fig. 10). If contraction rate variation with change in carbon percentage is a fundamental property of the iron it should occur independently of the casting, although the magnitude of the contraction rate may be affected by the casting design and cooling rate. Furthermore, the contraction rate should vary with carbon content whether the test bar is restrained or free.

The foregoing observations were found to be true in experiments with the straight stud-end bar. The relationship of carbon content to contraction rate in test bars freely contracting, and also those contracting against spring C is illustrated in Fig. 15. The data reported in Fig. 15 for bars cast in restraint were obtained by determining the slope of the time-deflection curves of the castings used in establishing the hot tearing strength vs. carbon content curve, Fig. 11 and Table 5. The contraction rate relationship

to carbon content is thus established as a fundamental property of the iron.

Hot Tearing Temperature: The temperature of hot tearing at the point of fracture of the test casting was determined. In addition a survey was made of temperatures prevailing in the casting at the time of tearing. Thermocouples were positioned on the center line of the mold at points of interest as shown in Fig. 16. A chromel-alumel thermocouple coated with refractory cement was embedded at one point in each casting. After pouring the casting, deflection and temperature data were taken simultaneously until hot tearing occurred.

Temperature as a function of time was recorded by means of a high speed electronic potentiometer recorder. The recorder was started at the moment of zero-time, the first positive movement in tension registered on the dial indicator of the hot tear apparatus. The aforementioned procedure was employed to correlate the temperature and deflection data on the time basis. Analyzing the temperature-time curves and the deflection-time curves obtained by this means provided the data given in Fig. 15. The temperature of hot tearing at the point of fracture of the standard casting was concluded to be 2030 F.

Figure 16 also shows that the indicated temperature at the time of the tear is at a maximum at the point of fracture and decreases in either direction from the tear. The temperature gradient results from the chilling action of the base on one end and the stud on the other end. Thus, hot tearing at the joint of the small and large sections is greatly favored by the sharply defined "hot spot" which fortunately exists in the casting. The amount of liquid metal re-

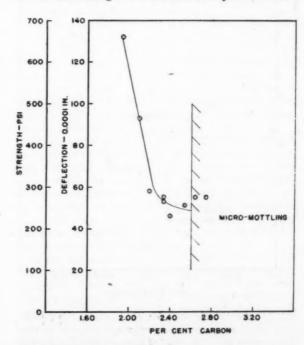


Fig. 14—Graph of stress necessary to cause hot tearing and percentage carbon in the iron (Table 6). Straight stud-end castings were poured from remelted cold-melt iron sprue held at 2700 F in an air atmosphere.

maining at each point at the time of hot tearing was calculated and is reported in Fig. 15.

The calculation was based on the recorded cooling curve at each point in the casting. The percentage of solidified eutectic at the time of tearing was determined by dividing the total time interval of eutectic freezing into the time interval of eutectic freezing prior to hot tearing, and multiplying by 100. The total percentage of solid metal present at the time of hot tearing was calculated for each carbon content from the relationship of percentage carbon to percentage of eutectic metal in iron-carbon, one per cent silicon alloys. The table in Fig. 16 points out that hot tearing occurs during the later stages of eutectic freezing and prior to complete solidification at the location of tearing.

#### Carbon Content Vs. Hot Tearing

The occurrence of hot tearing during the freezing of the eutectic serves to explain the shape of the curve relating carbon content to resistance to hot tearing (Fig. 11). In the lower carbon ranges much more solid metal forms prior to eutectic freezing than solidifies in the higher carbon iron. The amount of solid metal formed prior to eutectic freezing determines the strength developed prior to hot tearing in the eutectic temperature range. The lower carbon

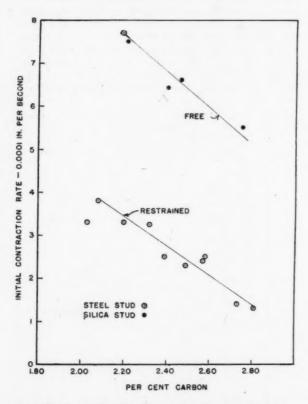
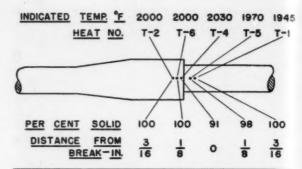


Fig. 15—Graph of contraction rate (slope of deflectiontime curve) as a function of percentage carbon in the iron. Upper curve gives the effect of carbon content on contraction rate of freely contracting bars. Lower curve gives the contraction rate of bars restrained by spring "C", data in Table 5. All castings were poured from 2700 F with remelted duplex iron sprue.



Heat	Carbon,	Temperate Eutection, Arrest		Percentage of Solid at Time of	Percentage of Eutectic Solidified at
No. %	Start	End	Tearing	Time of Tearing	
T-1	2.45	2065	2030	100	100
T-2	2.62	2050	2020	100	100
T-4	2.45	2055	2020	91	60
T-5	2.19	2130	1945	98	76
T-6	2.24	2105	2010	100	100

Fig. 16—Diagram showing the temperatures existing in the straight stud-end type "G" casting at the time of hot tearing.

irons therefore have much more resistance to hot tearing, i.e., they have more high temperature strength above 2030 F.

Furthermore, mottling at high carbon contents will increase the high temperature strength because it occurs during the freezing of the eutectic liquid. Depending on the degree of mottling the amount of contraction occurring during the freezing of the eutectic will be greatly influenced, decreasing with increasing mottling tendency. Decrease in contraction during eutectic freezing will, of course, increase the strength since it permits more metal to solidify before producing stresses high enough to cause hot tearing. The influence of carbon contents on mottling, high temperature strength, and contraction rate thus appear to be primary factors influencing the hot tearing tendencies of white cast iron.

#### Discussion

The results obtained with the hot tear testing apparatus indicate a relationship between per cent carbon and the strength and contraction characteristics of white cast iron. The latter two properties, plus ductility, are combined in their effect in the particular test casting studied. The test casting figures greatly in the results obtained since its dimensions and cooling characteristics serve to focus the effect of restrained contraction at the intended point of fracture. The intent of the casting design is to cause hot tearing at a location reproducible in all castings.

The test piece design adopted has been demonstrated by the data as one which will provide a definite point of hot tearing and one which will reflect differences in the strength and contraction properties of the iron. Furthermore, temperature gradients at the location of the hot tear show that a definite hot spot exists at the location of tearing. Conditions which involve all the inherent metal properties af-

fecting hot tearing are therefore considered to prevail in the test casting adopted.

The loading spring is another factor which infinences the results obtained with the apparatus. Its deflection-load characteristic determines the stress developed in the test bars. The effect of the loading spring on the data was experimentally verified when tests showed a spring weaker than the "C" spring was found unable to cause hot tear rupturing. Returning to the casting, it is evident that the length of the casting determines the total contraction at any moment of cooling and, therefore, the load applied to the bar by any given spring.

The combination of contraction by the casting and restraint of contraction by the spring causes hot tearing of the casting. The actual load which will cause hot tear rupturing of the bar is a composite result of contraction and loading which can vary, depending on the characteristics of the test bar and spring employed. The combination selected in this apparatus was one which gave a suitable range of values for

experimental purposes.

Because the aforementioned factors all enter into the data obtained with the present hot tear test, it must be recognized that the experimental results are relative in nature. For example, the stress values plotted in Figs. 9, 11 and 14 may not represent the true strength of the metal, since they are calculated stress values based on spring deflection relative to the particular casting (type G) being studied.

#### Tests Indicate Trends

The value of the data rests principally in the possibility of making comparisons involving the particular casting. In this sense, the hot tear test is like the spiral fluidity test in that a combination of effects is measured. Therefore, in considering the graphs and other observations reported it should always be recognized that they were obtained by the particular procedure employed and that their magnitude would be altered by other combinations of testing condi-

General trends are, however, held to be indicated by the present tests. In addition it appears that any variables such as chemical composition, melting practice, melting furnace atmosphere, etc., which might have a notable effect on the elevated temperature strength, ductility or contraction characteristics of the metal would be revealed by the hot tear test in

its present state.

Foundry Applications: Interpretation of the test results in terms of foundry hot tear defects requires caution. The data show decreased strength and failure at a smaller contraction value at higher carbon contents in the iron. However, it cannot be concluded that a casting which is hot tearing in the range of 2.50-2.60 per cent carbon would show marked improvement or cease hot tearing when the carbon is lowered to 2.20-2.30 per cent. Such an expectation would assume that the only factor influencing resistance to hot tearing is carbon content of the iron.

One reason for development of the test was to study the influence of melting furnace atmospheres on the hot tearing characteristics of white irons. By employing the test it has already been discovered that high carbon iron melted under one atmosphere may be stronger than low carbon iron melted under another atmosphere. These results are reported in another paper.6 It is evident that many factors other than carbon content of the iron may influence hot tearing in a specific application in the foundry. For instance, the many complicating factors introduced by the mold and cores may overshadow differences in the metal.

Considering only inherent properties of the metal related to carbon content, those which favor decreased hot tearing tendency at the extremes of carbon percentages are tabulated below:

#### FACTORS FAVORING DECREASED HOT TEARING TENDENCY Low Carbon Iron High Carbon Iron

- 1. Higher strength before failure
- 1. Lower contraction rate Greater deflection or strain 2. Longer time period from pouring to the beginning of coherence
  - 3. Micro-mottling

In actual foundry conditions the above listed factors may become important to varying degrees depending on the nature of the mold and casting. For example, when the metal is absolutely restrained as by a metal core, its strength and ability to withstand strain before failure are very important. This was experimentally demonstrated in the present work from observations of cracks developed in the test bar end containing the stud (Fig. 13). When the 3/8-in. diameter steel stud was cast into a 1/2-in. diameter test bar end, wide tears always occurred in the end at the high carbon levels but were absent or very fine when at the low carbon level.

As another case, consider a mold containing cores having a certain degree of collapsibility. The higher carbon material being molten and noncoherent for a longer time might permit sufficient burn-out and collapse of the core to minimize restraint and consequent hot tearing. Under the same conditions, a lower carbon composition becoming coherent more quickly might not give sufficient time for core collapsibility to become a factor.

On the other hand, another weaker core might be mechanically compressed by the contracting iron and hot tears thus reduced by the stronger low carbon metal, while the weaker high carbon material might hot tear consistently. The foregoing conditions are all postulated on the basis of variations in mold and core conditions which may cause one of the properties of the metal to become determinant in aggravat-

ing or minimizing hot tearing.

Actual evaluation of mold and core effects on hot tearing in castings requires a knowledge of their collapsibility characteristics which is not now available. The casting shape and size contribute additional problems. These factors of mold and casting which are external to the metal complicate the problem and must be considered when trying to relate test results to the foundry defect. The principal aim of the test has been to isolate and determine variations in the magnitude of properties inherent in the metal and entering into hot tearing. This aim has been accomplished and a tool provided which, with some modifications, may be used to study many of the factors which influence hot tearing.

#### Summary

A testing method for evaluating inherent metal properties affecting hot tearing has been presented in this report. Employing this hot tear testing procedure the following conclusions were reached:

1. In the particular white irons investigated, strength or resistance to hot tear rupturing decreases with increasing percentage carbon in the iron.

2. Less strain occurs prior to hot tear rupturing as percentage carbon in the iron increases.

3. Time interval from pouring to the development of positive coherence increases with increasing per-

centage carbon.

- 4. The rate of contraction of the test bar after coherence develops is a function of percentage carbon in the iron whether the bar contracts freely or is restrained. Contraction rate increases with decreasing carbon content to a maxium rate at 1.80–2.0 per cent carbon, and then decreases within the limits investigated.
- 5. The test casting elongates under restraint during cooling.
- 6. Iron containing mottling visible in a fractured surface does not hot tear under the conditions of these tests.
- 7. Micro-mottling occurs at carbon contents lower than those which produce visual mottling and results in an increase of the resistance to hot tearing.

The hot tearing properties of castings made from two different melting stocks were evaluated.

9. A visible relationship exists between the hot tear crack developed at the stud end of the casting and the magnitude of the stress necessary to cause hot tearing as measured by the testing apparatus, low resistance to hot tearing and wide tears occurring at 2.50-2.60 per cent carbon.

10. The temperature of hot tearing is about 2030 F in this casting. Hot tearing occurs during eutectic freezing and while about the last 10 per cent or less

of the metal is solidifying.

In conclusion, it may be said that a test which is capable of detecting changes in the inherent hot tearing properties of white cast iron has been developed. Being based on fundamental properties, the test should be adaptable to the study of the inherent characteristics which influence hot tearing in many other metals.

#### Acknowledgments

The authors wish to express their indebtedness to many persons for their assistance in carrying out the work reported in this paper. Among those offering

their help were the following:

1. Members of the A.F.S. Malleable Iron Research Committee—C. F. Joseph, chairman; W. D. McMillan, vice-chairman; H. Bornstein; W. A. Kennedy; R. Schneidewind; Milton Tilley; R. P. Schauss; S. C. Massari, A.F.S. Technical Director; J. H. Lansing. White iron sprue for melting stock was supplied by the group through W. D. McMillan, International Harvester Co., and The Pratt Manufacturing Co.

2. University of Wisconsin administrative leaders, Professor K. F. Wendt, Associate Director, the Engineering Experiment Station, and Professor G. J. Barker, chairman of the Department of Mining and Metallurgy.

3. Colleagues of the Department of Mining and Metallurgy—Professor P. C. Rosenthal; Research Engineer Lew F. Porter; and Mechanician Ray C. Wiggen, who constructed the hot tear testing apparatus.

#### Bibliography

- 1. H. F. Hall, "The Strength and Ductility of Cast Steel During Cooling from the Liquid State in Sand Moulds," Second Report of the Steel Castings Research Committee, The Iron and Steel Institute.
- C. W. Briggs, "The Metallurgy of Steel Castings," McGraw-Hill Book Co., New York.
   D. C. G. Lees, "Factors Controlling the Hot Tearing of
- 3. D. C. G. Lees, "Factors Controlling the Hot Tearing of Aluminum Casting Alloys," Foundry Trade Journal, Aug. 18, 1949.
- 4. A. R. E. Singer and P. H. Jennings, "Hot Shortness of the Aluminum Silicon Alloys of Commercial Purity," *Journal*, Institute of Metals, 1947, 73, 197.
- 5. W. E. Sicha and E. E. Stonebrook, "Correlation of Cooling Curve Data with Casting Characteristics of Aluminum Alloys," A.F.S. TRANSACTIONS, p. 489, 1949.
- A.F.S. TRANSACTIONS, p. 489, 1949.
  6. E. A. Lange and R. W. Heine, "Some Effects of Temperature and Melting Variables on Chemical Composition and Structure of Gray Irons," A.F.S. Transactions, vol. 59, p. 472 (1951).

#### **APPENDIX**

Detailed drawings of individual parts of the hot tear testing apparatus are set forth in the accompanying illustrations. No complete assembly drawing is given since it is believed that the individual parts may be readily assembled with the help of the photograph and schematic drawing of the apparatus (Figs. 2 and 3) in the first section of this paper.

The parts detailed in the drawings are:

1. Chill end block (base)

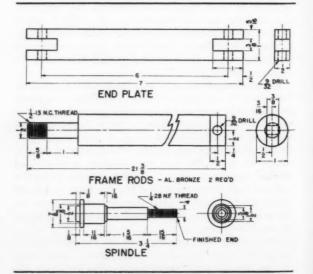
2. Split chill plug

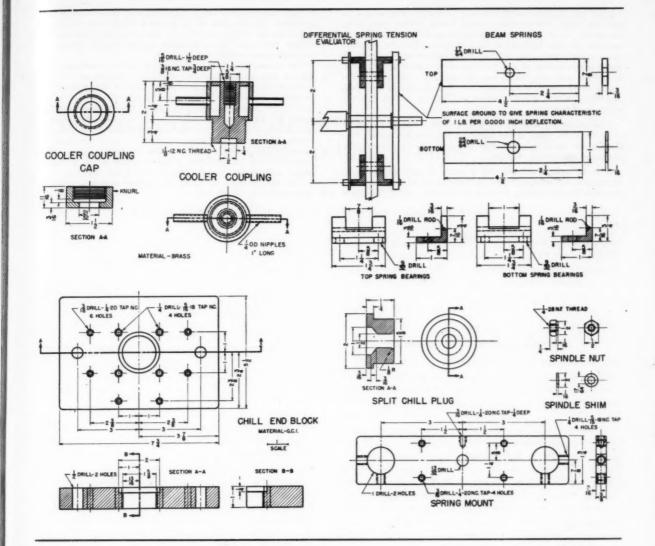
3. Frame rods and end plate

4. Cooler coupling

5. Spring mount

- 6. Differential spring tension evaluator and parts
- 7. Beam spring
- 8. Spindle and parts.





#### DISCUSSION

Chairman: W. K. Bock, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: F. B. ROTE, Albion Malleable Iron Co., Albion, Mich.

J. E. Rehder (Written Discussion): ¹ The authors, and the A.F.S. Malleable Iron Research Committee, are to be commended for the attempt described in this paper to evaluate quantitatively the practical foundry problem of hct-tearing. It would appear that a useful approach has been made, and that a practical test for hot-tearing tendency has been developed. With further refinement of the technique, many of the former use!ess arguments about hot tears can be settled or avoided by actual experimental test.

It may appear difficult to analyze precisely what is being measured by the test procedure, since contraction and stress or load are interacting, but since these conditions are similar to those probably existing in an actual casting solidifying in a mold, the test results should have practical value. It has been the writer's impression that in malleable foundry practice hot tears were more likely to occur at lower carbon contents, which is the reverse of the findings in the paper under discussion, but in a commercial casting the conditions are complicated by stresses set up by unequal cooling rates at various locations.

¹ Foundry Engineer, Bureau of Mines and Technical Surveys, Ottawa,

The authors have shown that hot tearing occurs at or near the eutectic temperature when the iron is nearly completely solid, and the results in Fig. 9 of hot-tearing stress versus carbon content could have been plotted as stress versus per cent eutectic.

The maximum in contraction rate noted in Fig. 10 at about 1.80 per cent carbon content is interesting, and, when the effects of silicon content on the composition points of the iron-carbon diagram is taken into account, corresponds to a carbon content of 2.0 to 2.2 per cent in iron-carbon alloys. This is approximate confirmation of the recent change in the iron-carbon diagram of from 1.7 to 2.0 per cent carbon as the point of zero eutectic content.

Comparison of Figs. 11 and 14, where the only important differences are in history of the raw material and in silicon content, shows that micro-mottling occurs at the same carbon content for both 1.10 and 1.30 per cent silicon content. One normally would expect that the mottling would occur at higher carbon content in the lower silicon iron, especially on fresh melts under similar conditions. Also, the minimum value of strength is lower for the higher silicon content iron, whereas silicon is usually regarded as a strengthener of iron. The author's comments on these points would be appreciated.

MILTON TILLEY (Written Discussion): 2 This method of test-

² Metallurgist, National Malleable & Steel Castings Co., Cleveland.

ing hot tear tendency is the best the writer has seen so far. It may not satisfy the physicist because of the stresses other than tension set up at the time of deformation and the inability to measure the area involved at time of tearing but the foundryman is interested in the relative tendencies to tear induced by metal characteristics of which he has some control and this test procedure indicates that it may do that. For instance, it has evaluated in a more quantitative way than heretofore the effect of carbon content and has shown a difference in two different methods of melting.

It is anticipated that the test will measure more quantitatively the correlation noted in the foundries between hot tearing tendencies and moisture content of the atmosphere and with the oxidation rate during melting.

It is the writer's opinion that it would be well to investigate thoroughly the effects of melting conditions as being a more fertile field than refining for most foundrymen realize that there is where metal for malleablizing is endowed with its main characteristics. At this stage of the art, it is known that mainly the properties are influenced more by oxidizing and reducing conditions than any other. Moisture content of the atmosphere probably comes under this category.

The authors are to be congratulated on the development of the test apparatus and on its preliminary application.

Messrs. Heine and Lange (Written Reply): Mr. Rehder's observations regarding the influence of carbon content on hot tearing are interesting. Since the paper reports progress on research still in progress, the full story is not told. It is believed Mr. Rehder's question will be adequately answered when the results of current work are published. Evidence has been obtained which shows that depending on furnace gases, white irons may tear more readily at lower carbon contents than at higher carbon contents, the observation of Mr. Rehder.

The authors wish to thank Mr. Rehder and Mr. Tilley for their discussion of the paper.

## INVESTIGATION OF PROPERTIES OF STEEL SANDS AT ELEVATED TEMPERATURES

R. G. Thorpe,* A. E. Riccardo,** P. L. Widener** and P. E. Kyle***

#### ABSTRACT

In the production of sand castings the surface layers are subject to rapid heating to rather high temperatures. The authors, in an earlier report on this project, expressed the opinion that if the mold-metal-interface conditions are such that the expansion of the sand results in a deformation which exceeds the ability of the sand to deform at that temperature without fracture, a surface defect will result.

This report shows conclusive laboratory evidence that the maximum temperature where the expansion exceeds the allowable deformation, varies with steel sand mixtures. On the basis of observations of scabbing tendencies in the foundry, it has been shown that as this temperature increases the scabbing tendencies increase.

#### Introduction

This report covers the work on the Sand Research Project sponsored by the Sand Division of the American Foundrymen's Society which has been completed at Cornell University during 1950 and 1951. Part I is devoted to the laboratory stress-strain and expansion studies of several synthetic steel sand mixtures at elevated temperatures. Part II is a summation of test casting results on the scabbing tendencies of the sand mixtures tested in Part I and an extension of work previously reported upon in the Tenth Annual Report. The appendices include additional data on problems which relate to the main objectives of the program.

The authors wish to acknowledge the many helpful suggestions and the guidance given by the A.F.S. Sand Division Committee on the Physical Properties of Steel Foundry Sands† at Elevated Temperatures, under the chairmanship of J. H. Lowe. The Sand Research Project is indebted to the New Jersey Silica Co., to Eastern Clay Products, Inc., and to the American Colloid Co. for all of the raw materials used in the investigations. The willing assistance of the Harry W. Dietert Co. in obtaining special items of equip-

ment is acknowledged.

The opportunity to test out the theories developed in the laboratory was made possible by the splendid cooperation on the part of the Wehr Steel Company (J. H. Lowe) and The American Steel Foundries (J. A. Rassenfoss), in whose plants all of the castings discussed in Part II were poured.

### PART I Stress-Strain and Expansion Test Data

The procedures used for the preparation of the test mixtures, as well as the techniques employed in specimen preparation and testing, have been described in previous progress reports, but briefly they are as

Mixture preparation: Base sand and clay are dried at 220-230 F for at least 1 hr and cooled to room temperature. Batch size is maintained at 5000 gm for a 24-in. Simpson Laboratory Muller with plow clearance of 1/16 in. The base sand is added to the muller, temper water added slowly and the mixture mulled for 5 min. Clay is uniformly distributed and the mixture mulled for 10 min. The mixture is then dumped, riddled through a four-mesh riddle directly into glass storage jars and covered tightly. The batch is allowed to stand for at least 2 hr prior to further use.

Specimen preparation: (a) 11/8-in. diam by 2-in. cylindrical test specimens are double-end rammed for stress-strain testing in a split specimen tube with three blows of the standard A.F.S. rammer. The green specimen is placed between mullite disks, inserted into the dilatometer furnace and tested in accordance with the procedure outlined in Appendix A, using an

exposure time of 20 min.

(b) The 1-in. x 1-in. x 8-in. expansion test specimen is rammed to the same density (as determined by sand weight) as the 11/8-in. x 2-in, stress-strain specimen (refer to Appendix E), transferred to a fused quartz plate and inserted into the heated hearth of the research furnace. Measurements of specimen length are taken optically by means of a telescope mounted externally to the furnace and the expansiontime curve plotted.

This phase of the research program has been concerned with mixtures formulated from the New Jer-

^{*}Asst. Prof. of Metallurgy, **A.F.S. Investigator and ***Prof. of Metallurgy, Cornell University, Ithaca, N. Y.

[†]Committee 8-L-J. H. Lowe, Chairman; H. M. Kraner, C. W. Briggs, H. W. Dietert, R. H. Jacoby, C. B. Jenni, R. E. Morey, J. A. Rassenfoss, H. F. Taylor, D. C. Williams.

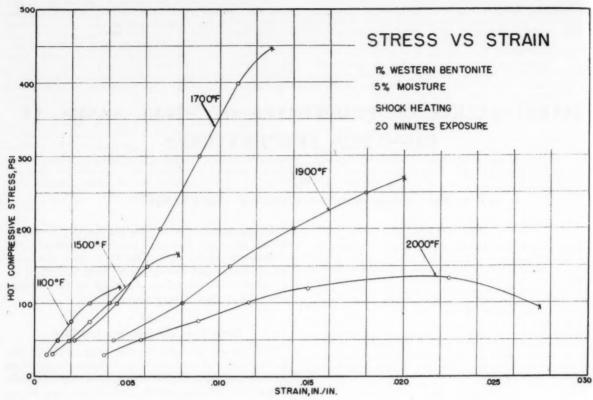


Fig. 1

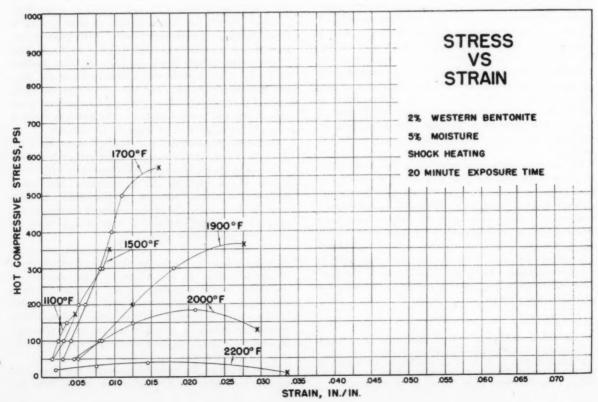


Fig. 2

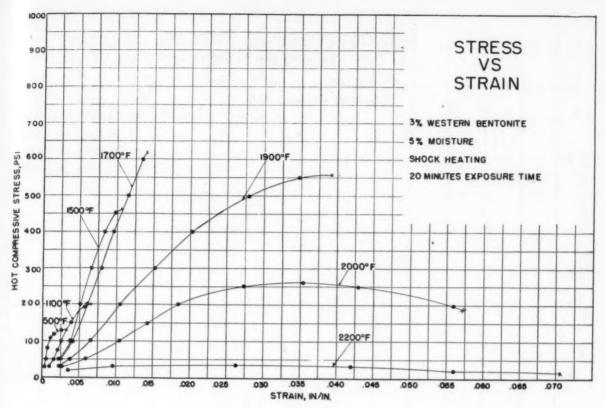


Fig. 3

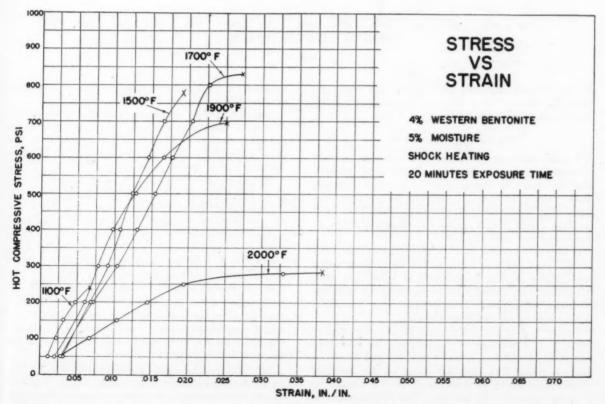


Fig. 4

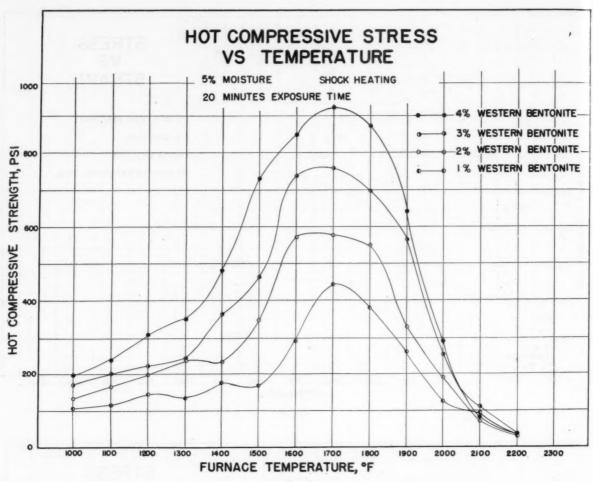


Fig. 5

sey No. 60 base sand and utilizing either fire clay or western bentonite as a binder, with moisture held constant at either 3 per cent or 5 per cent.

#### Typical Stress-Strain Curves

Typical stress-strain curves for bentonite-bonded mixtures are given in Figs. 1, 2, 3, and 4. It is noted that the shape of the stress-strain curve is approximately the same for any selected temperature, regardless of the binder concentration. The axis units for Figs. 1 and 3 were specifically chosen in order to emphasize this characteristic.

It should be noted that specimen failures are brittle and take place in shear in the range from room temperature to that temperature at which the peak hot compressive strength is developed by the mixture. At higher temperatures, the specimen fails plastically due to flow of the bonding material.

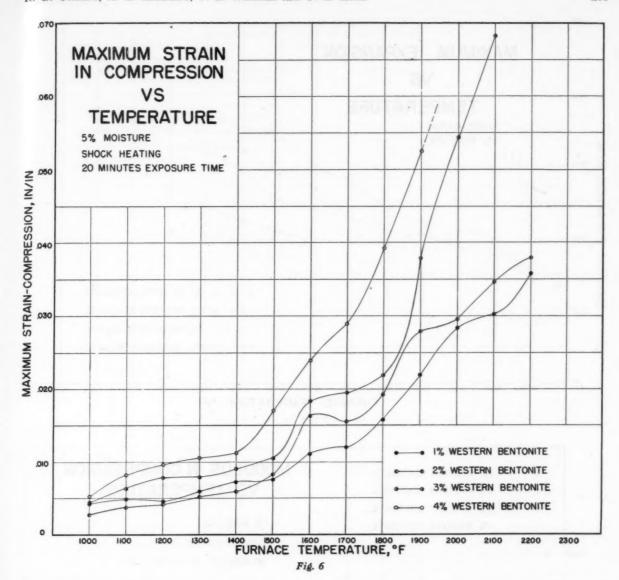
Figure 5 is a composite plot of hot compressive strength as a function of temperature for bentonite concentrations of 1, 2, 3 and 4 per cent and bears out the recognized fact that the shape of the strength-temperature curve is determined by the binder or binders present in the mixture. The displacement of the individual curves in Fig. 5 may indicate that the

thickness of the bond film has considerable influence on the strength of the mixture.

Maximum strain (ductility is shown to be closely related to bond film thickness in Fig. 6, and quite obviously due to the fact that the bentonite is the most easily deformable component in the mixture over the entire temperature range.

Maximum free expansion on shock heating is given in Fig. 7 as a function of furnace temperature for the bentonite-bonded mixtures. Since the clay is the only material in the sand mixture which exhibits a marked contraction over the entire temperature range from room temperature to the usual steel pouring temperatures, it is not surprising that the expansion of the mixture should vary inversely with clay content. It should also be noted that the expansion-temperature curve is always positive (shows no contraction) because any large unit contraction of the minor constituent (clay) is masked by the low unit expansion of the sand grains since they exist in the mixture as the major constituent.

Modulus as a function of temperature for the bentonite-bonded mixtures is plotted in Fig. 8. For each of the stress-strain curves there appears to be a portion where stress is proportional to strain. The slope



of the curve in this straight-line portion is the value recorded as the modulus. These slopes were taken from the autographic records.

The values from a series of tests made on one mixture at any temperature varied considerably and the plotted results represent the best average which could be obtained. No attempt will be made at this time to interpret these data since it is not certain if the accuracy of the determination is adequate. Further work is necessary also to determine if the mixture is elastic over this range of proportionality.

#### References

1. R. G. Thorpe, P. E. Kyle and J. P. Fraser, "Tenth Annual Report on the Investigation of Steel Sands at Elevated Temperatures" Transactions, A.F.S. vol. 58, pp. 133-143, (1950)

tures," Transactions, A.F.S., vol. 58, pp. 133-143 (1950).

2. R. E. Morey and C. G. Ackerlind, "A Study of the Effect of Various Binders and Additives on the Hot Strength of Molding Sands," Transactions, A.F.S., vol. 58, pp. 411-422 (1950).

3. J. P. Fraser and P. E. Kyle, "Ninth Annual Report on the Investigation of Steel Sands at Elevated Temperatures," Transactions, A.F.S., vol. 57, pp. 89-99 (1949). 4. D. C. Williams and P. E. Kyle, "Seventh Annual Report on Investigation of Properties of Steel Sands at Elevated Temperatures," Transactions, A.F.S., vol. 55, p. 607 (1947).

#### PART II

## Strain-Expansion Characteristics of the Mixture and Their Relationship to Scab Formation

On the basis of a limited amount of laboratory test and foundry casting data, a tentative correlation between the fundamental strain-expansion characteristics of several sand mixtures and the tendency of those mixtures to form defects of the mold wall fracture type was proposed in the Tenth Annual Report.¹ The two sand mixtures tested in this phase of the work both used the New Jersey No. 60 base sand, one bonded with 4 per cent Western bentonite, the other bonded with 10 per cent fire clay; both maintained at a moisture level of 5 per cent. It was determined that the fire clay-bonded mixture showed a more pronounced tendency to form the scab defect than did the bentonite-bonded mixture.

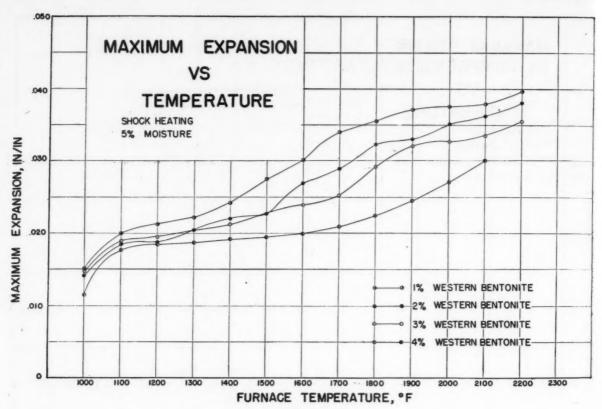


Fig. 7

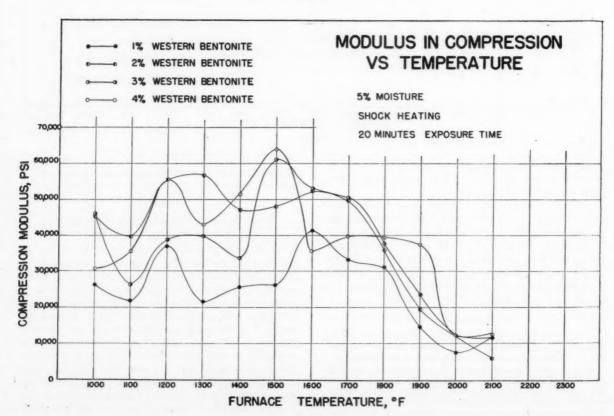


Fig. 8

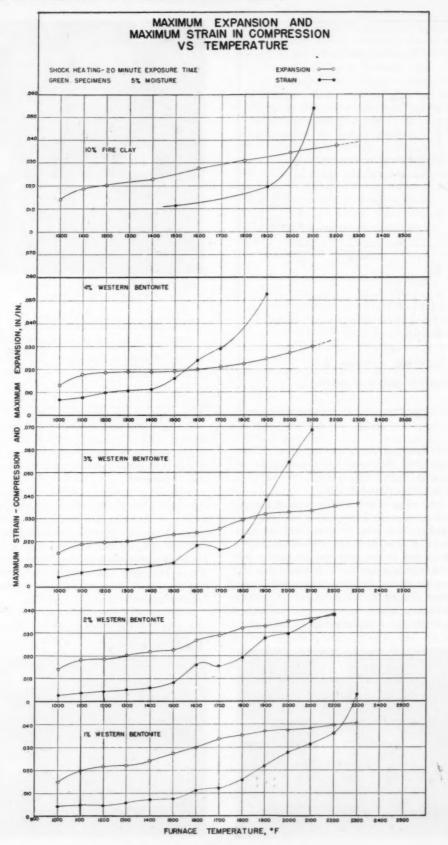


Fig. 9

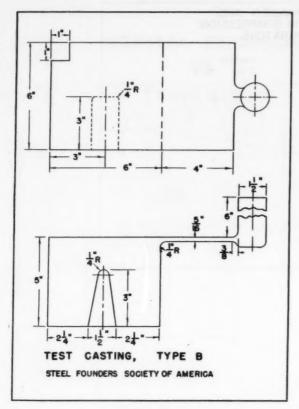


Fig. 10

As suggested mixtures for a continuation of this work, the Committee indicated that a reduction in clay concentration might also produce a scabbing mixture in a bentonite-bonded sand.

Expansion and stress-strain studies were made on sand mixtures bonded with 3 per cent, 2 per cent and 1 per cent western bentonite, misture held constant at 5 per cent. The new data are plotted on Fig. 9 with the data for the 10 per cent fire clay and 4 per cent Western bentonite mixtures. It is noted that the temperature at which the free expansion of the mixture is less than the ability of the sand to deform under load without failure moves toward a higher value as the bentonite concentration is reduced.



Fig. 11—Test castings showing scabbing. See Table 2.



Fig. 12-Test castings showing scabbing. See Table 2.

TABLE 1

Binder		Concentration, %	Critical Temperature, F
Western	Bentonite	4	1550
**	9.9	3	1870
9.9	99	2	2150
913	**	1	2240
Fire Cla	y	10	2040

It might be predicted, therefore, that the mixtures bonded with 2 per cent and 1 per cent bentonite would form scab defects even more readily than the 10 per cent fire clay mixture, if tested under comparable conditions.

In order to confirm the laboratory data, test castings were poured in steel, using the Steel Founders' Society of America test casting (Fig. 10). The first series of castings poured with moisture held at 5 per cent for all the mixtures showed defect formation for all the mixtures, ranging from slight scabbing in the 4 per cent bentonite-bonded mixture to very badly scabbed areas with the 1 per cent bentonite-bonded mixture (Figs. 11 and 12).

Moisture was then reduced to 3 per cent for all mixtures and the series repeated. Figure 13 shows a

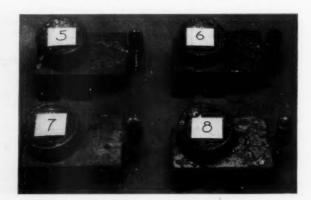


Fig. 13—Test castings showing scabbing. See Table 2.

decrease in scab severity as the bentonite content is increased.

A description of the sand mixtures used in producing the castings shown in Figs. 11, 12 and 13 is given in Table 2.

TABLE 2-TEST CASTINGS

Casting	Bind			
No.	Туре	Content, %	Moisture, %	
1	Western Bentonite	1	5.0	
2	**	2	5.0	
3	22	4	5.0	
4	Fire Clay	10	5.0	
5	Western Bentonite	1	3.0	
6	P9	2	3.0	
7	**	4	3.0	
8	Fire Clay	10	2.9	
9	Western Bentonite	1	4.2	
10	**	2	4.5	
11	**	4	4.5	
12	Fire Clay	10	4.4	

#### APPENDIX A

### Description of Hot Strength-Deformation Recorder

The Hot Strength-Deformation Recorder automatically records the values of stress and strain produced by loading a sand sample from zero load up to its breaking point. The sensing elements for recording load and deformation are differential transformers. The transformer for the deformation recording, T_D, is shown on Fig. 14, and the transformer for the load recording is in the recorder box. These transformers have one primary winding and two secondary windings on a common axis around an open center spool. The iron cores of the transformers inside the spools are free to move axially with respect to the windings and can be adjusted by thumbscrews, S.

When the iron core is centered in the transformer, equal voltages are generated in the secondary windings. Axial displacement from this center position produces a measurably larger voltage generated across the closer secondary winding, and a correspondingly smaller voltage across the secondary winding further

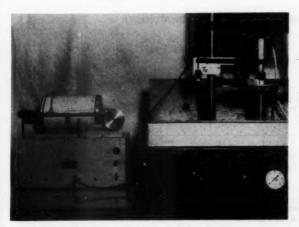


Fig. 14—High temperature load-deformation recorder.

from the core. This change in voltage is amplified by a power pack which, in turn, drives the recording motors in the direction indicated by the secondary winding of the highest voltage.

The change in secondary voltage varies linearly with core position provided that the core is within 0.100 in. of the center position. Therefore, the core has a range of movement of 0.200 in. or 0.100 in. on either side of the center position. Outside of this range, the output voltage will not be linear and hence the recorded values of load and deformation will not be accurate. It is necessary, therefore, to keep the differential transformers in proper adjustment.

The deformation pick-up assembly consists of a lever, L, with a variable ratio, and the differential transformer mechanism discussed earlier. With a lever ratio of 1:1, the deformation range of the recorder is 0 to 0.080 in./in. Changing the ratio to 1:2 gives a range of 0 to 0.160 in./in., while a ratio of 1:3 extends the range from 0 to 0.240 in./in. With a lever ratio of 1:1, the graph reads directly in thousandths in./in., using the scale provided. In the range from 0 to 0.160 in./in., the deformation scale readings must be multiplied by 2 and in the 0 to 0.240 in./in. range, the readings must be multiplied by 3.

The deformation pick-up assembly is clamped to the left hand post of the dilatometer. With the tip of lever, L, resting on the metal collar which holds the lower ceramic loading post, the assembly is adjusted until the lever is level, as judged by eye. Using the 0 to 0.080 in./in. range, the transformer core is centered by placing the 1.920-in. metal specimen between the ceramic caps on the lower loading post. The dilatometer is now loaded until the upper post is contacted. The transformer core should now be at the center of the differential transformer, but it can be positioned exactly in the center of the transformer by using a low range AC voltmeter to record the exact voltages across each of the secondary windings.

The load recording mechanism is a Bourdon Tube with a differential transformer mounted on its tip. The Bourdon Tube is connected to the high pressure side of the hydraulic system of the dilatometer. The position of the transformer core in relation to the windings is set correctly at the factory, but can also be checked in the same way as the deformation transformer. Again movement of the core inside the transformer gives different voltages which are amplified in the power pack. This signal is sent to the motor which, in turn, drives the drum.

#### Procedure

In actual use, a sheet of recording paper is placed on the drum and held in place with two coil springs at either edge of the paper. The recorder is now set at zero psi by loosening the thumbscrew,  $S_L$ , which holds the drum in place, and rotating the drum until the stylus is at the zero position on the chart. The thumbscrew should be tightened after positioning. With pencil lead in the stylus and the stylus on the paper at zero psi, the transformer thumbscrew is adjusted so that the stylus moves to the left hand margin of the paper.

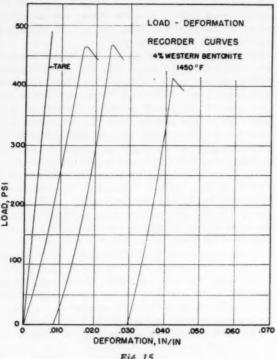


Fig. 15

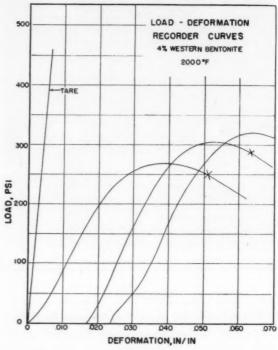


Fig. 16

are up, the recorder is turned on and allowed to A 11/8-in. by 2 in. long green sand specimen is prepared in the manner prescribed by the A.F.S. test warm up. With the stylus set at zero strain and zero procedure.* This specimen is placed in the dilatometer and allowed to soak at temperature for a

*See FOUNDRY SAND HANDBOOK, 6th edition, p. 107, published by the American Foundrymen's Society, Chicago (1952).

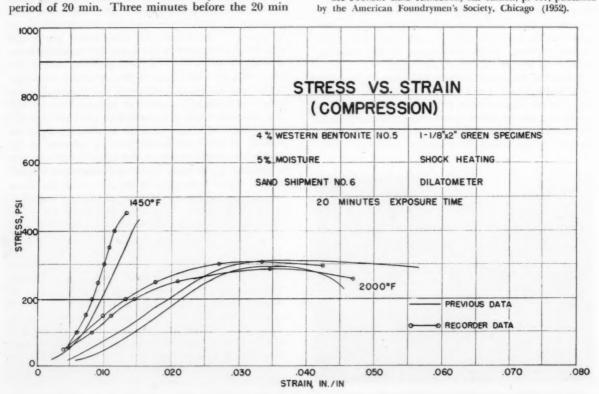


Fig. 17

stress, the specimen is loaded to failure. Curves such as Figs. 15 and 16 are obtained.

A tare curve is run by placing a cold ceramic cylinder equal in length to the sand specimen plus two caps on the lower loading post. Loading is started as soon as the cylinder is raised to contact the upper loading post, and continued to the maximum expected value of load. Such a curve gives the deformation of the loading system. At each value of stress, the corresponding value of loading system deformation is picked from the tare curve and subtracted from the experimental curve to give the true value of strain for the particular specimen in question. Curves such as those shown in Fig. 17 are the stressstrain curves plotted using data from the load-deformation curves made with the recorder. At least five specimens are tested and the average values from the curves are used as characteristics of the sand mix-

#### Comments on Operation

There is an inherent time lag in the strain recording mechanism as can be seen from the curves of stress vs strain for bentonite, Fig. 17. At 2000 F, there are two separate curves of stress vs strain from both the previous data¹ and the recorder data. These show the variation which is obtained at a single temperature. These curves indicate that for a given stress, the corresponding strain as compared with previous data taken with a movie camera lags by as much as 0.005 in./in. No doubt, a time lag also exists in the stress recording mechanism. However, this lag cannot be detected by eye. Values of stress observed on the dilatometer pressure gage correspond very well with values observed on the recorder during loading.

The time lag in the strain mechanism is due to the short times necessary to pick up, transmit, and retransmit the signals from the cores to the motor driving the stylus and the drum. There is a further time lag in overcoming the inertia of starting these elements. Both of these lags would have the effect of moving the curves, Fig. 17, to the left.

Copies of the typical curves as obtained from the recorder are included, Figs. 15 and 16. At 1450 F, it can be seen that a sharp peak is obtained due to the rapid, brittle failure of the specimen. From this peak, values of maximum load and deformation can readily be obtained. However, at 2000 F some difficulty is encountered in obtaining the value of maximum strain. The strain at maximum stress is the highest point on the curve, but the strain at failure (maximum strain) occurs somewhere to the right of the peak due to the gradual plastic failure of the specimen. Failure does not occur at the point of highest stress, although for all practical purposes, the sand may no longer have significant load carrying ability.

On most of the curves at these high temperatures, the load decreases after the maximum and at the point of failure the curve becomes a straight line, due to the recoil of the drum. The beginning of this straight line portion of the curve (marked with an "x" on Fig. 16) after maximum stress is reached has been taken as the point of failure and the point of

maximum strain. On many curves this point of failure is difficult to obtain because the straight line portion of the curve is approximately tangent to the load deformation curve before failure.

The new data on hot compressive strength vs temperature for both fire clay (Fig. 18) and bentonite (Fig. 19), are within the experimental limits of accuracy when compared with previous data. However, the value of strain appear to be substantially lower than those obtained previously and are apparently not within the limits of experimental accuracy. These differences are attributed to the lag in the strain recording mechanism already mentioned. When these low values of strain were obtained, the expansion characteristics of the sand mixtures were checked using the research furnace and found to agree with values for similar mixtures previously tested, indicating that the sand mixtures had not changed.

#### APPENDIX B

#### Research Furnace Control System

The original research furnace control system was designed as a high-voltage capacitor-modulated unit.⁴ After the capacitor bank burned out in July 1950, it was decided to improve the control system and to eliminate the operating hazards which had been inherent in the original set-up.

A General Electric Reactrol unit was purchased, installed and tested satisfactorily. Essentially, the new system is a low-voltage, high-current proportional type controller which varies the power applied to the heating units in infinitely variable increments by means of a saturable core reactor. Several safety devices are incorporated into the system to prevent furnace failure due to overheating, with the result that the furnace may now be left unattended for long periods of time while operating at 2500 F.

General apparatus arrangement is shown in Fig. 20. Note that the entire unit is extremely compact.

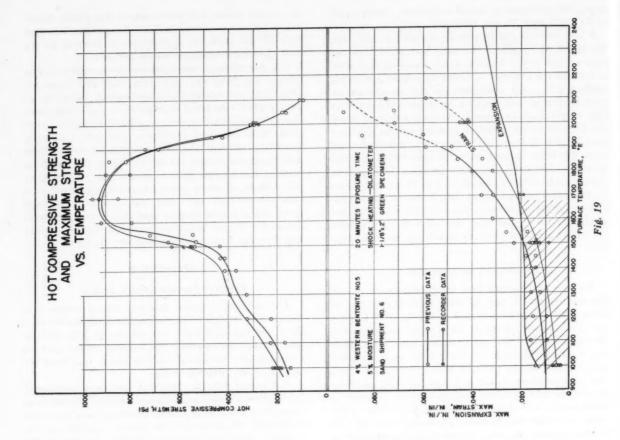
A considerable amount of time has been devoted to the installation and testing of the new control system to insure that this unit was capable of meeting the design demands. It is felt that the apparatus is completely satisfactory in all respects.

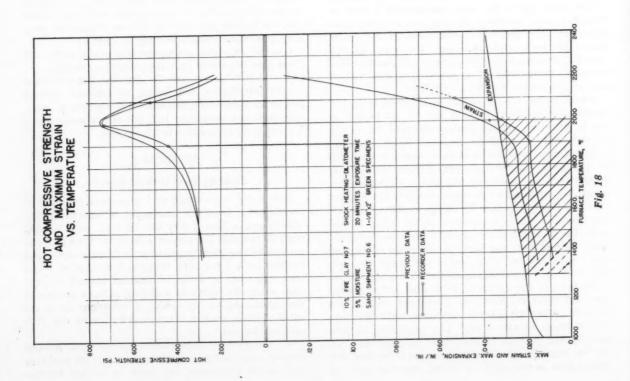
#### APPENDIX C

### Scabbing Tests at Wehr Steel Co.

Exhaustive stress-strain and expansion testing had revealed that from the standpoint of theoretical considerations, a New Jersey No. 60 base sand bonded with 10 per cent fire clay at a moisture level of 5 per cent should be more prone to scab than the same sand bonded with 4 per cent Western bentonite at 5 per cent moisture. A series of castings poured in gray iron using the Fairfield scabbing pattern confirmed the laboratory testing work. A complete discussion of this phase of the research is given in the 10th Annual Report.¹

At the time of presentation of the 16th Annual Report, steel melting facilities were not yet available at Cornell University. In order to further confirm





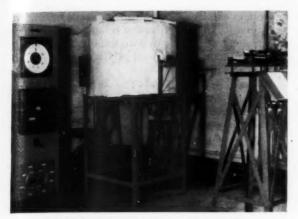


Fig. 20-Research furnace and controls.

and extend the work, a similar study was undertaken at Wehr Steel Co. foundry, Milwaukee, August, 1950.

It was found that the Fairfield scabbing pattern could be used for work in steel if modified by (1) removal of two opposed pattern sides so that liquid metal is forced to flow over a large green sand core, and (2) the addition of a large riser to the top of the pattern in order to provide an adequate feed reservoir. (See Fig. 21.)

The plan of the investigation was essentially a duplicate of the founding work at Cornell University except that three base sands were used, each bonded with 4 per cent Western bentonite and 10 per cent fire clay, with moisture held at 5 per cent. Screen analysis data for the three base sands is given in Table 3.

TABLE 3—SCREEN ANALYSIS FOR WEHR BASE SANDS

US Screen Mesh	N.J. 60-2	N.J. 100-1	Portage	
12	_	_	_	
20	_	_	_	
30	0.2	_	-	
40	1.4	0.2	2.0	
50	11.2	0.6	20.3	
70	47.0	13.0	35.5	
100	29.6	41.2	28.5	
140	9.4	31.6	11.0	
200	1.6	10.0	1.6	
270	0.2	2.6	1.1	
Pan	0.2	1.0	_	
I.F.S. GFN	61.3	89.4	62.4	

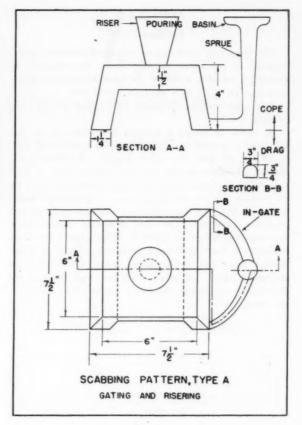


Fig. 21

All molds were rammed on a production jolt-squeeze machine to a mold hardness of 67-75 with the sand mixture under test used as facing sand over the top and sides of the pattern to a rammed depth of at least 1½ in. All molds were poured in acid-electric steel. A representative portion of the founding data is given in Table 4.

It is noted that fire clay bonded sands showed a much higher incidence of scab formation than did Western bentonite bonded sand. The above data, while interesting, is not regarded as being highly conclusive due to the fact that stress-strain and expansion data are not available for the mixtures using the Portage and N.J. 100-1 sands. Such data will not be determined until the load-deformation recorder becomes available.

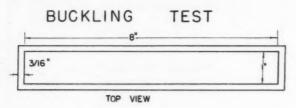
TABLE 4

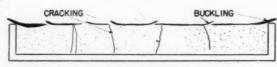
No.	2	3	0	8	x	2a	y	10
Base Sand	NJ 60-2	NJ 60-2	NJ 100-1	NI 100-1	Portage	Portage	Portage	Portage
Moisture, %	5.1	4.8	5.4	5.4	5.0	4.8	4.2	4.2
Clay Type	W.B.	F.C.	W.B.	F.C.	W.B.	W.B.	F.C.	F.C.
Clay Conc., %	4	10	4	10	4	4	10	10
Permeability	140	135	80	70	95	95	100	95
Green Comp. Str.	3.1	1.8	3.3	2.1	4.3	4.3	2.5	2.4
Mold Hardness, Cope	50-65	65	65	60-70	60-65	65-70	60-70	55-65
Mold Hardness, Drag	70-75	65	70-75	75	65	70-75	70	65-70
Scab Occurrence	No	Yes	No	Yes	Yes	No	Yes	Yes
Defect Area = 2 in.2	No	Yes	No	Yes	No	No	Yes	Yes

# APPENDIX D Buckling Test

The buckling tendency of any sand mixture, when exposed to radiant heat, may easily be observed by making a suitable mold of the sand mixture in question and pouring only enough metal into the cavity to heat the sand at the mold cavity face in the cope section.

In order to create a comparable situation in the laboratory, a sand specimen must be completely confined on all but one side. Figure 22 illustrates a special specimen holder for the 1-in. x 1-in. x 8-in. expansion specimen made of fused quartz and the type of surface failure observed with a green specimen made of N.J. No. 60 sand and bonded with 10 per cent fire clay at 5 per cent moisture under shock heating conditions at 2500 F. This shows the very marked effect of rapid heating on the expansion of the surface layer and the buckling due to its inability to deform without fracture. This test will be used in the future to observe buckling tendencies in various mixtures.





SECTION THROUGH FUSED QUARTZ CONTAINER
Fig. 22

# APPENDIX E Notes on Expansion Testing

A-Specimen Geometry-An investigation of the effect of specimen geometry on the magnitude of the

maximum expansion developed on shock heating reveals that maximum expansion is independent of specimen size and shape. An 8-in, gage length is ideally suited to the apparatus used at Cornell and will be used in the future.³

B—Specimen Preparation—The specimen press formerly used for the preparation of expansion specimens has been discarded in favor of a modified transverse test core box and rammer. Comparison tests run on specimens prepared by pressing and ramming to the same density show that mode of specimen preparation does not influence the specimen behavior on shock heating.

C—Support Refractories—In the Cornell expansion test procedure, the slabs cut from a zirconite brick have been used to support the specimen during the test. While zirconite is highly resistant to heat shock, frequent breakage of the support slabs has promoted a search for a better material.

Specimens of fused mullite, zirconium oxide, zirconite and vitreous silica have been tested. Vitreous silica is by far best material found to date. A series of expansion tests were run on a representative mixture to insure that the support slab material or geometry had no effect upon the expansion of the specimen. Since results obtained with the vitreous silica slabs check with the results obtained with the old zirconite slabs, the vitreous silica slabs will be used for all future expansion testing.

#### DISCUSSION

Chairman: H. W. DIETERT, Harry W. Dietert Co., Detroit. Co-Chairman: E. C. Zirzow, Werner G. Smith, Inc., Cleveland. Recorder: R. H. JACOBY, The Key Co., East St. Louis, Ill.

J. B. Caine: ¹ There are too many test casting designs in use. The time has come to standardize on one pattern in place of the 27 different cnes used. A committee is now being formed to set up a uniform test casting. All foundries considering the use of test patterns should contact the American Foundrymen's Society for specific design.

W. B. PARKES: ^a We tried to determine at what temperature sand scabs occur. A riser on casting was used with thermccouple hook-up. Results were widely scattered from 573-1100 C (1065-2010 F).

¹ Consultant, Cincinnati, Ohio.

² British Cast Iron Research Association, England.

## RISERING OF GRAY IRON CASTINGS

Progress Report No. 3

By

W. A. Schmidt* and H. F. Taylor**

#### Authors' Foreword

On January 5, 1950 the Gray Iron Research Committee of the American Foundrymen's Society met at M.I.T. and outlined the following program:

(1) A correlated bibliography would be completed, and

(2) A fundamental program of research would be undertaken to rationalize the complex problem of risering gray iron castings.

"Fundamental research" is construed to mean an approach which will lead to a basic understanding of why gray iron castings behave in anomalous and undependable fashion as regards shrinkage characteristics. Some of the many sponsors of this current work (members of A.F.S.) may think an indirect approach, using conventional risers and making castings of various shapes and sizes, would be a more fruitful avenue of attack. Since no real progress has been made to date by this approach, the more fundamental method seems indicated.

The following is a "boiled-down" description of work done to date at M.I.T.

#### I. Introduction

Progress Report No. 1 was an extensive bibliography on practices and theories on gating and risering of gray iron. Report No. 2 was a very short synopsis of studies made on types of shrinkage encountered in commercial gray iron castings; these were of typical grades of commercial gray iron furnished by the Gating and Risering Committee.

Last year (1951) an oral report was made on research at M.I.T. in which spheres of different grades of cast iron were rotated during solidification. It was found the type of iron affected the degree of shrinkage to a very small extent; the biggest factor governing shrinkage behavior being the mold into which the metal was poured. Dry sand molds either eliminate or greatly reduce the amount of shrinkage in gray iron castings, while green sand molds are very prone to

produce large cavities (Fig. 1). This led to some work on the effect of various gases on the shrinkage behavior of iron. The results of this work showed that hydrogen, oxygen, and water vapor all produced small effects tending to increase the amount of shrinkage present, but not of a magnitude large enough to be a controlling factor in the problem (Fig. 2). Rotation of spheres did not prove necessary so the apparatus is not described; spheres cast statically gave the same results.

While this work was being done, it was observed that two distinctly different types of shrinkage cavities were recurring. The most prominent type, due to its frequent occurrence, was the conventional "V" shaped cavity (Fig. 2-4). The other and more obscure type of cavity was one with smooth parallel walls, having a diameter 1/4 to 1/2 in., penetrating the casting to a depth of about 1 in. (Figs. 2-3 and 2-6). The shape and character of this second type cavity suggests it is formed by a sudden and urgent demand for feed metal while the casting is in a mushy or semi-solid state. The first and more common type cavity is formed by progressive feeding of the casting.

#### II. Description of Current Work

It has long been suspected that mold wall movement has been responsible for shrinkage in gray iron castings; it is known that castings showing unexpected cavities are often larger than those that show no shrinkage under similar conditions of risering. Besides the work of this research, it has lately been indicated in the literature that mold wall behavior is an important item; so the current research (described herein) includes:

1. measurement of dilation of mold and casting during solidification. (these measurements to be made in both green and dry sand);

2. measurement of shrinkage of iron in green sand molds, its magnitude and relation to temperature of casting;

 dimensional measurements of solidified castings;
 an approach that would determine whether or not dilation in green sand molds could be restricted by mechanical methods;

^{*} Research Assistant and ** Associate Professor Massachusetts Institute of Technology, Cambridge, Mass.

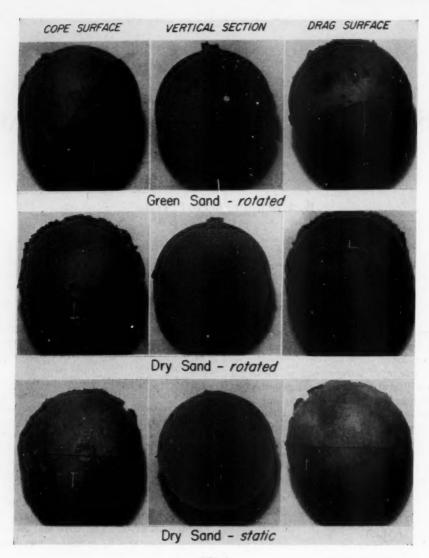


Fig. 1

5. density determinations; and6. qualitative study of the effects of thermal gradients in solidifying castings.

#### III. Procedure

In all experimental work reported hereafter, sand and metal composition were kept constant. The sand used was washed silica (No. 80 A.F.S. grain fineness), bonded with 4 per cent Western bentonite, ½ per cent cereal, ½ per cent dextrine, and 3½ per cent water. The metal was iron from a master heat, remelted by induction in a graphite crucible. The analysis of the metal was: C—3.02; Si—1.42; Mn—0.96; S—0.09; P—0.09.

Method for Studying Effect of Mold Restraint—As a preliminary step, the effect of mold restraint on dilation was studied. This was entirely qualitative due to difficulties in maintaining constant mold conditions. A thin steel sleeve was faced inside with thicknesses of sand ranging from ½ in. down to a mold

wash, in 1/g-in. steps. These castings were small cylinders 11/2 in, in diameter and 3 in. high. The pipe was varied in diameter to accommodate added thicknesses of sand. The manner in which they were adjudged for degree of dilation was the relative size of the shrink hole formed and the measured diameter.

Method for studying Mold Movement (Dilation)—To study dilation of castings as they solidified, a small 2-in. diam cylinder, 5 in. high (bottom-gated), was used. An Ames Dial was placed 1 in. from the bottom of the casting and another 1¾ in. from the top (Fig. 3). A connecting link of carbon rod transmitted movement of the mold-metal interface to the Ames Dial. To provide a bearing for the carbon rod at the interface, a thin steel plate about ¼-in. square was molded in the sand. Temperatures were taken on a recording potentiometer with a Pt-Pt, 10% Rh thermocouple placed at the center of the casting. With an electric clock to indicate time, all instru-

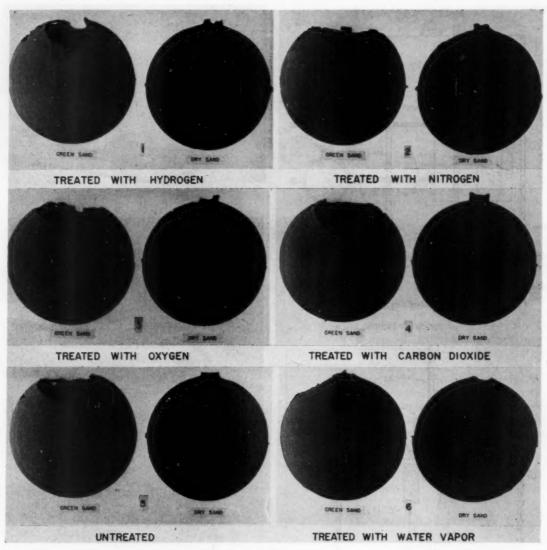


Fig. 2

ments and dials were photographed with a motion picture camera to permit a simultaneous and permanent record (Figs. 3 and 5). Castings poured in both green and dry sand molds were measured with this apparatus.

Method for Studying Rate of Metal Shrinkage—To measure the volume of shrinkage occurring with green sand molds, a larger casting, 3 in. in diameter and 6 in. high, was used. This was necessary to permit easier and more accurate measurements. Temperatures were recorded on a potentiometer using a Pt-Pt, 10% Rh thermocouple at the center of the casting. An open riser 1½ in. in diameter, 2 in. high, was placed atop the casting for feeding and to facilitate measurement of the shrinkage. The riser was formed by a thin shell of sand, ½6 in. thick, backed with 1 in. of vermiculite, for insulation. The casting was poured through the riser, and the riser was covered with rice hulls immediately after pouring.

To measure the level of the liquid metal in the riser as the casting solidified, a thin metal probe was lowered into the cavity through the rice hulls. When the probe touched the metal surface, it completed an electrical circuit, indicating contact (Figs. 4 and 6). In this way the rate at which riser metal was drawn into the casting could be followed accurately. Upon sectioning the riser, the volume and rate at which the riser fed could be calculated. Figure 9 shows the correlation of shrinkage volume with temperature of the casting at three different pouring temperatures.

Qualitative Evaluation of Thermal Gradients—To alter cooling rates in these castings, a set of three dry sand and three green sand molds were made of the 3 x 6-in. cylinder. The top was left open on all molds. Immediately after filling the molds, one pair was covered with an eighth-inch layer of clean dry sand, one pair with rice hulls and the last two castings were left open to the atmosphere.

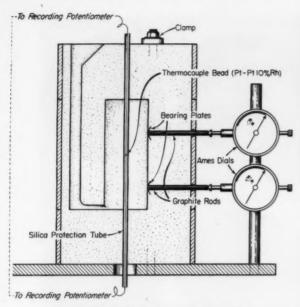


Fig. 3-Dilation measurement apparatus.

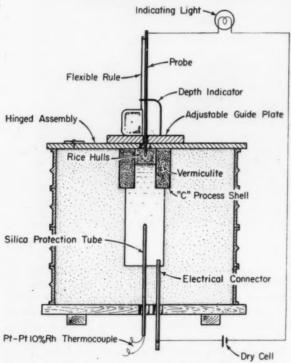


Fig. 4—Assembly for measuring rate of shrinkage.

All castings made in these experiments were carefully measured with micrometers on the diameter over the entire length at ½ in. intervals and then the casting was sectioned. Several measurements were made on each plane to obtain an average value.

Density determinations by the water displacement method were made on 1-in. cubes taken from the centers of several of the castings.

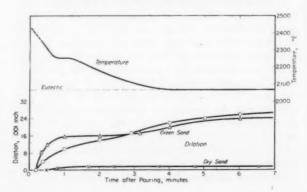
Fig. 7 (Right)—Dilation of green and dry sand castings.

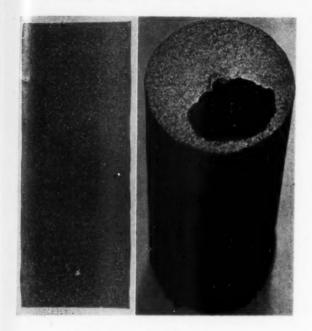


Fig. 5-Dilation apparatus.



Fig. 6-Apparatus for measuring shrinkage.





#### IV. Discussion of Results

The amount of dilation was found to be reduced by mechanically restraining the green sand mold wall. The principal observation made in this connection was that the steel sleeve had to be placed very close behind the facing sand before its effects were observed. This distance, between the casting and sleeve, was in the range of 1/4 to 1/8 in. When the casting was poured directly against the sleeve, coated only with mold

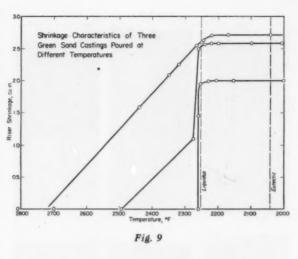


Fig. 8—Typical dry sand and green sand castings made in dilatometer experiments. Pouring temperature, 2525 F.

wash, dilation increased somewhat. This could be attributed to the fact that the sleeve heated up more than before and expanded during the critical solidification period of the casting.

The data obtained from the Ames Dials, were used to plot curves of dilation versus time and temperature (Fig. 7). The exact amount of dilation observed was reproducible only in dry sand molds but the results were consistent even for green sand. These curves

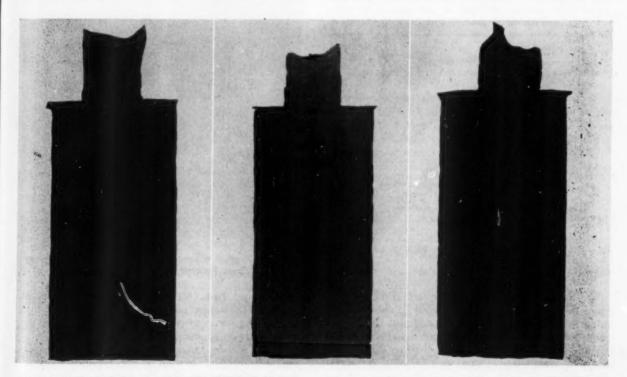


Fig. 10—Castings made to determine shrinkage characteristics using insulated risers in green sand molds. Pour-

ing temperatures (left to right), 2300 F, 2535 F and 2730 F.

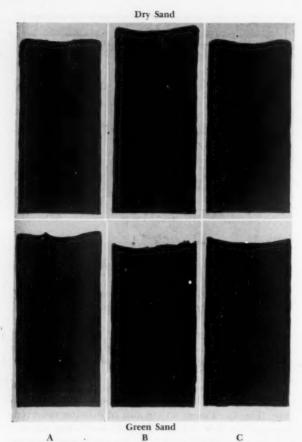
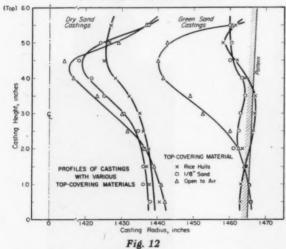
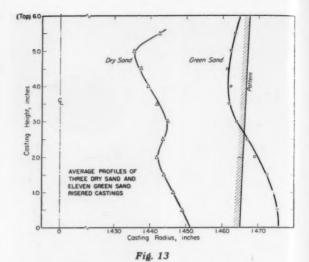


Fig. 11—Effect of top-covering materials on dry sand and green sand castings. A, Open to atmosphere; B, Thin layer of dry sand; C, Rice hulls.



show that dry sand molds dilate very little. Green sand molds show a consistent rapid dilation, beginning as soon as the mold is poured and continuing through a point slightly above the eutectic. On further cooling, dilation proceeds at a much slower rate, stopping completely at some point during the eutectic hold.

Profiles of all the castings showed a slight dishing



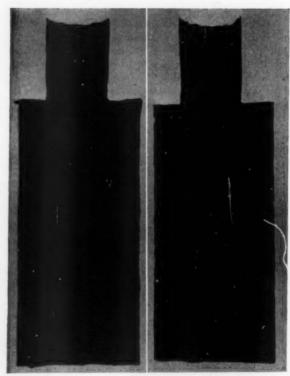


Fig. 14—Typical risered dry sand and green sand castings. Pouring temperature, 2470 F.

in the upper half. All attempts to pick this up on the Ames Dials were unsuccessful. It is believed that this dishing occurs after the major portion of the observed dilation is passed.

Figure 9 shows that feeding in green sand castings is completed as soon as the liquidus temperature is reached, and that the amount of feeding is not dependent on pouring temperature alone. In

other words, any dilation occurring after the liquidus is reached will cause shrinkage of some sort, since no more riser metal is available.

Comparison of Figs. 7 and 9 shows that these two experiments are in relative agreement regarding dila-

tion as a cause of shrinkage.

Measurements of the diameters of all the castings show that those made in dry sand are always smaller than those made in green sand. Figure 13 shows the average shapes of risered 3-in. cylinders. It will be noticed that the dry sand castings adhere more closely to patternmakers' shrinkage rules; and that the distortion present at the bottom of the castings is more pronounced in the case of green sand.

Density determinations made on many of the castings show no trend towards greater density in green sand castings. This is in contradiction to results we have reported previously on another group of experimental castings. It appears density is not an im-

portant parameter.

In all castings of green or dry sand, up to this point, there were no signs of interdendritic shrinkage. The open castings, using various covering treatments to obtain different cooling gradients, were used to induce some interdendritic type shrinkage; the technique proved wholly successful (Fig. 11). castings, except the one in green sand with the top exposed to the atmosphere, exhibited some form of interdendritic shrinkage. The dry sand castings covered with rice hulls or sand, and the green sand casting covered with rice hulls had shrinkage connected with the atmosphere through the surface. The green sand casting covered with sand had a shrinkage area close to the top surface. The dry sand casting exposed to the surroundings had shrinkage located near the center. The dimensions of these castings were measured and their profiles are shown on Fig. 12.

#### V. Summary of Results

The results of this work have shown that:

1. Dilation could be restricted by mechanically restraining the movement of the green sand mold wall.

2. Dilation of the casting occurs rather rapidly, most of this movement occurring before the eutectic is reached, but continuing to a lesser degree through the eutectic hold.

3. The volume of shrinkage occurring in a green sand casting is not necessarily a function of pouring

4. Risers of the class of metal used in this work stopped feeding at a temperature slightly below that

of the liquidus.

5. Dimensionally, green and dry sand castings made from the same pattern fall into two separate groupings-all dry sand castings being smaller than the green sand, and following patternmakers' shrinkage rules more closely.

6. The density of green and dry sand castings of the size used in this work does not differ in a man-

ner which will form a pattern.

7. Under certain unique combinations of mold material, metal composition, thermal gradients, etc., a casting will develop an interdendritic type shrinkage in one instance and be sound in another.

#### VI. Conclusions

We have shown that for the type iron used in this work, a considerable amount of dilation occurs before any substantial amount of metal has solidified; that dilation continues through the eutectic hold; that feeding of the riser stops while about 70 per cent of the metal is still liquid; and that cooling rates and/or gradients are influential in determining the type shrinkage, and the degree of soundness in the casting.

From this, it is possible to hypothesize:

1. That eutectic expansion accounts for some of the dilation observed.

2. That the green sand mold is not a rigid container.

3. That gray iron does not form a solid strong skin.

4. That ordinary large risers form no useful purpose other than creating larger ferrostatic pressures (which can be harmful in that this can cause greater dilation in the casting) and feeding part of the dilation shrinkage. Large risers do influence temperature gradients, sometimes being more harmful in this respect than if no riser were used.

5. Proper cooling conditions must be maintained to insure sound castings. What these conditions are, or should be, will be determined in the individual instance by the size, shape, and composition of the casting. It seems that a uniform cooling rate over the entire casting is most desirable; attempting to obtain directional solidification, and consequently preferred feeding from risers, will not work because the risers will not feed below the liquidus temperature. Changes in section size will continue to create trouble until their effect on thermal gradients can be minimized.

In Summary

This report is concerned only with an accounting of work to date. Naturally, some philosophy on the feeding of gray iron castings has developed as a result of this work, but the foundation is still too untenable to present for public consumption. No attempt is made to do more than offer a concise interpretation of the results, sans elaborations.

#### DISCUSSION

Chairman: J. S. VANICK, International Nickel Co., New York. Co-Chairman: T. E. EAGAN, Cooper-Bessemer Corp., Grove City, Pa.

Recorder: J. H. SCHAUM, Bureau of Standards, Washington,

J. T. MACKENZIE: 1 Future work along this line by the authors should be done using oil sand cores. I noticed the casting walls "dished-in."

RECORDED SCHAUM: "Tucking-in" of casting walls might be due to development of a partial vacuum within the casting when solidification shrinkage takes place after the flow of feed metal has been shut off. The combination of atmospheric pressure acting on the outside of the solidified casting wall and a partial vacuum on the inside is often sufficiently great to move the relatively weak wall inward to relieve this pressure differential. T. W. Curry: ^a Did you run gas analyses in the experiments

in which the metal was gassed prior to pouring spheres.

Mr. TAYLOR: No gas analyses were run.

Technical Director, American Cast Iron Pipe Co., Birmingham, Ala.
 Dir. of Mfr. Research, Lynchburg Foundry Co., Lynchburg, Va.

## VEINING TENDENCIES OF CORES

#### A Progress Report by

#### A.F.S. Sand Division Committee 8-J*

#### Introduction

The material presented in this Progress Report deals with factors that affect the presence of metal fins, not at the core parting, but on cored surfaces of castings. Metal fins of this nature are caused by metal flowing into a break on the core surface. A defect of this nature is termed veining. Since this Committee's task is to study molding materials for iron castings at elevated temperatures, the study of veining is limited to iron castings. All of the test castings were made at the University of Michigan in Ann Arbor, and all the tests on cores were made at the laboratories of the Harry W. Dietert Co., Detroit. Meetings in the foundry were held on Dec. 6, 1950, Jan. 5, 1951, Jan. 31 and Feb 1 and 2, 1951, and Sept. 17, 18 and 19, 1951.

#### Properties of Cores Tested

The composition of the various types of core mixtures used in this study is tabulated in Table 1.

The properties of cores, such as baked and hot properties, together with casting condition, are tabulated in Table 2. The test data describes very closely the condition of cores used in the foundry for test castings, since both the laboratory and foundry cores were baked at the same time in the University of Michigan foundry core oven.

#### Test Procedure

#### Core Sand Mixing Procedure

- 1. Weigh all ingredients
- Mix sand in 18-in. laboratory muller-3 min dry, 4 min with water, and 4 min with oil.
- Clean mixer well between each type of core sand mixture.
- 4. Run dummy batch before mixing sand con-

- taining different types of binder from previous batch or when mixer has not been used for 30 min.
- 5. Place mixed sand in sealed container.

#### Core Making

- Make ten A.F.S. transverse cores of each test batch of sand. Ram with four drops of weight.
- Make following test cores:
   2-A.F.S. 2 x 2-in. specimens
   6-A.F.S. tensile specimens
   20-11/8 by 2-in. specimens
   3-Hot gas pressure specimens.

#### Core Baking

- Place cores in oven at 425 F, automatically controlled and frequently checked with a thermocouple placed close to the cores and connected to a reliable potentiometer.
- Bake long enough to develop optimum hardness.
- 10. Store cores in sealed containers.

#### Moldin

- Following molding procedure as specified for the Committee's Scabbing Research Project.
- 12. Set four test cores in each mold.
- 13. Make sufficient molds so that two cores are used from each type of core mixture. Preferred method: four different types of cores in one mold.

#### Meta

- University of Michigan scrap. Add sufficient phosphorus to obtain 0.50 to 0.60 per cent P in melt.
- 15. Melt in electric furnace.
- 16. Pour from hand ladle at 2650 F.

#### Shakeout

17. Shakeout castings 2 hr after pouring.

#### Cleaning

18. Sand blast castings.

#### Core Testing

19. Baked hardness Baked permeability

[•] Membership of the A.F.S. Committee on Physical Properties of Iron Foundry Molding Materials at Elevated Temperatures consists of the following: V. M. Rowell, Chairman, R. L. Doelman, W. A. Spindler, R. W. Bennett, K. S. Brooker, J. E. Coon, H. W. Dietert, H. H. Fairfield, J. A. Gitzen, R. A. Green, John Grennan, H. J. Jameson, J. D. Johnson, Roy Korpi, Henri Louette, R. W. Mason, A. E. Murton, E. J. Passman, D. E. Shiels, E. W. Smith, Michael Warchol, R. E. Wilke, J. T. Zaynor, E. C. Zirzow, and E. C. Zuppann.

Tensile strength
Transverse strength
Transverse deflection
Gas pressure
Core collapsibility, 5-psi load at 2000 F
Core expansion
Screen analysis of base sands.

#### Pattern

The test pattern selected for this investigation is a 5-in. cube containing four core prints in the drag to receive test cores each 1 in. square. The test cores projected into the metal a distance of 1 in. The test casting is shown in Fig. 1.

All laboratory tests were made in accordance with A.F.S. test procedures where available. The test procedure used for hot properties deserves explanation. The dilatometer in this test is rigged as for the hot compressive strength test. A 5-psi load is applied to the specimen, and as the specimen expands the movement of the lower table is read from the deformation dial indicator. This movement is divided by two to express growth of specimen in in./in. of length. This is often termed "unaccommodated expansion." The Committee suggests that this test be termed "expansion under 5-psi load." In this test the time that the test core would support the 5-psi load was also recorded.

#### Order of Tests

The tests were arranged so as to change one item at a time as far as possible. The test castings were made in the following order:

- Test A: Veining tendency vs. degree of ramming.
- Test B: Veining tendency vs. degree of baking.
- Test C: Veining tendency vs. metal pouring temperature.
- Test D: Veining tendency vs. grain size of sand.
- Test E: Veining tendency vs. moisture content of the core mix.

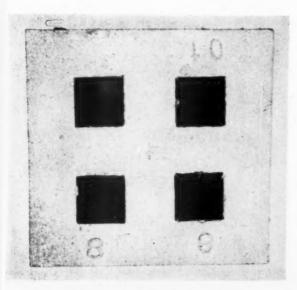


Fig. 1—Test casting used for veining study.

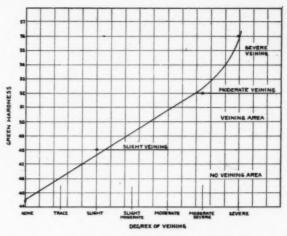


Fig. 2—Veining tendency increases with increasing green hardness.

- Test F: Veining tendency vs additives.

  (The preceding tests were all made with
- silica sands.)

  Test G: Veining tendency vs. change in base sand.
- Test H: Veining tendency vs. change in base sand
- Test I: Veining tendency vs. percentage of cereal binder.
- Test J: Veining tendency vs. additives.

in Test E.

- Test K: Veining tendency vs. bentonite additions.
- Test L: Veining tendency vs. varying blends of base sands.
- Test M: Veining tendency vs. type of cereal binder.
- Test N: Veining tendency vs. type of core oil.
- Test 0: Veining tendency vs. synthetic resins and other binders.
- Test P: Veining tendency vs. grain size of silica sand with less binder than in Test D.
- Test Q: Veining tendency vs. percentage of oil, with less cereal than in Test H.
- Test R: Veining tendency vs. moisture content of core mix, with different base sand than

Veining vs. Green Hardness: The hardness to which a core is rammed will influence the veining tendencies of a core. The relationship between green hardness of the core and degree of veining is shown in Figs. 2 and 3.

No veins were present on castings made with cores rammed to a 44.5 green hardness. As the green hardness was increased from 48 to 56, the degree of veining was increased (Fig. 2). A marked increase in the expansion under 5-psi load and a slight increase in collapsing time is secured with an increase in green hardness (Fig. 3). It is not recommended that cores be rammed to a low green hardness because this will cause weak cores and invite metal penetration. Ram hard, keeping in mind that with hard rammed cores other means for eliminating veining must be provided.

Square vs. Round Test Specimens: Early in the test work it developed that a 1 in. square specimen was superior to a 11/8 in. round test specimen for study-

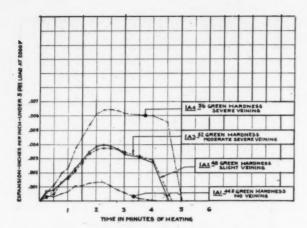


Fig. 3—Degree of veining increases with an increase of expansion under 5-psi load when collapsing time is substantially the same.

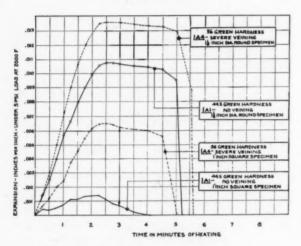


Fig. 4—Square specimens show veining tendency more than round specimens.

ing veining tendencies. A graphical illustration of the sensitivity of the square vs. round test specimen is shown in Fig. 4. Comparison is made between cores rammed to 44.5 and 56 green hardness. The difference in height of graphs between the expansion under 5-psi load for the 44.5 and 56 green hardness is greater in the case of the square test specimen than for the round test specimen. The square test specimen was thus used for all expansion under 5 psi load-collapsing time tests.

Veining vs. Metal Temperature: Since metal temperature at time of pouring the mold is an influencing factor in casting quality, it is of interest to know how the metal temperatures will influence veining. The relationship is shown in Fig. 5. With a 2400 F metal temperature no veins were present on the test castings. At 2600 F slight veining tendencies were present, and at 2700 and 2750 F severe veining was experienced. On the basis of this test 2650 F was selected for the pouring temperature of the remaining castings.

Veining vs. Baking Time: The degree of core bak-

ing in this study was altered by varying the length of baking time so that the cores were obtained of the following order of bake: green, under-baked, optimum-baked and over-baked. The influence of the degree of bake on veining is illustrated in Fig. 6. The green cores produced no veins on any of the test castings. The under-baked cores produced moderate veining, while the optimum- and over-baked cores produced severe veining defects. Thus a well-baked core will produce veining tendencies, while underbaking will not produce veining tendencies. Good coreroom practice, such as hard ramming and good baking, is conducive to veining, and corrective means must be provided to avoid veining.

Veining vs. Moisture: An increase in moisture content increased the veining tendency of a core containing 1.43 per cent linseed oil and 1.93 per cent cereal binder in fine silica sand. A moisture content of 2.9 per cent produced only moderate veining, while both 3.9 and 4.8 per cent moisture produced severe veining. The curves in Fig. 7 show how expansion under 5-psi load and collapsing time were affected by a change in moisture in this sand.

When a mixture of  $\frac{2}{3}$  lake sand and  $\frac{1}{3}$  bank sand was bonded with 1 per cent cereal binder and 1 per cent linseed oil and tested there was no veining

b

C

sl co se w sl tr

fie

h

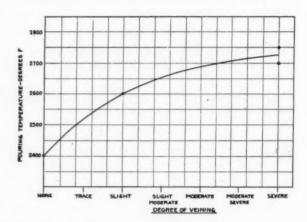


Fig. 5—Increasing pouring temperature increases severity of veining.

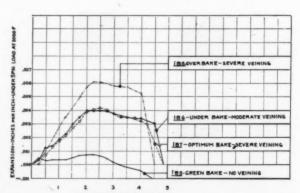


Fig. 6—Longer baking time increases veining by increasing unaccommodated expansion.

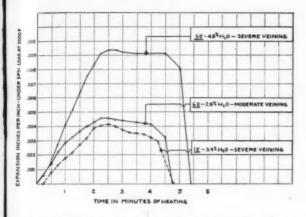


Fig. 7—Increase in moisture content increases unaccommodated expansion under 5-psi load, and also increases veining without materially changing collapsing time.

throughout a moisture range of 1, 2, 4 and 6 per cent. Veining vs. Grain Size: The grain size of silica core sands studied ranged from 129.5 to 36.1 A.F.S. Fineness Number. These washed and dried sands were bonded with 1.47 per cent linseed oil and 1.93 per cent cereal binder. All of these cores showed severe veining tendencies. When 1 per cent oil and 1 per cent cereal was tried the severe veining persisted.

Veining vs. Base Sands: Cores No. 1, 11, 12 and 13 were used to study the effect of changing the base sands on veining. The composition of these cores is shown in Table 1. A graphical presentation of the expansion under a 5-psi load and collapsing time of cores is shown in Fig. 8. The No. 1, Test G cores, composed of washed and dried silica sand, showed severe veining tendencies. The cores No. 11, Test G, with Michigan lake sand as the base sand, showed slight to moderate veining. This sand contains a trace of clay substance and fluxes which prove beneficial in retarding veining. The cores No. 12, Test G, have a base of  $\frac{2}{3}$  Michigan lake sand and  $\frac{1}{3}$  Michigan bank sand. The latter contains clay substance on the order of 0.5 per cent. This base sand produces cores resulting in slight to moderate veining.

The No. 13, Test G cores, in which a washed and

dried silica sand was mixed with an equal part of heap sand and bonded with pitch, produced only a trace of veining. It should be recognized that the test casting constitutes a severe veining test.

Veining vs. Additives: Additives were added to the No. 1 core mix which veined severely. This core mix is composed of 96.6 per cent 129.5-mesh washed and dried silica sand bonded with 1.47 per cent linseed oil and 1.93 per cent cereal binder. When 2 per cent of a cellulose flour type product was added, the cores produced moderately severe veining (Fig. 9). An addition of 0.75 per cent western bentonite resulted in cored surfaces showing severe veining. The same is true when 9.8 per cent silica flour is added. However, when 2 per cent iron oxide was added no veining occurred (No. 7, Test F cores).

Veining vs. Percentage of Oil: The next several tests were made using  $\frac{2}{3}$  lake sand and  $\frac{1}{3}$  bank sand. In this test no cereal was used and moisture was held constant at 2 per cent. Linseed oil (0.5, 1 and 1.5 per cent) produced no veining. When the oil content of the mix was raised to 2 per cent, moderately severe veining resulted. This test was repeated with 1 per cent cereal added to the mixtures. Again no veining was encountered with less than 2 per cent oil. When 2 per cent oil was used, severe veining was encountered.

Fig. 8—The unaccommodated expansion of a group of cores containing different base sands.

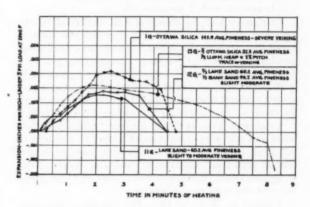


TABLE 3-SCREEN ANALYSIS OF BASE SANDS

	Retained on Screen, %										
Screen No.	Muskegon Lake Sand	Juanita Bank Sand No. 1	Juanita Bank Sand No. 2	Wedron W. & D.	80 Mesh Ottawa W. & D.	60 Mesh Ottawa W. & D.	35 Mest Ottawa W. & D.				
6			*								
12											
20											
30	0.1					0.1	0.3				
40	5.4	0.4	0.1		0.2	0.1	49.5				
50	57.9	2.2	0.3		1.7	0.2	34.4				
70	40.2	10.6	2.6	0.2	29.3	36.4	9.7				
100	5.8	39.4	41.4	9.4	40.2	58.2	3.2				
140	0.5	30.6	32.6	48.2	19.6	4.8	0.5				
200	0.2	14.0	16.6	26.0	7.0	0.1					
270	0.1	2.6	4.4	10.2	1.6	0.1					
Pan	0.2	0.1.	2.0	5.9	0.3						
.F.S. No.	49.46	89.58	96.9	129.5	77.05	64.21	36.1				

Veining vs. Percentage of Cereal Binder: In this series the oil was held constant at 0.75 per cent, the cereal was 0.5, 1, 1.5 and 2 per cent with moisture adjusted for proper workability. No veining was en-

countered with any of these cores.

Veining vs. Additives: This series of tests was made with the  $\frac{9}{3}$  lake and  $\frac{1}{3}$  bank base sand which seemed to be a characteristically non-veining sand. It was bonded with 1 per cent cereal binder and 2 per cent linseed oil which, in Test 18-H, promoted moderately severe veining. An addition of 2 per cent iron oxide eliminated all veining. An addition of 1 per cent perlite reduced the veining to slight. Two per cent of a formulated binder containing high- and low-temperature ingredients added to this sand eliminated all veining. An addition of 2 per cent wood flour reduced the veining to a trace.

Veining vs. Bentonite Additions: The same base sand was used in this test. No cereal was used, and the moisture was held to 2 per cent. The percentage of oil was adjusted to compensate for the absorbtive effect of the bentonite. With 1.5 per cent oil and no bentonite a trace of veining was observed. This trace of veining was unchanged by 0.5 per cent western bentonite with 2 per cent oil, and 1 per cent western bentonite with 2.5 per cent oil. When 0.5 southern bentonite and 2 per cent oil was used the veining became slight to moderate. It is interesting to note that the mix with 0.5 per cent southern bentonite and 2 per cent oil produced 129 psi tensile strength against 77 psi tensile strength from 0.5 per cent western bentonite with 2 per cent oil.

Veining vs. Varying Blends of Lake and Bank Sands: One per cent cereal and 1 per cent oil was used to bond the sands in this series. The all lake sand cores produced a trace of veining. Seventy-five per cent lake sand, 25 per cent bank sand, 50 per cent of each, and 25 per cent lake sand–75 per cent bank sand produced castings with no veins. The natural impurities in these sands appear to reduce veining tendencies.

Veining vs. Type of Cereal Binder: Two parts of lake sand and one part of bank sand with 0.75 per cent of oil were combined with 2 per cent each of heavy-weight corn flour, lightweight corn flour, dextrin, and a lignin liquor (powder form). None of these mixes showed any veining tendencies.

Veining vs. Type of Core Oil: In this series lake

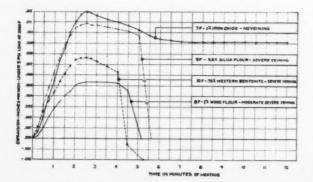


Fig. 9—Core expansion under 5-psi load may be high and still not cause veins providing collapsing time is high.

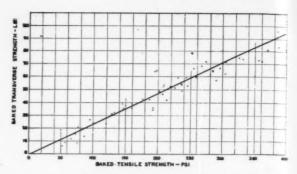


Fig. 10—Relationship between baked tensile and transverse strengths,

sand-bank sand blend was mixed with 0.5 per cent cereal, 2 per cent moisture, and 1 per cent each of A.F.S. reference linseed oil, an amber colored core oil with 156 saponification number, a dark colored petroleum polymer core oil with 84 saponification number, and a fast baking core oil. No veining tendencies appear in this group.

Veining vs. Synthetic Resins and Other Binders: The same base sand and 0.5 per cent cereal was used in this series. With 1 per cent liquid urea-formaldehyde resin or 1 per cent liquid phenolic resin no veining appeared. The mix bonded with 2.5 per cent oilless binder produced moderate veining. A mix bonded with 2 per cent resin binder produced slight veining.

#### Baked Tensile and Transverse Strength Relationship

The question frequently arises as to the relationship between baked tensile and transverse strength of cores. The Committee tested all cores, numbering to 58 different mixes, for both A.F.S. tensile and transverse strengths.

Plotting the test data so secured produces a straightline relationship graph (Fig. 10) between tensile and transverse strengths. No particular advantage was indicated by either test with reference to which provides the better measure for strength of cores. The cores in both methods of testing fail in tensile stress. The top section of the transverse core is in compressive stress while the lower section is in tensile stress. Since the tensile strength of a core is lower than the compressive strength, the core in transverse fails in tensile first.

#### Summary

The Committee has established that the veining defect can be produced or eliminated by selection of base sands, binders, metal pouring temperature, degree of ramming, degree of baking and the percentage of binders and, in some cases, moisture. The early work seemed to indicate that a good correlation between expansion and collapse time of the cores and veining could be expected. However, too many exceptions were found. As yet no test or combination of tests that will enable a foundryman to predict the veining tendencies of cores with certainty has been discovered. The Committee will continue to explore elevated temperature testing particularly, with confidence that the answer to this elusive problem will be found.

# PRODUCTION OF HEAVY GEAR WHEELS IN SPHEROIDAL GRAPHITE CAST IRON

By

Dr. Corrado Galletto*

#### Introduction

This paper depicts a characteristic Italian application of magnesium-containing spheroidal graphite cast iron in the manufacture of heavy gears with cast teeth. The use of heavy gears weighing from 2 to 5 tons is a well established practice in Italy and it is hoped that the illustration of a method used by an Italian foundry will interest many foundry technicians and engineers.

This new material which we call spheroidal graphite iron or in short S.G. iron, and which has been referred to by others as nodular or ductile iron is well known, so are its properties which make it different from the other types of cast iron. Therefore it is not thought necessary to refer in detail in this paper to such points as shape of graphite, structure of matrix, production method, types of magnesium addition alloys, response to thermal treatment and engineering properties in the as-cast and heat-treated conditions.

#### Organization of the Foundry

G. Tagliabue & Co., of Milan, Italy, operates a jobbing foundry of medium capacity with a high industrial status. In addition of producing a wide range of castings for use by themselves and by others, they have specialized in the manufacture of crown gear wheels and of gears of all types and forms ranging in diameter from few inches to 20 ft, and in weight from few pounds to 15 tons.

In the molding shop there are special installations for the production of gear wheels.

Melting equipment comprises four cupolas from 36 to 30 in. in diameters, a graphite-lined electric arc furnace of 3 tons capacity, and two electric resistor units of about 700 lb each.

There is a well equipped machine shop for finishing of general castings and gears in particular.

The works has a chemical and physical laboratory equipped for the control of molding sand, for chemical and microscopical examination and for mechanical and physical testing.

#### Production of S.G. Iron

At the time of the first announcement with regard to S.G. iron, Messrs. Tagliabue, having foreseen that the properties of this material would open a wide range of applications in their own specialized field, were among the first Italian foundries to apply for license to manufacture.

The first tests were carried out in May 1950. The results which were satisfactory from the beginning, strengthened the opinion of Tagliabue regarding the suitability of S.G. iron for their products and stimulated production on an industrial scale.

At present the metal poured into small castings is melted in the resistor furnaces, while that for heavy castings is taken from the cupola or the electric arc furnace. Occasionally, when it is desired to increase the speed of treatment, use is made of cupola and arc furnace metal together.

The pig iron from Cogne, used as the basis of the charges, has the following composition:

These charges also include selected mild steel scrap varying in proportion according to the required composition of molten iron and the method of melting employed.

Generally foundry returns from previous melts are not used because their nickel contents are preferably reclaimed by their inclusion in charges from which alloy cast irons of flake-graphite types are produced.

The composition of the liquid iron from the electric furnaces prior to treatment is normally:

and the tapping temperature is in the range 2550 to 2650 F.

Magnesium alloy, in the form of large lumps, is added in the required proportion. The magnesium-treated iron is then inoculated by adding granulated Fe75Si and the properties exhibited by the as-cast iron are normally in the range:

^{*} Centro d'Informazioni del Nickel, Milano, Italy.





Fig. 1 (Above)—Large scale divider for molding of teeth. Fig. 2—A close-up view of Fig. 1 showing in detail the

pattern holder; a second metal tooth pattern can be seen on the sand mold.

T.S. = 92,000 to 100,000 psi % Elong. = 3 to 5 per cent

Brinell = 220 to 230 measured 0.2 in. below the metal surface as determined in test pieces cut from 1-in. thick keel blocks.

The composition of the cupola metal is:

C Si Mn S 3.70-3.80 1.65 0.40 0.07

and its temperature is in the range 2520-2550 F. After the magnesium treatment and inoculation, the properties obtained on the 1-in. thick keel block are:

T.S. = 100,000 to 107,000 psi.

% Elong. = 4 to 5 per cent on 2 in.

Brinell = 240 to 250 measured 0.2 in. below the metal surface.

The composition of the molten iron is adjusted in accordance with the properties desired in the finished castings and depending on the preference for highest tensile strength, or maximum elongation, or greatest hardness for resistance to wear.

The machinability is always good and in the ascast condition S.G. iron can be machined as readily as high-strength, flake-graphite iron despite its higher hardness.

The fluidity of S.G. iron is equal to and even better than that of high-duty, flake graphite cast iron, and for the casting of heavy pieces no more precautions are necessary than those commonly taken for highstrength irons.

Practical experience has shown that, contrary to original expectations, contraction across the diameter, which is a point of great importance in castings of circular shape, does not greatly differ from that normally encountered in similar castings poured in high-strength, flake-graphite irons.

#### Method of Molding Gear Wheels

The method followed in the preparation of gears with cast teeth is characterized by the fact that full patterns are not employed. Use is made of the equipment and procedure now to be described.

The main outlines of the wheel are first strickled from a central spindle. Subsequently, the teeth are individually molded in position, by use of a highly finished metal pattern held against the face of the swept mold. The shape and dimensions of the tooth pattern and the circumferential distance between the

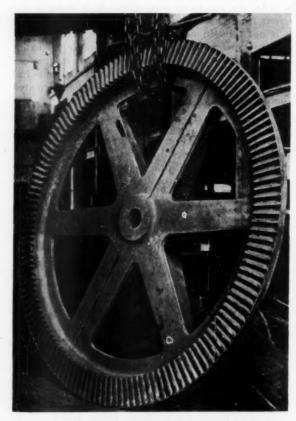


Fig. 3—A bevel gear made in S.G. iron—diameter 12 ft, weight 3500 lb.

teeth centers are varied according to the design of the gear. A circular table on which the mold is positioned and which is rotatable forms part of each machine. Its movement is controlled by a system of gears which are contained in the attached mechanism. This also provides a head which is capable of movement outward and inward, and up and down. The complete machine is shown in Fig. 1.

The equipment can be regarded as a large scale set of dividers in which, by suitable adjustment of the actuating train of gears the movement of the table effected between the ramming of each individual tooth is exactly that required to provide the specified number of equally-spaced teeth on the wheel. Much care is taken to ensure absence of "play" in the transmission and in the table bearings, so that movement is precisely controlled. The tables used for molding of smaller gears are manually operated but the larger ones are electrically driven because the heavier weights and greater diameters involved would make manual operation laborious and would affect the accuracy of the procedure.

The arrangement for positioning of the tooth pattern will be clear from Fig. 1. The hand-operated gear which provides outward and inward movement of the main arm can be seen at the top right-hand side of the photograph and it will be noted that the combined use of a long sleeve and opposed splines prevents movement of the head in any direction other than that described.

The second hand-operated gear, which raises and lowers the vertical arm is clearly seen above the head of the operator and in this instance also, use of sleeve and double spline effectively prevents any movement other than up and down.

Movement of the pattern holder is effected by a third hand-operated gear which is shown in greater detail in Fig. 2. This fixture effects movement of the pattern holder and pattern outward from the mold face and it is therefore normally unnecessary to use the other movements during the ramming of the wheel and after the initial positions of the horizontal and vertical arms have been established. In those instances in which there is insufficient space behind



Fig. 4—Equipment for molding of teeth in large gear wheels which are pit molded. Note the front hand wheel which moves the mechanism controlling angular movements.



Fig. 5—Crown wheel after stripping with the two downsprues, and the continuous runner with the ingates and feeder heads.

the pattern to permit withdrawal in this manner, the operator, by use of the second gear earlier described and which is easily within his reach, can free the pattern from the mold by upward movement, and so clear the table ready for rotation to the next position. At this point it should be noted that the tooth pattern is always deep. This makes it possible to use the same pattern for a considerable range of wheel depths.

When the swept face of the mold is ready the operator fixes to the holder the pattern which corresponds to the type of teeth required and by means of the counter-poised crank handle he moves the pattern away from the face of the mold.

Then by movement of the large horizontal arm, he positions the pattern almost in contact with the swept face approximately in the working position. He then secures the large arm in the position thus reached.

The operator is provided with a data card from which he can readily determine the assembly of gears and the number of precise revolutions of the main gear required to move the table exactly the amount which corresponds with the pitch of the teeth. Then the gears are assembled and the machine is ready for molding of the teeth.

By operating the counterbalanced crank, the molder then places the pattern in slight contact with the vertical face of the swept mold, locks it in this position and, by movement of the vertical arm, he then lowers the pattern until it reaches the bottom of the swept cavity. Thereafter into the cavity which exists between the tooth pattern and the wall of the mold, he introduces a suitable sand mixture which he then rams by hand until the cavity is filled to its upper face.

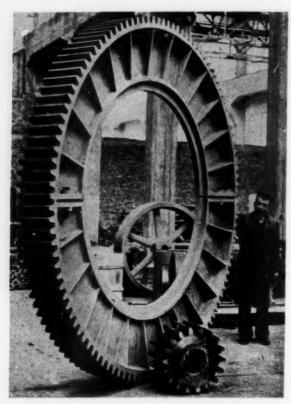


Fig. 6—Heavy crown gear wheel, weighing 9900 lb, with pinion.

To ensure a smooth surface and perfect contour, the sand used normally consists of about equal proportions of graphite and a strong molding sand. Where the teeth are sufficiently heavy to require reinforcement, supporting irons are rammed in position.

When ramming of the first tooth is completed, the molder unlocks the pattern holder and then raps the pattern slightly before withdrawing it from the mold face by use of the crank. If there is insufficient space for the pattern to be disengaged completely from the mold by horizontal movement in this way, the vertical arm is raised in order to clear the mold. The table is then rotated the prescribed amount and molding of the second tooth proceeds by the same sequence of operations. This procedure is repeated until all the teeth are molded in positions as precisely set as in a machined gear.

Bearing in mind the fact that the teeth are perfectly molded and completely ready to be cast, the method described does not occupy more time than molding from a full pattern. In the latter case the teeth contours are seldom perfect and repair of the mold is almost necessary.

In the case of bevel gears such as that shown in Fig. 3, the molding process remains the same with the sole difference that the teeth are molded on the conical face corresponding to the outer part of the gear instead of being applied vertically on the cylindrically swept mold as in the case of normal gears.

Even heavy helicoidal driving gears are made in

the same way by using special tooth patterns which do not vary only according to the modules but also according to the transmission ratio.

Gear wheels up to 12 ft in diameter are molded in the manner which has been described and for this purpose six machines of various sizes are employed. The molding boxes used for gear wheels so produced should be circular and should also be of strong and rigid design in order to resist deformation.

Larger gears are pit molded in the manner illustrated in Fig. 4. It will be seen that in this procedure, the mold cannot be moved and therefore an equipment for suspension of the tooth pattern, mounted on the central spindle, is caused to move around inside the swept mold. The mechanism for controlling angular movements between the ramming positions remains the same.

#### Gating and Feeding

Spoked wheels, such as that shown in Fig. 3, are gated directly into the spokes. Spokeless wheels are gated in the manner shown in Fig. 5, that is, by use of two downsprues and a long continuous runner supplying many well distributed ingates.

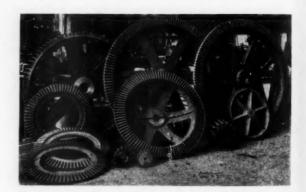


Fig. 7—Spoke and spokeless bevel gears, crown wheels and pinions of various sizes.

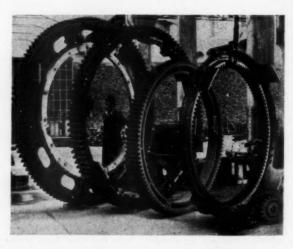


Fig. 8—Other types of crown gear wheels with internal or external teeth.

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Spoked wheels are fed from heads positioned above the hub and at intervals around the crowns. Spokeless wheels are provided with feeder heads positioned above the heavy sections which serve as the points of attachment to the driven body.

The metal in all feeder heads is covered by an exothermic compound to maintain temperature during solidification of the casting.

#### **Practical Results**

The castings which are put into service without heat-treatment normally exhibit following mechanical properties:

Ultimate tensile strength: 92,000 to 100,000 psi Elongation: 3 to 5 per cent Hardness tested on the teeth is 270-290 Brinell.

Impact tests conducted on notched and unnotched specimens of the Charpy type have given the following values:

Test piece 10 x 10 x 55mm unnotched: 1.5 to 3.0 Kgm/cm² Test piece 10 x 10 x 55mm notched: 0.5 to 0.7 Kgm/cm²

These excellent properties, especially the high hardness of the surfaces which are subject to wear and the good resistance to impact, allied with the



Fig. 9—Side frames of rivet press weighing 3000 and 6000 lb.



Fig. 10—Parts of rivet press made in S.G. iron. Total weight of S.G. iron in presses: 3500-7250 lb.



Fig. 11-Bogie Wheel



Fig. 12-Flexible joint

good surface obtained on the cast teeth which are produced with high precision to theoretically perfect profile, make it possible to replace, with advantage, cast steel gears with machined teeth and which are more laborious and more expensive to produce.

Figures 6, 7 and 8 show various types of gear wheels which are in regular production in S.G. iron in the foundry of G. Tagliabue. In equipment for manufacture of cement, such as rotating kilns and grinding mills, for mixing of rubber, for milling of ceramics and other materials, in hoisting mechanisms, and in many other driven constructions these gears are finding ever-increasing applications.

Comparative service data are not yet fully available since the first installations of S.G. iron gears were made only about a year ago. Nevertheless, the indications from the early observations, especially the checks on resistance to wear in the important initial stages of adjustment of the teeth with one another show that there is likely to be a big field of applications for this new iron in such gears with considerable saving in weight and with service life at least doubled.

#### Other Tagliabue Products in S.G. Iron

As stated initially, the production of this foundry is not confined to gear wheels and many types of castings are made, some in large numbers. The illustrations which follow depict some of the more important applications which have already been proved in service and for which repeat orders have been placed.

One such application is shown in Figs, 9 and 10. These illustrate, respectively, the side frames, supports and various components of rivet presses, all made in S.G. iron. The side frames shown weigh approximately 3,000 and 6,000 lb and the total weight of S.G. iron in the presses ranges from 3,500 to 7,250 lb.

A casting of relatively small size, but which is being made in hundreds, is the bogie wheel shown in Fig. 11. These wheels are fitted to skips which are used for conveyance of material in mines and on civil engineering projects of many types.

The flexible joint in Fig. 12 uses S.G. iron for the male and the female sections. The inserts of hard rubber which engage with the driving 'dogs' are clearly seen in the driven section of the coupling. Such joints, which are a Tagliabue design, are used for heavy duties of many types in cement mills, paper making equipment, rubber manufacturing plants, etc.

Other items include bodies for submersible pumps, impellers; chain wheels, etc. Of particular interest are the end rings for collecting the current in electromagnetic joints and reversers employed for example in planers and other machines which require reversible motion.

#### Acknowledgments

In closing I wish to thank Ing. Habe, technical Manager of Tagliabue foundry, for his kind collaboration.

#### DISCUSSION

- Chairman: J. S. Vanick, International Nickel Co., New York. Co-Chairman: W. S. Pellini, Naval Research Laboratory, Washington, D. C.
- Washington, D. C.

  Recorder: Chas. Mooney, Olney Foundry Div., Link-Belt Co.,
  Philadelphia.

- MEMBER: Were the castings you mentioned stress relieved.

  MR. GALLETTO: No, the castings were in the as-cast condition.
- Member: How long have the gears you illustrated been in operation?
- MR. GALLETTO: The first gears have been cast late in 1950 and have started actual service in March 1951. Service data are not yet available as such gears are supposed to last from 5 to 8 years. It has been observed, however, during the period of setting of the teeth, that their wear resistance is better than that possessed by prior materials, consequently final results should be far better.
- MEMBER: Do you use feeding risers on the large gears?
- MR. GALLETTO: Yes, see Fig. 5; the spokeless wheels are provided with a feeder which is located on the heavy sections which serve as attachment to the driving body. Spoked wheels are fed by heads positioned above the hub at intervals around the crown wheel. In all feeder heads molten metal is covered with an exothermic material.
- MEMBER: What sand mixtures were these gears cast in? What is the green strength of these sand mixtures?
- MR. GALLETTO: Small gears are cast in green sand while gears up to 10 ft in diam are molded in flasks and furnace dried. Gears of large diameter, which are pit molded, are dried on the spot by means of hot air.
- For sweep molding, use is made of a strong sand called "Refrancore" having a P.C.E. cone of 2.5 mm in the refractory test at 2570 F and melting point of approximately 2570 F.
- For mold facing the new sand is mixed to 60 per cent of used sand. Binders are not used because this sand contains about 17 per cent clay substances.
- In molding of small teeth a mixture of 50 per cent graphite plus 50 per cent "French sand" is used. This sand (finely ground sand suitable also for bronze castings) contains 20 to 24 per cent clay substances; moisture content is 3 to 4 per cent.
- In molding large teeth we use 50 per cent graphite mixed with 50 per cent weak sand of Bassano which contains 3 to 5 per cent of clay substances; it softens at P.C.E. cone of 1 mm in a refractory test at 2372 F and has a melting point at 2912 F. 4 to 5 per cent bentonite is added; moisture content is 3 to 4 per cent. All sand mixtures for molding of teeth should be thoroughly mulled and mixed. This is of great importance.
- MEMBER: What is the tensile strength and elongation of the iron?
- Mr. Galletto: The tensile strength of the iron is 90,000 to 100,000 psi. The elongation in 2 in. is 3 to 5 per cent. These values are determined on test bars from the keel block. Hardness tested on teeth is 270-290 Brinell.

# REFINING AND ELIMINATION OF GASES FROM METALS AND ALLOYS

By

#### Henry Lepp,* Florent Monneret,** and Paul Rame;

#### ABSTRACT

Starting with the theory of refining and degasification of metals through the method of selective oxidation, the authors proceed with description of application of this process.

After description of the reverberatory furnace they show how the brass castings of high mechanical characteristics can be made from reconditioned scrap metal without the use of ingots. The paper is divided into three parts:

- Resumé of the method of selective oxidation.
   Description of a furnace allowing the application of the procedure on industrial scale.
- 3. Description of one melt or heat:
  - a. Performance of furnace.b. Control of metal before pouring.
  - c. Results obtained.

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A few years ago there was a common belief among foundrymen that most problems they encountered in production of sound castings were due to oxidation of metal. Today this opinion does not hold true. Numerous research works have established that oxidation of metals was less dangerous than the contamination by reducing gases, hydrogen particularly. Several procedures of gas elimination have been advocated, such as melting in vacuum, pre-solidification, and passing of "neutral" gases through the bath of metal, etc.

In spite of usage of the above-mentioned methods, the problem of metal degasification was not entirely solved. Some of those procedures provided degasification only in part, the others required special and expensive equipment.

To overcome those inconveniences a special method has been developed and applied over a period of 25 years with satisfactory results. This metallurgical process is based on the principle of selective oxidation of impurities which may be metals or metalloids.

The authors have previously published the fundamentals of this method of "selective oxidation" of un-

desirable impurities.^{2,3} To clarify their statement the authors will briefly describe it.

The essential requirement for effective operation of selective oxidation of an undesirable impurity is that the resultant oxide has a higher heat of formation than the oxide of the metal itself (or metals composing an alloy).

More precisely it is necessary that the variation of free energy ( $\triangle F$ ) in action during the process of oxidation of impurity takes place in temperature which is higher than that of the formation of oxide of the metal or alloy to be refined.

The type of oxidizer should be selected depending upon the type of metal to be treated and impurity to be eliminated.

It is therefore essential that the atomic oxygen of the oxidizing agent if free and available at the temperature proper for its combination with the impurity without coming in contact with the basic metal. The following chemical reaction should proceed from right to left.

$$R + (O) \leftarrow RO + Q cal$$

Where RO = Oxidizing agent

R = Chemical element (metal metalloid or their compound).

(O) = Oxygen available at the temperature required for reaction with the impurity.

Q = Heat of formation of RO

There are many types of oxidizers which can be used for treatment of metal on a large industrial scale. It has to be mentioned, however, that the simplest, namely, the oxygen from atmosphere, unfortunately cannot be used. Such a process would require the dissociation of the  $(O_2) = 2$  (O) which would be difficult to control and stop at the required moment. In other words the oxidation method using the oxygen from the air could not be "selective" with regard to the respective impurity.

Following are the oxides listed in the sequence of their efficiency:

Cr₂O₃, Fe₂O₃, MnO₂, CuO, ZnO peroxides BaO₂, etc., manganates, chromates, etc., finally nitrates, chlorates, perchlorates, etc.

^{*}Doctor Engineer, **Civil Engineer in Maritime Engineering, and †Engineer of Arts and Trades and Head of Foundry School.

The manuscript was translated from the original French by H. W. Zimnawoda, Export Manager, National Engineering Co., Chicago.

For the illustration of the principle, let us take the following two examples:

 Addition of the oxides of copper to the copper bath.

It is evident that CuO or  $Cu_2O$  cannot oxidize the copper. They can be dissolved in the fluid copper or oxidize the lower grade chemical elements which could be found in the bath, for example:  $Zn \rightarrow ZnO$ ,  $P \rightarrow P_2O_5$ ,  $Fe \rightarrow FeO$ , etc.

2. Addition of ZnO to the copper bath.

This addition cannot be active against either copper or zinc. However, it is able to oxidize other metals or elements having the heat of oxide formation (or  $\triangle F$ —variation of free energy) higher than of ZnO.

The addition of these oxidizing agents can be measured and consequently the process of oxidation of impurities can be controlled and limited if desired. Beside that, this procedure is selective regarding the choice of oxidizing agents.

The products of oxidation, the oxides, are practically not soluble in the liquid bath. When they appear in the form of gases, they escape easily into the atmosphere. In the liquid form they come up to the surface of the bath. As solids they have the tendency to remain in suspension.

A fluid slag should be created to assist the reaction of metal refinement by absorption of oxides to be eliminated which are the chemical compounds of third degree, such as silicates, manganates, borates, aluminates, etc. In most cases these compounds are stable and hard to reduce.

A practically complete degasification can be achieved by proper selection of oxidizers and composition of slag, capable to retain the oxides.⁴

The type of oxidizing agent to be used for a given alloy and impurities to be eliminated should be established by thermodynamic analysis. This system of analysis and method to be applied was described before.^{3,5,6}

Practical application of the method of selective oxidation adapted to different metals and alloys enhances the physical characteristics.^{2,3,4}

In this paper the authors present certain details on the application of the above mentioned method of selective oxidation as used in the production of screw propellers made of special brass of high tensile strength.

The scarcity of raw materials makes the problem of metal refining an important one. The non-ferrous industry might find itself in very unfavorable position if this problem were not solved satisfactorily.

Utilizing large tonnage of metal for manufacture of propeller screws, the authors studied the question of a furnace adaptable for treatment of scrap metal and rejects by the degasification method described below.

To obtain best results from this procedure it was necessary to provide for intimate and prolonged contact between the charge of the metal, the oxidizing agent and the slag.

It was decided that a reverberatory type furnace would best serve the purpose. The authors used a furnace of this type of 13 to 15 tons capacity.

#### 13- to 15-Ton Reverberatory Furnace

The shape of the furnace, that is its length and diameter were selected to attain the most favorable proportion or ratio between the volume and the surface of the bath of the metal. The problem of heat transfer has been taken into consideration.

The next problem was to provide a burner capable of burning 350 kg of fuel per hour.

The burner should allow:

- 1. To decrease melting time to a minimum.
- To obtain an adjustable length of flame to be regulated according to the obstacles provided by the metal charge.
- A long or short flame with the possibility of having it oxidizing, neutral or reducing.

The last problem was solved by:

- a. Use of high pressure blower.
- Flexibility of burner performance described below.

The fuel is introduced into the burner through a movable tube which terminates in a nozzle (Fig. 1). The movement of this tube provides a change of the angle of the flame.

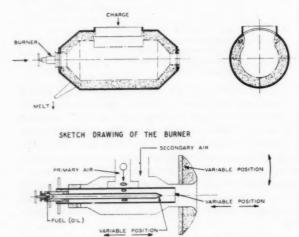


Fig. 1-Sketch of furnace used in this study.

Air is introduced through a lateral or side pipe. It is used as primary and secondary air for pulverization of the fuel.

The regulation of flow of air, its high pressure and the possibility of the regulation of the nozzle results in a short or long flame of high temperature well located inside the furnace. In addition the burner can be moved around its horizontal axis, thus enabling to direct the flame to any desired spot.

This flexibility of burner operation results in maximum thermic efficiency and shortest possible melting time.

By accurate control of the flow of the air and fuel it is easy to obtain the required atmosphere inside the furnace: oxidizing, neutral or reducing. This feature is important from the metallurgical point of view.

Oscillating Movement of Furnace - The oscillating

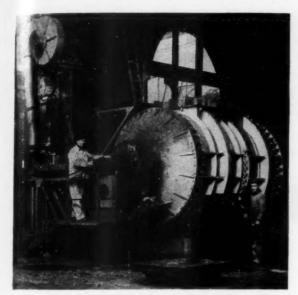


Fig. 2-Photo of furnace used in this study.

movement is effected by means of a chain connecting the furnace with an hydraulic cylinder. The movement can be automatic or manual. The moving operation of the furnace is fast and easy. The furnace can be started, put in reverse, stopped and started again almost instantly.

Charging the Furnace—A door located at the top of the furnace permits charging of large pieces. There is no need to cut material into small pieces. The savings obtained are appreciable and the charging can be done rapidly. (Figs. 1 and 2.)

Due to all the refinements of design, as described above, the operations of charging, melting and tapping require the service of only one man.

The melting time of a 13-ton charge (most suitable for this type of furnace) starting from cold scrap to the uniform metal bath temperature requires 5½ to 6 hr. It has to be noted that the charge is composed of old propeller screws, scrap and turnings.

#### Melting and Refining Procedure

To protect the metal bath during melting against the direct contact with the flame the authors add to the charge 3 to 4 percent of slag.

To avoid excessive metal losses during the melting process of brass with high percentage of Zn it is necessary to maintain a reducing atmosphere in the furnace

It is known that up to approximately 1200 C, CO₂ acts as an oxidizer of zinc. This reaction should be avoided. Above 1200 C, the increase of entropy or zinc gases is sufficient to allow for co-existence of CO₂ without its entering into chemical reaction. At that moment (1200 C), to assist the degasification of the bath, the flame has to be changed from neutral to slightly oxidizing. This is also the time when the refining process starts.

The method of selective oxidation is performed by judicious addition of a mixture of slag, copper oxide and manganese dioxide.

To check the efficiency of this treatment and its influence on the composition of the alloy the authors use the following control tests:

#### 1. Composition of the Alloy.

Rapid methods of chemical analysis are used to check the composition of the liquid metal and to make the corrections of certain elements such as Fe, Mn, Al, Zn, etc. The chemical analysis which require over 1 hr could be shortened considerably by application of spectrographic analysis.

#### 2. Micrographic Examination.

This test consists of a comparison of the cast specimen, cooled according to established procedure, with a standard specimen. This check requires 10 to 15 min and the results are sufficiently accurate for production control purposes,

After the desired chemical composition is obtained, the authors proceed with tests to verify the efficiency of the method of refining and degasification.

Following are tests used: (1) Shrinkage test, (2) Torsion test, and (3) Density test.

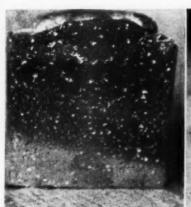


Fig. 3—Photo of cross-section of untreated metal. Note gas pockets.



Fig. 4—Photo of cross-section of treated metal. Note freedom from gas pockets.



Fig. 5—Photo of another sample of treated metal. Also note freedom from gas pockets.

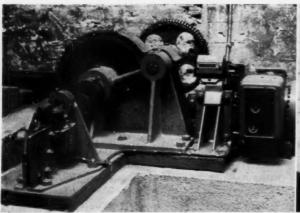


Fig. 6-Photo of torsion testing machine.

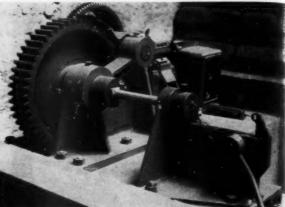


Fig. 7—Torsion testing machine with specimen ready for test.

#### 1. Shrinkage Test.

The specimen used for this test is cast in open and dried molds; it is 100 mm in diameter and 120 mm long.

After solidification the examination of shape of the upper surface gives the first indication of the degree of degasification.

Figures 3, 4 and 5 show the cut specimens and demonstrate the excellent results obtained. It can be seen that the very porous metal was changed to sound metal.

#### 2. Torsion Test.

When elimination of gases is achieved and the shrinkage test considered satisfactory, the authors proceed with the torsion test.

This test is made in a special machine. See Figs. 6, 7 and 8. The as-cast specimen is twisted until rupture occurs. Its results indicate the following:

- 1. A figure in relation to the tensile strength,
- 2. A figure in relation to the elongation.
- 3. Degree of degasification.

The first is shown on the wattmeter installed in the electric supply line to the motor of the testing machine.

The second gives the angle of twist at the rupture of the specimen. The scale of the tachometer is calibrated so that one revolution corresponds to 1 degree of torsion. When the specimen breaks, a device stops the machine and the angle of twist can be established within 5 degrees accuracy.

The third (degree of degasification) can be checked by examining the fracture of the specimen.

#### 3. Test of Density.

The specimen prepared for this test is cast in a sand mold similar to that for the torsion test. The density test contributes considerably to the establishment of the degree of degasification of metal. As the authors have mentioned above, the increase of density corresponds with the increase in mechanical properties.

As an example, and for the purpose of illustration of the procedure of degasification, we present figures obtained during one melt or heat. The shrinkage specimen has indicated satisfactory results of degasification. The density specimens have given the following figures: 7.9; 8.2 and 8.35. The torsion specimen taken before addition of the mixture of oxidizing agents has shown 285 degrees and 345 degrees angle of twist. After addition of oxidizer (approximately 20 min) an angle of 490 degrees was obtained (Fig. 8).

The density has also improved, at the end of treatment the density value being 8.6.

The mechanical characteristics of the test bars attached to the propeller screw, cast in sand without chills, were as follows:

 $R = 50.9 \text{ Kg per mm}^2$ 

A = 23.5 percent.

 $E = 22.7 \text{ Kg per mm}^2$ 

R + A = 74.4

In this short paper the authors attempted to indicate the technical advantages and economics to be derived from the application of the method of refining and degasification through selective oxidation of impurities.

To be able to use this method on a large scale, the authors briefly described the design of the furnace.

The principle applied allowed them to re-use the metal of old propeller screws, scrap and turnings and pour the metal direct into molds without making



Fig. 8—Photo shows specimen broken in torsion testing.

ingots or pigging. The mechanical characteristics of the metal were in accordance with the specifications of the French Navy.

The melting losses depending upon the composition of charge (turnings, small scrap, etc.) were 5 to

The following are the average figures of material properties taken from different melts:

 $R = 45-51 \text{ Kg per mm}^2$ 

A = 20-40 per cent

 $E = 20-28 \text{ Kg per mm}^2$ 

The chemical analysis of the metal has shown approximately 1 percent Fe and Al and 2 percent of Mn. The authors allowed up to 1 percent of Sn and 0.6 per cent of Pb.

The variation of the mechanical properties of the metal from one casting was about 3 percent.

It has to be mentioned that the uniformity of the results obtained is not solely the merit of the metal-lurgical process of metal refinement.

The successful result depends also upon proper sand control and completely dried molds to avoid the reverse reaction of regasification or absorption of gases by the metal during the pouring operation.

#### Acknowledgment

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#### References

- 1. H. Lepp, Bulletin, Assn. Technique de la Fonderie, March, 1937, or Metal Industry (London), vol 53, pp. 27, 59, 79, 103, 131 (1938).
  - 2. H. Lepp, Journal, Institute of Metals, vol. 57, p. 31 (1935).
- 3. H. Lepp, Revue Metalurgique, pp. 443-446 (1937).
  4. H. Lepp, "Mechanical Properties of Some Tin Bronzes,"
  Foundry Trade Journal, vol. 57, No. 1105, pp. 321-322 and 328,
  Oct. 21, 1937.
- 5. H. Lepp, Bulletin, Assn. Tech. Fonderie (1944).
- 6. H. Lepp, Metal Treatment, Summer, 1949.
- 7. Congress International De Fonderie, Prague 1933, pp. 95-119.

# INFLUENCE OF THE NATURE OF OVEN ATMOSPHERE ON DRYING OF MOLDS AND BAKING OF FOUNDRY CORES

#### By

#### Georges Ulmer* and Maurice DeCrop**

Does the nature of the atmosphere of a foundry oven in which molds are heated for drying or cores for baking have an influence on the progress of the operation of drying or baking? Such is the question which we wish to ask when, for example, the regulation of the incoming fresh air in an electric core oven is being considered. So, one might ask, what are the advantages of circulating mold ovens in which the hot fluid is only superheated steam circulating in a closed circuit.

The problem which arises is two-fold, for the dry-

ing of molds as for the baking of cores.

We must know if heating in closed circuits (confined atmospheres) is more economical than heating in open circuits (changing atmospheres), which could be determined by the comparison of the thermal output of the two processes, and if the quality of the oven, which one can evaluate by the absence of cracked molds and high strength in cores, depends on the nature of the atmosphere.

This study first shows the determination of the physical mechanism of drying with the aid of some theoretical considerations. At the same time we will give an insight into the work which has already been done on this subject. Following this we will show the results of the experiments we have carried out and the conclusions which can be drawn from them.

#### I-A Study of the Drying and Baking of Sand

#### A-Transmission of Heat in Sand

In a mold or a core the silica grains are surrounded by a film of a mixture of water and binder (clay in the case of molding sand, agglomerating agent in the case of core sand) which is bonded to them by

adhesion and capillarity. Heat transmission takes place in two ways: by conduction and by vaporization.

(1) Conduction—Heat is transmitted by the ingredients, that is to say by the sand grains, the binder and the moisture. Since the conductivity of the water and the binder is better than that of dry sand, it seems likely that the flow of heat is through the agglomerating paste and that its intensity will diminish as the drying and baking progresses, acting to cause the conductivity to approach that of the dry sand; moreover, a mold or a core which is moist and filled with the agglomerating agent will have a better conduction at the beginning of heating than at the end of the process, because its structure will then be much more open, and its density much less (presence of insulating voids between the sand grains).

This influence of moisture on the conductivity of sand is far from being negligible; an experimental study has shown that for a theoretical granular substance (composed of balls of glass of the same diameter as the silica) the addition of a very small quantity of water 3g/1) causes an increase in conductivity

of 50 per cent.1

It should be noted that in the case of cores, the decrease in the amount of heat due to conduction, during the course of drying, is compensated by the heating due to the exothermic oxidation of the binder: this is the reason why at the end of the process it is possible to have the temperature within the core higher than the ambient temperature in the oven.

Finally the conductivity for a porous substance is a phenomena which is comprised of one part the convection of the air in the pores of the material and the mutual radiation of the walls of these pores.

(2) Vaporization—While the temperature is increasing, the vapor tension of the volatile substances increases in the sand and these vaporized elements migrate toward the regions of lower vapor tension, that is to say starting from the warm areas toward the interior cooling regions where they condense, and toward the dry and warmer surface areas where

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they are superheated before being expelled. This displacement of vapors in the material goes on at the same time as the displacement of calories from the outside to the inside which is thus heated by the sensible heat of the vapors and by their heat of condensation.

B-Mechanism of the Drying of Sand

If one considers a mass of wet sand enclosed in a flask and bounded, on the other hand (side) by two plane surfaces, one sees during heating, a loss of moisture at the same time toward the outside and toward the inside where the water content has a tendency to rise in an early part of the process (the case of superficial drying). The vapor tension of the water which increases with temperature reaches the value of the atmospheric pressure (760 mm Hg) for the temperature of 100C (212F) and from this moment the vapor tension can no longer increase. Vaporization continues at constant temperature as long as any moisture remains.

In following the drying of such a mold with the aid of thermocouples placed at various depths in the sand, one will see at each point these phenomena follow in order:

- increase in temperature to 100C or 212F (evaporation).
- leveling of temperature at 100 C or 212F (vaporization).
- increase beyond 100C or 212F (drying ended, superheating).

A 100C (212F) temperature zone therefore will be displaced to the interior of the sand, from the surface toward the middle (Fig. 1) and its travel on a level with a thermocouple will be distinguished by a leveling of temperature, a classical phenomena which has been distinctly observed in the drying of molds. In the case of cores, where the moisture content is low enough, (of the order of 1 per cent), the level-

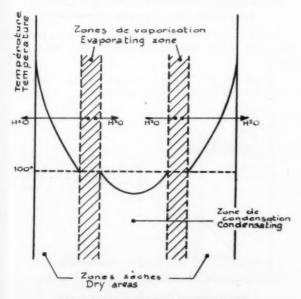


Fig. 1-Drying process for a sand mold.

ings are not very long although clearly noticeable.

In the case of a foundry oven, where the temperature is always in excess of 100C (212F) the moisture is thus driven from the sand in the form of superheated steam; one knows that beyond 100C (212F) water can no longer exist as a liquid at atmospheric pressure, and that there is no longer a degree of saturation limiting the content of the water vapor of the oven atmosphere.

It is useful to determine here that when drying at a temperature over 100C (212F) one encounters a different phenomena of drying than simple evaporation, which is carried out by the action of ordinary dry gas and at a low temperature (less than 100C or 212F) at which the efficiency is proportional to the difference between the water saturation content of the gases for their temperature and their effective content. Above 100C (212F) drying can be done as well in steam as in dry air, because nothing limits the steam content of the air-steam mixture in which the sand is soaking to dry. There is, therefore, no cause for fear that the drying may be diminished in proportion to the moisture content of the convection gas; the hygrometrical condition of the atmosphere in particular will not interfere and it will not be important to use a fuel having large or small amounts of moisture in the fumes.

The process of drying to the middle of a mold depends essentially on the flow of heat which is transmitted through the outer zones already dried toward the inner zones still moist and which gives rise to a flow of superheated steam from the opposite direction, from the interior to the exterior and toward the superficially dried zone.

The nature of the convection fluid does not influence the rate of drying (in depth) which is indirectly proportional to the convection coefficient of heat transfer. We will show in the following, by calculation and experiment, that this influence is slight when going from dry air to superheated steam. On the contrary, the effect of the temperature of the hot fluid on drying is predominant. On diagramming the phenomenon one can show this influence on a mass of very wet sand, bonded by plane surfaces (full flask constituting a test mold).

Consider the flow of heat transmitted by the convection fluid to the sand of a mold during the time dt; it is equal to

$$\frac{\mathrm{dQ}}{\mathrm{dt}} = K (T-\Theta)$$

with K = convection coefficient

T = temperature of the hot fluid

e temperature of the sand surface.

This heat flow is transmitted by conduction through the dry zone to the zone of vaporization which goes to a depth in the mold while maintaining itself at 100C (212F).

We will assume in the first approximation and to simplify the calculations, that a constant part and moreover the most important of this flow is used in vaporizing the water in the sand and that the other part equally constant is used in heating the sand, preheating the water and superheating the steam.

This hypothesis premits the writing of an equation for the fraction of transmitted flow:

$$m \frac{dQ}{dt} = \frac{\lambda}{e} (\Theta - 100)$$

with  $\lambda = \text{conduction coefficient of dry sand}$ e = depth of the dried zone

m is a constant [which we will call the coefficient "moyen de rendement"] which is equal for a mold of a given analysis to the ratio between the heat of vaporization of mold water and the total heat used in drying the mold.

The quantity of heat supplied to the zone of vaporization permits the drying of a layer de such that:

$$\frac{d\Theta}{dt} = m \frac{T-100}{l \quad me} = h L d e \tilde{\omega}$$

$$\frac{1}{k} \frac{me}{\lambda}$$

with h = moisture content of the sand

L = heat of vaporization of water at 100C (212F)

 $\tilde{\omega}$  = the specific weight of sand*

One can deduce by integrating that equation, the drying time as a function of the thickness reached and the depth

$$t = \frac{e^2}{2m} \frac{hL\tilde{\omega}}{(T-100)} \left( \frac{m}{\lambda} + \frac{2}{Ke} \right)$$

The temperature is therefore an essential factor in the speed of drying [through the depth a hyperbolic function] for a sand mold soaked in a convection fluid at over 100C (212F).

To obtain a more rapid rate of drying, the temperature of the convection fluid should be as high as possible.

C-Baking of Cores

The baking of cores to give the hardness and strength necessary for their use is at first a drying, then a process of chemical oxidation and polymerization. The two phases of the baking (properly called drying and baking) can be considered distinctly as the temperature at the center of a core is noted with the aid of a thermocouple embedded in the sand (Fig. 2). First the temperature increases to 100C followed by a leveling of drying, then a second more rapid increase in temperature due not only to the heat of convection gases but also to the heat of reactions of the oxygen and the polymerization of the binder which begins at middle of the sand.

All authors agree in recognizing that presence of oxygen is necessary for baking; in an atmosphere of nitrogen and carbonaceous gas the core would develop no strength. On the contrary, a core placed in an

oxygen atmosphere can have its temperature brought to more than 400C (750F) in 2 min with destruction of the binder.2 Baking must then be a controlled operation. For linseed oil, for example, the oxidation occurs slowly at first, increases very much between 175C (345F) and 250C (480F) and breakdown occurs at 260C (500F). The chemical reactions of baking are not well determined and it is difficult to calculate the quantity of oxygen needed; yet Fiegehen⁸ has made measurements on this point from which he finds that for the oxidation one kilogram of linseed oil requires about 2.7 cu meters (95.34 cu ft) of air. One can predict from these results that a slow renewal of air will be sufficient for the heating of a charge of cores in which the linseed oil content is about 1 or 2 per cent.

A systematic study has been made elsewhere on the effect of the introduction of air on the mechanical properties of linseed oil cores;⁴ it would result that the amount of fresh air admitted during the baking has little effect on the values of these properties; the volume of air in the oven proved sufficient to oxidize the binder: on the contrary an appreciable quantity of replacement air would tend to decrease the value of the mechanical properties studied (cohesion, bending, and hardness) and a minor quantity would be necessary (10 to 15 per cent) to give sufficient hardness.*

For those concerned with comparative baking time in an atmosphere of dry air and of steam, tests recently made in the sand laboratories of the Department of Mines of Canada showed that baking in an atmosphere of steam does not decrease the strength of cores after baking but increases the time of baking by about 50 per cent.⁵

We will now show the results of tests which we have made to compare the drying of molds and cores in confined atmospheres and changing atmospheres.

#### II—Comparative Tests on Core Baking in Confined Atmosphere and in Changing Atmosphere

Before giving the results of tests made on an industrial scale, it is interesting to determine what the Com-

^{*} On this last point we are not in agreement; the laboratory tests to which we refer later do not bear this out.

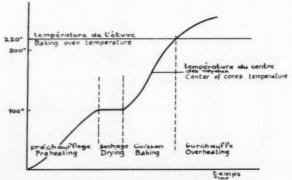


Fig. 2-Baking stages of an oil sand core.

[•] One can consider that the moisture content is affected by the heat transmission on a thickness de which represents the thickness of the imaginary layer in which the moisture content going from a value h to a value O, gives the equivalent variation to that of the moisture content of the mass of the mold during the time dt.

mission compétente du Centre Technique des Industries de la Fonderie which is engaged in confirming a new specification for tests of core binders on test cylinders considers the need for introducing a certain amount of air in laboratory ovens for the correct baking of specimens. Of the numerous tests made in the Dietert oven it has been shown that it would not be necessary to have a completely confined atmosphere permitting any renewal of air because in that case the baking of the core is retarded and the value of maximum cohesion is decreased for a given mixture.

The industrial scale tests which were the primary objective of our study on the baking of cores, have been carried out in an electric oven of current design, with metal-walled heating furnaces, forced convection, of 1 cu meter (35.31 cu ft) volume (effective volume 0.8 cu meter or 28.25 cu ft) and of 7.5 kw power.

#### A-First Test Series

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In a first test series, conducted on a normal load of cores, we were able to compare the heating time for operation with all dampers closed or allowing a continual change of air; the time to reach 250C (480F) was then 2 hr 45 min. to 4 hr, on account of the appreciable stack losses which reduced the heating in the second case. We have even verified that the opening of the door during operation to permit charging, resulted in a drop in temperature such that a half hour of heating was necessary to compensate for it.

The saving realized by confined heating being appreciable, we have undertaken to show directly that the required baking conditions were realized in the oven atmosphere, in spite of the complete closing of the dampers. To do that after having charged a complete set of cores, we connected to the oven a device for continuous aspiration of gases of which the ana-

lysis was made by a recording apparatus; a device for condensation of water vapor was placed in the circuit to allow verification that the humidity of the oven atmosphere was negligible. We have further noted the presence of CO₂ at the start of the binder oxidation, but the CO₂ gas content did not exceed 1 per cent. Finally the direct Orsat analysis allowed verification of the presence of oxygen in excessively large quantities.

Below are two of these analyses, one taken at the beginning, the other at the end of baking:

	CO ₂ , %	O2, %
First Analyses	0.5	20.2
Second Analyses	0.75	20.

One must conclude that the leakages of the oven are largely sufficient to renew the inside air; nevertheless the metal walls were absolutely air tight and the entry of the air could only come from the joint in the door; but the reduced pressure of the warm atmosphere of the oven was sufficient to result in an appreciable aspiration through the joint. With the view to determining our experimental conditions from the point of view of leakage, we have undertaken to measure directly the amount of leakage of our oven.

#### B-Measurement of the Amount of Oven Leakage

(1) Principle—One measures the dilution of a substance introduced in an oven in known quantities. We have chosen to measure the water vapor content obtained in the oven by the evaporation of a certain weight of water.

(2) Experimental Arrangement—We placed in the oven, before placing it in operation, a known weight of water P (Fig. 3). During the operation a weight p was evaporated, measured at the end by weight difference. On the other hand a continuous aspirating

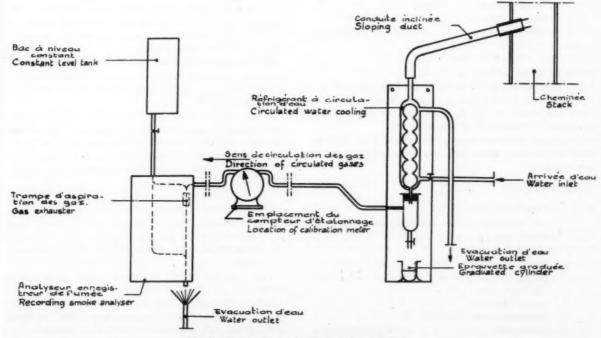


Fig. 3—Device for measuring smoke-moisture.

device (water jet pump) was connected to the oven which we calibrated by means of a gas meter, the test giving the output aspirated per unit time. A condensing apparatus (refrigerant descending in the circulating water) placed in the circuit of the oven outlet, gave the quantity of water condensed in a known time, and therefore permitted the measurement of the water vapor content of the aspirated gas.

#### (3) Results of Measurements

(a) First test-7 hr duration

-Weight of wa-: p = 15.750 kg 875 mols ter vaporized -Water condensed during the entire op- : 200 cm³ 11.1 mols eration -Total value of gas aspirated by the pump : 1,260 liters 56 mols water content of aspirated : _____ lb = 20%gas

Volume of dilution air =4 times the volume of vapor being 875 x 4=3,500 mols in 7 hr, the volume of the oven being 44 mols (1 m³), the replacement of air was 3.456 mols, that is 494 mols hr, representing a volume of 11.115 m³ hr of cold air.

#### (b) Second test

—Weight of wa- ter vaporized	: 2.24 kg	124 mols
-Condensation		1.7 mols
—Total volume aspirated by the pump	: 450 liters	20 mols
	: 8.5% 124 x 91.5	40.00
tion air		1.335 mols

being 30 m³ of cold air in 2½ hr.

Replacement: 29 m³ or 11.6 m³/hr. of cold air. These tests showed that an appreciable volume of fresh air passes by the door joints and the leaking dampers—of the order of 11 m³ per hr.

We have tried to reduce this entering air by cementing the door joints. We then obtained the following results:

#### (c) Third test

-Weight of wa-				
ter vaporized	:	8.570 kg	476	mols
—Condensation			25.5	mols
—Total volume aspirated by				
pump	:	763 liters	34	mols
-Mean water				
vapor content	:	75%		
-Volume of di-		476 x 25		
lution air	:	=	159 mole	ecules
		75		

being 3.6  $m^3$  of cold air for the duration of the test (4 hr, 15 min).

Replacement: 2.6 m³ or 0.6 m³/hr.

#### (d) Fourth test

-Weight of wa-				
ter vaporized	:	11.790 kg	655	mois
-Condensation	:	594.5 cm ³	33	mols
-Total volume				
aspirated by		5 hr, 45 min		
pump		at 8 mol/hr	46	mols
-Mean water				
vapor content	:	72%		
-Volume of di-		655 x 28		
lution air			254.7	mols
		72		
Replace-				
ment air			210.7	mols

being 4.74 m³ of cold air in 53/4 hr or 0.82 m³/hr.

The above values show that the loss to the stack is not negligible in the last two tests.

With all dampers closed, about 11.3 m of cold air pass through the joints of the door resulting in a loss of 800 cal/hr (a mean temperature of 250C (480F) was taken as the ambient temperature of the oven.)

The oven having a power of 7.5 kw, this loss represents about 12½ per cent of the energy consumption. Cementing the door reduces appreciably the volume of air admitted during the oven operation and reduces this loss to 50 cal per hr or 0.7 per cent.

Having thus determined the air tightness and obtained a confined atmosphere (up to a mean water vapor content of 80 per cent) we repeated our tests of drying in a confined atmosphere.

#### C-Drying of Cores in a Confined Atmosphere

We had available a complete charge of cores in the oven and placed in this charge, some test pieces of predetermined strength to permit a concrete measurement of the drying quality.

Six thermocouples, of which four were placed in the oven atmosphere and two in the sand in the experimental cores, were permitted to follow the ther-

mal heating cycle.

When the charging was completed, the oven door was cemented and thus an air tightness (previously determined) was obtained which was far superior to that which can be obtained in industrial practice.

The oven atmosphere was analyzed at the middle of the test by two methods:

 Continuous analysis of CO₂ content by means of a recording apparatus,

—Intermittent analysis of the CO₂ (check) and the combustible gas content (CO, H₂) by means of an Orsat apparatus equipped with a platinum spiral.

Here are the results obtained under these conditions:

When the mean core temperature reaches 225C (437F) (mean ambient temperature: 240C or 464F, heating was stopped and the oven door opened.

The results of the atmosphere analyses are given in

Fig. 4. The CO₂ content reached a maximum of 5.3 per cent, CO 1.5 per cent, and H₂ (or hydrocarbon) 0.4 per cent. This indicates there was always sufficient oxygen remaining (minimum 14 per cent) for the oxidation of the binder. In effect the drying of the charge was satisfactory, the setting of the binder was effective, the cores were hard and the strength tests obtained on test specimens gave a value of the resistance to the traction entirely normal and satisfactory for the materials used.

One can conclude from these results that it is practically impossible to realize, in a recirculating foundry oven, a sufficiently confined atmosphere in order that the drying of cores can no longer be done therein. An ordinary recirculating oven, even completely closed, in general will allow sufficient air to go through its joints to assure oxidation of the binder. If admission of fresh air is allowed in the recirculation circuit, it will be necessary to regulate the entry frequently through small openings. This will avoid a wasting of calories by unnecessary replacement.

It will be interesting from the point of view of thermal efficiency to obtain an air tightness as perfect as possible, particularly along the door joints.

The improvement in efficiency which results from a replacement of a minimum amount of air and by good air tightness, is clearly shown by the thermal balance sheet given in the appendix.

#### III-Study of Drying Molds in Steam Atmosphere

#### A-Theoretical Considerations

Before giving the results of tests we have carried out, we will first make a comparison of the physical properties of air and water vapor likely to be encountered in the process of heat transmission.

	Figures gi	iven by		
		Moser	Erk	Cammerer
	100 C (212 F)	0.020	0.0204	0.0259
Conductivity	200 C (392 F)	0.02558	0.0266	0.0310
(K Cal/M hd°C)	300 C (572 F)	0.0315	0.0331	0.0361

Results given b	y Landolf and Bernstein
Water Vapor	Air

We see first from this table that the thermal conductivity of liquid water

$$[\lambda = 0.49 + 0.1 \frac{t}{100} \text{ (between 0C or 32F and } \\ 90\text{C or 194F]}$$

is about 20 times as great as that of air, that of water vapor, however, is much less than that of dry air. But in a foundry oven the coefficient of conduction

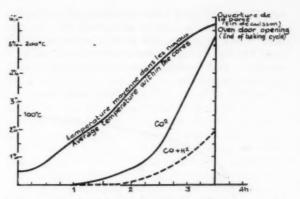


Fig. 4—Atmosphere analysis of core baking oven operating with closed and tight circuit.

does not only enter in the transmission of heat; the heating taking place by convection of the gas, it is necessary to compare the convection coefficients.

Resistance to heat transmission by convection is localized on the surface where a thin layer of fluid adheres which retards the passage of calories by its insulation characteristics. The convection coefficient K is defined by the Newton formula

 $Q = KS (t-\Theta)$ 

giving the quantity of heat Q passing through surface S in a unit time when the temperature differential is  $(t-\Theta)$ .

This coefficient K depends on fluid convection which interferes with the conductivity, but its specific weight, and by the film thickness on the surface (which is also dependent on the fluid viscosity). In addition the coefficient K varies with the regulation of fluid flow and there is a relation between the regulation of the temperature in the stream and that of its speed. In particular, the existence of the thin film which we mentioned above where heat is transmitted by conduction, the temperature, which varies linearly, is related to that of the boundary laminar layer located along the wall where the speed also varies linearly.

The thickness of these two layers are related the equation

$$\delta_t = \delta_d S^{\frac{1}{u}} \qquad \text{where} \qquad$$

 $\delta_t \equiv$  thickness of secondary thermal laminar layer  $\delta_d \equiv$  thickness of secondary dynamic laminar layer

$$S =$$
the Stanton number  $S = \frac{\lambda}{C_{n-1}}$ 

(where v is the dynamic fluid viscosity) and n is an experimental exponent.

For the gases, the Stanton number S is independent of temperature and pressure and varies depending on the complexity of the molecule.

The analogy of the regulation of speed and of temperature can be quite exact in the case of laminar flow along a plane; we know that there exists in this case a boundary layer where the fluid speed variations are localized in the vicinity of the wall and that this layer is located on the interior of a parabola. In the same way the thermal boundary layer is limited by a parabola (Fig. 5), the ordinates of the two parabolas for the same abscissa:

$$OH = X \\ \delta_t = \delta_d S \%$$
 
$$\delta_t \text{ and } \delta_d \text{ such that}$$

Fig. 5—Thermal limit layer in case of laminar flow along a plane.

One sees finally that the coefficient of convection K will vary with the speed of the fluid since when the speed increases the insulating film decreases in thicknesses and breaks which increases the flow of transmitted heat.

Finally for turbulent flow, with an average Reynold's number, along a smooth plate of little roughness, the convection coefficient takes the following form (mean values between 0 and X on the abscissa):

$$K_{o}{}^{x}=~0.036~(rac{\dot{\lambda}}{S_{r}})^{~0.20}~.~(CpP\mu_{o})^{~0.80}$$

where  $\mu_0$  is the speed of flow and p the specific mass of the fluid.

For a flow characterized by a large Reynold's number, and where the wall roughness becomes preponderant, the convection coefficient takes the form

$$K = k \; C_p p N_o$$
 with  $k = \frac{Y}{2}$  ;  $Y = coefficient$  of friction.

The roughness increases the heat exchange.

Let us compare the numerical values of the coefficients for air and water vapor

at 250C or 482F-smooth surface-

$$\begin{split} &\frac{K\left(H_{2}O\right)}{K\left(air\right)} = & \left[\frac{\lambda_{H2O} \cdot S_{air}}{\lambda_{air} \cdot S_{H2O}}\right]^{0.2} \left[\frac{C_{p} \cdot P\left(H_{2}O\right)}{C_{p} \cdot P\left(air\right)}\right]^{0.8} \\ &= 1.12 \end{split}$$

-a rough surface-

$$\frac{K (H_2O)}{K (air)} = \frac{C_p p (H_2O)}{C_p (air)} = 1.16$$

From that comparison we can draw the following conclusions:

—At the same temperature the convection of water vapor is a little higher than that of air, but the difference is slight and should result in little effect. In the same way the quantity of heat stored in a volume V=1 of water vapor  $(C_p \times p)$  is slightly higher than that in the same volume of air.

To return to the problem of heating, one can therefore predict that for the same power consumed, the equilibrium temperature of the gas mass, assumed in a closed circuit, which is made up of the oven atmosphere will be lower in the case of the water vapor than in the case of air. But one can hardly predict that there will be an appreciable difference in the efficiency of heat transfer of the convection fluid in molds.

#### B—Tests on Drying of Molds in an Atmosphere of Superheated Steam

To study this problem in as complete a manner as possible, three types of tests were carried out.

- —laboratory tests on moist sand samples in an electric oven.
- —semi-industrial scale tests on small molds in a 1 cu meter electric oven which has previously been used to study the baking of cores.
- —industrial scale tests on very large molds in a large brick (masonry) oven.

(1) Laboratory Tests—The test apparatus was a small electric oven with revolving hearth and a forced circulation which could function as an open, semiopen or closed circuit.

Tests were conducted on two different types of sand. Each test was made on 16 A.F.S. test cylinders of 50 mm diameter and 50 mm height. They weighed about 170 grams and were weighed exactly. During the time needed for preparation and weighing, the test specimens were placed in a hermetically sealed container to avoid drying.

The oven was heated until the desired uniform temperature was obtained.

In one series of tests, with the view to studying drying in a dry atmosphere, we placed at the air intake of the oven, a container of dehydrated silica gel which absorbed the moisture in the air and we opened a large exit orifice for the air.

In other series of tests with the view to studying drying in moist air, we closed the intake and exit air orifices to provide a closed circuit, and we placed inside the oven a 250 x 180 mm heating plate which was supplied in boiling water (atmosphere of superheated steam).

The 16 test specimens were placed at the same time on the oven trays and during the test the trays were continuously rotating. The air was circulated by the fan and its temperature was maintained by passing it over electrical resistances.

The zero time is the moment when the specimens were placed in the oven. An electric meter placed in the circuit recorded the current consumption at each instant during the test.

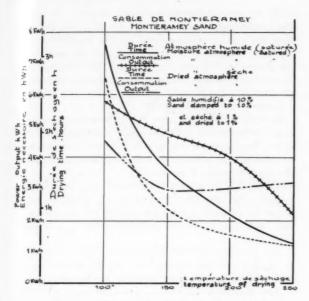


Fig. 6—Drying tests on Montieramey sand under damp and dry conditions.

Each 30 minutes two specimens were removed from the oven and weighed immediately. The authors then computed the per cent of water evaporated.

The temperatures chosen were 100C (212F), 150C (302F), 200C (392F), 250C (482F), and 280C (536F).

The results of these tests are shown in Figs. 6 and 7.

The curves in Fig. 6 show for a first quality sand the time and current consumption needed for drying as a function of temperature, first for a dry atmosphere and then an atmosphere of superheated steam.

One verifies that the drying is faster and requires less energy in a dry atmosphere, yet without this difference being very significant.

One notices on the other hand that the drying time is a hyprebolic function of temperature which was shown by calculations.

The curves in Fig. 7 corresponding to a second quality sand, indicate results comparable to the preceding ones except with a little more scatter.

The drying time increases when one goes from a dry atmosphere (open circuit), to a partially moist atmosphere (closed circuit), then to a moist atmosphere (closed circuit with superheated steam).

The differences in drying time are, however, of little importance and practically non-existent in the temperature range present in foundry ovens (200C or 392F to 300C or 572F). From the first series of tests, therefore, one can conclude that the oven atmosphere has no appreciable effect on the efficiency of drying and the most important factor is temperature.

#### (2) Semi-Industrial Tests-

From these tests it is possible to compare on a larger scale the drying efficiency in warm dry air and an atmosphere of superheated vapor.

To compare the two methods of drying, it was

necessary to hold one of two factors constant: temperature or power consumption. We preferred, as in preceding tests, to hold temperature constant because it is difficult to maintain the same power in both cases; in fact, it is necessary in the case of drying in dry air, to admit a replacement of air and discharge the steam (whence a stack loss) and in the case of drying in a steam atmosphere, to place in the oven with the mold to be dried, a certain mass of water which, while evaporating during heating, will insure an atmosphere containing a high steam content (whence the use of calories to vaporize the water).

To compare the two drying methods we therefore must operate at equal energy consumption, but for like temperature increases obtained, must make manual adjustments for the electrical power consumed.

(a) Experimental Apparatus—The tests in a steam atmosphere were made in a completely closed oven, the door joints being closed with refractory cement. A container was placed inside with ten kilograms of water which evaporated gradually during the heating and the apparatus previously described for measuring water vapor content was placed in position.

The tests with a dry atmosphere were carried out with intake and exit dampers partly open and the water vapor was controlled to remain nil during the test. The two tests were carried out following the same curve of increasing temperature.

Couples were placed in such a way as to permit the measurement of the following:

- —the ambient temperature of the oven (high and
- —the temperature at various depths in the experimental sand mold (6.25, 4.5, 2.5, and 0.5 cm).

The test mold having dimensions of 450 x 355 x 125 mm was made in the same manner for the various tests (same weight, density of ramming, and even

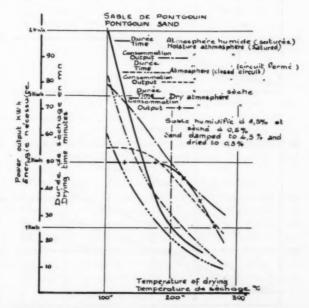


Fig. 7—Drying tests on Pontgouin sand in damp atmosphere and in closed circuit and in dry air.

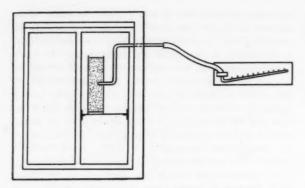


Fig. 8—Pressure measured within the sand during oven operation.

moisture content) and the weight of water lost in drying was determined by weighing before and after.

The weight of water to maintain the steam atmosphere was largely calculated in such a way that the liquid remains to the end of the test and the weight of the water evaporated was determined by difference.

Two effects of drying in a steam atmosphere could be expected:

1. The effect on speed of drying into the mass and on the drying time to the center.

2. The effect on the drying quality and particularly on the chances of crack formation. In an attempt to show the results obtained in this respect, we raised the temperature of the sand (at 0.5 cm). We thought that the cracks being caused by the appearance on the sand surface of a hard crust due to rapid and complete heating of the surface, the decrease in chances of cracking would be accomplished by a more gradual increase in surface temperature with perhaps a leveling off at 100C (212F).

Besides, we thought that the appearance of cracks could be associated with the steam pressure inside the sand. We therefore arranged a series of tests planned to measure the pressure in a mold during drying and

the corresponding temperatures.

To do this we placed in the sand, at a fixed depth, a glass tube, the other end of which extended from the oven through a hole for taking the temperature and was joined to an inclined water manometer (Fig. 9).

(b) Test Results—The results are particularly clear as regards the time necessary for drying to the center. (See the thermal balance sheet in the appendix).

The dropping of the temperature level to 100C (212F) at a depth of 6.25 cm was accomplished for the cycle of heating adopted, at the end of 51/4 hr, regardless of the atmosphere where drying occurs (Fig. 9).

These tests allowed us also to verify the formula we suggested for the calculation of heating time and depth:

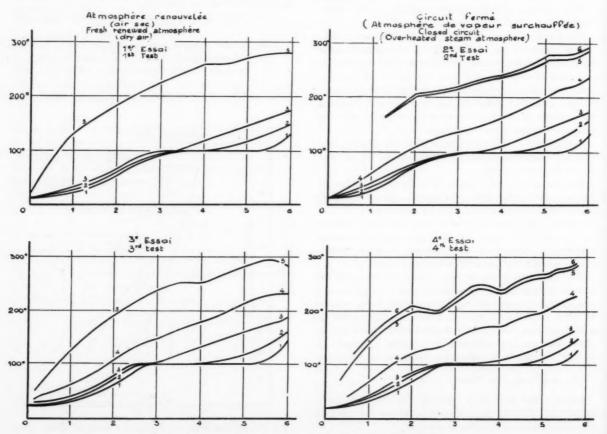


Fig. 9-Sand temperatures vs depth during drying of test mold.

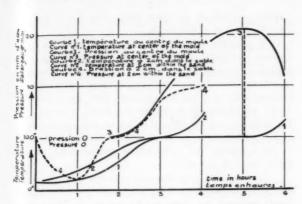


Fig. 10—Steam pressure within the sand vs depth during drying of test mold.

$$t = \frac{e^2}{2m} \cdot \frac{h \ L \ \tilde{\omega}}{(T\text{-}100)} \cdot \left(\frac{m}{\lambda} + \frac{2}{Ke}\right)$$

in which

e = 6.25 cm

h = 7.9%

 $\tilde{\omega} = 1.89 \text{ kg/dm}^3$ 

T = 260C (500F)

 $\lambda = 0.5 \text{ cal/m/h/}^{\circ}\text{C}$ 

L = 540 cal

m = 0.62

 $K = 9 \text{ cal/M}^2/h/^0C$ 

which gives for t a time of  $4\frac{3}{4}$  hr, very close to that determined experimentally.

For the above calculations we have adopted a value of 9 for the coefficient K resulting from the application of the formula

$$K = \frac{Y}{2} CpPN_o$$

proposed earlier for the coefficient for rough violent flow, with the given values as follows:

Y = 0.015

Cp = 0.245 cal/kg (air at 250C or 482F)

 $p = 0.675 \text{ Kg/M}^3$ 

 $\mu_0 = 2$ m/sec = 7200 m/hr

As for the coefficient m we have calculated it taking into account the values found for the different situations of the thermal balance sheet in the appendix which gives the number less than 0.62.

Besides the preponderant role of temperature, the proposed formula above the heating time shows again the importance of decreasing of the convection coefficient *K* to a value which increases the thickness of sand dried.

Consequently, if we increase K in increasing the speed of circulation (use of recirculation) the effect obtained will be less sensitive the thicker the mold. The transmission of heat is then retarded especially by the conductivity at the interior of the sand mass. The principle of recirculation is especially interesting

for shallow drying, that is, for cores and small molds. These considerations are clearly confirmed by our tests of drying of large molds in the foundry.6

Regarding the quality of drying the authors have not been more successful in showing improvement; the increase in temperature at 0.5 cm depth is the same in a steam atmosphere as a dry one and shows no more leveling at 100C (212F) in one case than another. Curve 4 (temperature in the sand at 0.5 cm) follows Curve 5 (ambient temperature) with simply a difference, the same in both cases, of the order of 80C or 176F. Accordingly, curves 2 and 3 (temperature at 4.5 and 2.5 cm. in the sand) can be a little divergent but this is not regular and can be explained by the inexactness in the placing of the tips of the thermocouples in the sand.

The results of pressure measurements are given in Fig. 10. The pressure at the middle of the mold increases with temperature and continues to increase during the leveling of drying at 100C or 212F. Then when the leveling is ended, the temperature rises about 100C or 212F (end of vaporization of moisture), the pressure decreases in the same time as the residual water vapor is eliminated. We have then another procedure for following the deep drying by the measurement of pressure. When the maximum of the pressure curve is passed, the drying is complete and the water entirely vaporized.

The value of maximum pressure in the course of our tests remains between 13.5 and 21 mm of water at a depth of 6 cm of sand rammed hard (moisture content of the sand 6 to 7 per cent). We have not noticed any variation of this maximum value particularly as a result of the nature of the oven atmosphere.

One conceives, in effect, that the internal pressure created as a result of a lack of permeability of the sand by moisture vaporization will follow the trend of temperature increase and will be influenced by the ramming density, the permeability, and the water content of the sand. But the phenomena we seem to note as indifferent to the nature of the atmosphere in which the drying is done and it is well that experience confirmed this.

Besides, we have accomplished a test for the measurement of pressure not only at the center of the mold but closer to the surface at 2 cm depth. This test has shown the existence at the start of drying of a phase of very marked reduced pressure (see Fig. 9), where one can see an experimental confirmation of the theoretical factors explained more above on skin drying.

The decrease in pressure verified would be explained by the driving off of vaporized moisture near the center of the mold. Then the pressure deviates in a positive direction, increases rapidly (driving off of moisture near the surface) and continues to increase even though the leveling off of drying has reached point where it is considered slow (driving off of moisture at center of the mold); the end of drying

^{*} This procedure has already been used by M. Blanchard in Paris for following the drying of furnace linings.

is indicated by a very marked inflection of the curve.

In short, the results of this second series of tests confirms those laboratory tests; at normal oven temperatures the nature of fluid convection has no appreciable effect on drying speed. No appreciable differences were found to favor a superheated steam atmosphere, in the range of temperatures and pressures in the same mass of sand likely to cause a more marked tendency to crack formation.

#### (3) Industrial Tests

These tests were carried out in a foundry during the operation of regulated ovens of large volume, functioning with a normal charge of molds but allowing in addition a test mold of 600 mm thickness, the other dimensions being greater than the depth; this mold consisted of a flask entirely filled with sand of normal moisture and density, in which were placed at various depths, near the middle, thermocouples to follow the progress of drying by temperature measurements.

These tests are not as systematic as the preceding because certain factors affecting the mechanism of drying can vary from test to test: moisture content, sand quality, and sand density.

Nevertheless, they are sufficiently similar to permit comparing in the different cases, the order of magni-

tude of the drying time.

The table below gives the drying time up to a depth of 10 cm for large test molds according to the oven type and temperature.

Oven with superheated steam at 260C (500F) 12 hr Oven without recirculation at 400C (752F) 7 hr Coke oven with circulation at 400C (752F) 8 hr

The temperatures indicated correspond to the maximum obtainable during normal functioning of the types of ovens in question. It is noted that the oven with superheated steam which cannot operate above 260C (500F) dries less quickly than the other two. The lack of temperature is not therefore compensated for by a convection power of the superheated steam, appreciably greater than that of the atmosphere of ordinary ovens.

On the contrary, the appearance of the molds was much better after drying in a steam atmosphere: the absence of cracks was particularly noticeable.

On the other hand, in an oven with superheated steam, the temperature distribution was more uniform, and resulted in uniform drying to the same depth of the entire load.

These tests confirm the two preceding test series made in the laboratory and at the semi-industrial level.

The nature of the warm fluid has no important influence on the efficiency of mold drying, the essential factor being the temperature.

Although we have not made clear the variations in the range of the temperatures and pressures in the mass of sand as one heats it in air or steam, it seems that the steam atmosphere is favorable to mold quality, especially that which is evident by the absence of cracks. This fact which we have been able to verify has already been pointed out.7

It is possible that the quality of drying and the absence of cracks, which were observed in ovens with superheated steam, is not due to the steam atmosphere but to the slow rate of temperature rise followed by a maintenance of a low temperature (less than 250C or 482F) which results in, on the one hand, a slow transfer of heat, and on the other hand, the greater heat capacity of a cubic meter of steam than a cubic meter of air lowering the equilibrium temperature of the warm fluid in the oven.

Too rapid a rise in temperature results in effect in the formation of a dry crust of low porosity which tends to crack to open up a passage for the steam resulting from the vaporization of the moisture of the deeper layers.*

In the exchanger ovens drying in an atmosphere of superheated steam, the temperature distribution is very good because the mixing of the warm fluid can be done in a completely closed circuit (the excess steam is removed by a valve or by the joints in the door) in the ordinary ovens, the temperature distribution is not as good because the recirculation circuit is not entirely closed since the combustion chamber produces an excess of warm gas when the chimney allows excess fumes and corresponding steam to escape.

On the other hand, the thermal efficiency of heat utilization is not as good in the exchanger oven because the overall efficiency of the oven is affected by the efficiency of the exchanger which cannot be 100

per cent.

At any rate, in all our tests we have never concluded that the steam atmosphere gives a heating rate and thermal efficiency higher than warm air contrary to that which is indicated in a recent article concerning the drying of granular materials;8 it is true that these tests have been made very differently and that the sand tested did not have the characteristics of a foundry sand and were not prepared in the same manner.

#### Conclusion

The results of the writers' tests show that the influence of the nature of the atmosphere, in which molds or cores are heated, is secondary in comparison with the essential factor which is the temperature of that atmosphere.

With regard to the baking of cores, it is observed that the oxygen necessary for the oxidation of the binder is found in always sufficient quantity in the atmosphere of an ordinary oven, thanks to the intakes of air through the joints.

^{*}We do not pretend in this study to be able to give the complete explanation of the phenomena of crack formation, which depends on a number of factors, the influence of which we have not studied, such as the physical-chemical nature of sand. We simply wished to show that the factor of atmosphere had only a slight effect as compared with temperature. The problem of crack formation has been studied in a more general manner and in collaboration with M. Nicolas, Chief of the Primary Materials Department of the Foundry Industry Technical Center. M. Nicolas will soon be in a position to report the results of his work on this subject.

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With regard to drying of molds, it is observed that the atmosphere of superheated steam is not more active than the hot air mixed with fumes, but that the steam produces a regular and gentle drying, favorable to a good appearance of the molds and reducing the formation of cracks,

To combine quality of heating and economy of calories, it is of interest to have the oven quite tight, to limit to the minimum the entry of cool air to be introduced in the recirculation circuit and to choose a thermal cycle giving a slow rise in temperature.

The tests mentioned in this thesis were made at the testing station of the Technical Center of Foundry Industries. The authors thank their General Director, Mr. Le Thomas, for the interest he has kindly shown in this work, by authorizing the authors to publish these results.

#### **APPENDIX**

#### Thermal Balance Sheet of Heating

The measurements made (amounts, temperature rises) permitted evaluation of several items of a thermal balance sheet.

#### (1) Balance Sheet for Mold Drying

The authors made two heats on maximum charges and compared the thermal balance sheet for these two operations for the same fixed time of 7 hr.

(a) First heating — all dampers closed — partial tightness.

Charge—356.85 kg for 0.8 cu meters of volume Moist sand—281.65 kg

Flask—75.2 kg
Temperature attained at end of 7 hr heating

in the atmosphere of the oven, 260C (500F); drying incomplete.

(b) Second Heating all dampers closed door

(b) Second Heating—all dampers closed—door joints cemented—tightness as good as possible (stack losses at a minimum)—

Charge—355 kg Moist sand—2'

Moist sand—279.8 kg Flask—75.2 kg

Temperature attained at end of 7 hr heating in the atmosphere of the oven, 290C (554F); drying completed at center.

#### Comparison of Balance Sheets

Energy Consumed (7 hr heating at 7.5 kwh =  $860 \times 7.5 \times 7$ = 45,150 cal)

	Heatin Atmos Parti Confi	phere ally	Heating in Totally Confined Atmosphere	
		Cal		Cal
<b>Evaporation of Moisture</b>	14.07 kg	9,005	18.4 kg	11,776
Heating of Sand	135 C (275 F Mean Temp 266 kg		155 C (310 Mean Te 261.4 kg	
Heating of Flask Heating of Drawers	260 C (500 F) Mean Temp		290 C (554 Mean Te	,
Heating of Drawers	671		7	25
Total Heat Utilized	20,462	Cal	24,020	6 Cal
Efficiency	45.39	6	53.5	2%

The tightness at the door joints was therefore sufficient to increase the efficiency of heat utilization by 7.9 per cent.

#### (2) Balance Sheet for Drying of Cores

(a) Balance sheet of heating 48 kg of core for 0.8 cu meter—starting cold—operating time 3 hr—efficiency of utilization 24 per cent.

(b) Oven of same charge with starting warm—operating time 3 hr to 2 hr—efficiency of utilization of heat 33.2 per cent.

Heating realized with greater charge density.

(c) 110 kg of core for a useful volume of 1 cu meter —starting cold—efficiency of utilization of heat 31.5 per cent.

(d) 75 kg of core for 1 cu meter—starting warm—34.1 per cent.

#### Remarks

1. For the balance sheet calculations above, we have included in the charge, the metal shelves of the oven which are charged cold with the cores. The tiered carriage which is not removed from the oven is not counted in the charge.

2. It was estimated that the contribution of heat from the exothermic reaction of the binder oxidation

#### Details of Balance Sheet Above

	Energy Consumed	Н	E A T U	TILIZE	D	Efficiency
	Consumed	Evap. of Moisture	Heating Sand	Heating Racks	. Total cal	%
48 kg—0.8m³	3 hr,	5.1%	8.9%			
$(60 \text{ kg/m}^3)$	7.5 kwh	992 cal	1717 cal	10%		
Start Cold	19,350 cal			1953 cal	4662	24
60 kg/m ³	2 hr,	1.2%	15.8%	16.2%		
Start Warm	7.5 kwh	150 cal	2040 cal	2090 cal	4280	33.2
	12,900 cal					
110 kg/m ⁸	31/2 hr.	1.9%	14.6%	15%		
Start Cold	12 kwh	704 cal	5280 cal	5408 cal	11,392	31.5
	36,120 cal					
75 kg/m ³	2 hr,	2.3%	16.4%	15.4%		
Start Warm	12 kwh	480 cal	3375 cal	3185 cal	7040	34.1
	20,640 cal					

compensated for the loss of heat by the evaporation of the binder.

3. The temperatures used for the calculation of the heating of the sand are the lowest temperatures evaluated according to the curve of temperature variation

in the depth of the sand.

4. Finally, all of the balance sheets above have been set up for ovens operating with all dampers closed. If one had operated while partially opening the dampers, the heating time would have been appreciably increased-for example, for the first balance sheet of core baking the heating time would have gone from 3 to 4 hr at a minimum and the efficiency from 24 per cent to 18 per cent.

#### References

1. "The Influence of Moisture on the Thermal Conductivity of Granular Materials," Journal of Research of C.N.R.S., no. 8. 2. E. C. Troy, "Factors Influencing the Baking of Cores,"

Foundry, June 1948, pp. 92, 95, 252, 254-6.
3. E. G. Fiegehen, "The Drying of Cores and Molds,"

Foundry Trade Journal, vol. 51, pp. 21-23 (1934).
4. S. H. C. Forelund, "The Effect of the Introduction of Fresh Air and the Volume of Core on the Mechanical Properties of Linseed Oil Cores," *Gjuteriet*, vol. 36, no. 6, pp. 97-104 (1945). 5. A. E. Murton, H. H. Fairfield and B. Richardson, "Core Oil Evaluation Method," A.F.S. Transactions, vol. 59, pp. 276-

6. G. Ulmer, "Experimental Study of the Efficiency of Recipculation in Foundry Ovens," Twenty-fourth Congress of the Foundry Technical Association, Paris.

7. George Wright and J. M. Sampson, "Humid Oven Atmosphere Eliminates Surface Cracks," The Foundry, vol. 55, no. 16,

Aug. 15, 1927, pp. 644-646.

8. L. Wenzel and R. R. White, "Drying Granular Solids in Superheated Steam," Industrial and Engineering Chemistry, Aug. 1951, pp. 1829-1837.

#### DISCUSSION

Chairman: O. J. MYERS, Archer-Daniels-Midland Co., Minne

Recorder: F. S. Brewster, Harry W. Dietert Co., Detroit.

CHAIRMAN MYERS: The general experience in this country does not agree with the authors' statement that cores can be baked efficiently in a closed circuit oven. To summarize, some of the important points the author mentioned are: (1) The temperature of baking is the prime factor, (2) Cores can be baked in super-heated steam, (3) Heat travels through a core via the binder rather than from grain to grain, (4) Drying molds in super-heated steam is slower than in air but seems to eliminate cracking, (5) Sufficient oxygen is always present for the oxidation of the oils even in a closed circuit oven:

How do the authors measure humidity in an oven?

GEORGES ULMER: By condensing the moisture from the quantity of oven air as measured by a flow meter and then weighing the water.

### EFFECTS OF MOLD MATERIALS ON LEAK TIGHTNESS AND MECHANICAL PROPERTIES OF 85-5-5-5 AND 81-3-7-9 ALLOY CASTINGS

By

J. G. Kura* and L. W. Eastwood**

#### Introduction

The effects of mold materials on the quality of copper-base alloy castings are very important. Mold materials may be responsible, at least in part, for many casting defects, including buckles or rat tails, washes or cuts, rough surfaces, metal penetrations, and blowholes formed by the entrapment of mold gases in the metal in the casting. The present investigation, however, is concerned with the effect of mold material on the absorption of gas by the melt poured into the mold cavity. This paper constitutes a progress report of a research project at Battelle Memorial Institute for the Brass and Bronze Ingot Institute, and the conclusions are not final.

The effect of mold materials will be evaluated primarily by measurement of leak tightness, density, and mechanical properties. By these methods it is possible to measure the relative amounts of gas absorption from the mold because the gas which is absorbed and later evolved during solidification forms unsoundness.¹ This unsoundness not only reduces the density and mechanical properties, but also increases leakage. Although a relationship exists among soundness, density, and mechanical properties, apparent anomalies also exist.

A previous article,² based on work conducted for the Brass and Bronze Ingot Institute, showed quite clearly that a mold which could be baked at 1600 F produced castings with the greatest degree of soundness and, consequently, the highest density and highest mechanical properties. Test-bar castings of 85-5-5-5 alloy made in this mold material contained approximately zero per cent voids. On the other hand, a synthetic green sand mold, prepared by bonding clean silica sand with 6.5 per cent fireclay and 2 per cent western bentonite with about 3 per cent water, produced 1.3 per cent voids with the same metal. A natural green sand mold produced 1.8 per cent voids on the average, whereas the synthetic sand baked at 450

F produced 2.2 per cent voids. These percentages are average values for a large number of heats and test bars. Although the synthetic green sand ranked second with respect to freedom from voids, it produced the poorest tensile properties because localized unsoundness produced shoulder fractures. At a result, a synthetic sand with or without baking at 450 F produced markedly lower properties than the natural green sand.³ In turn, the properties of test bars made in the natural green sand were lower than those of test bars made in the refractory mold baked at 1600 F.

#### **Experimental Procedure**

The general experimental procedure included the preparation of molds of the various materials to be studied. The castings poured consisted of test bars and leak-test cylinders. Melts of 85-5-5-5 and 81-3-7-9 alloys, prepared in various ways to provide melts of either high gas content or low gas content, were poured at one or more pouring temperatures. More detailed information on the various phases of the experimental procedure is presented in the following pages.

Mold Materials: Five different mold materials were employed. A description of their compositions and properties is as follows:

Calcined-Clay Mix: This mixture consisted of the following: 84.3 per cent calcined-clay particles; 4.9 per cent ammonium acid phosphate; 1.4 per cent 150-mesh magnesia; 9.4 per cent water.

The mix was mulled dry for 1 min and then mulled wet for 2 min. The mold was made and air dried for 15 hr, oven baked at 1600 F for 1 hr, and cooled to and held at 400 F prior to being placed on the floor. The mold temperature was about 300 F at the time of pouring. The A.F.S. green permeability of this mix was 52, and its dry permeability was also 52. It had a green compressive strength of 5 psi and a dry compressive strength of 160 psi.

Coated Sand: Silica sand particles coated with an organic material were mixed with 3.7 per cent western bentonite and 4.5 per cent water. The A.F.S. green permeability was about 50, and the green compressive strength was 7 psi.

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^{**} Formerly Supervisor at Battelle Memorial Institute, now Assistant Director of Research at Kaiser Aluminum & Chemical Corporation, Spokane, Washington.

Green Synthetic Sand: The green synthetic sand consisted of 92.2 per cent southern silica sand, 3.8 per cent western bentonite, and 4 per cent water. The A.F.S. Grain Fineness Number of the silica sand was 72. The mix was mulled dry for 3 min and the water was added to temper to produce a green compressive strength of 8 psi. A.F.S. green permeability was 58.

Baked Synthetic Sand: This mixture was the same as the green synthetic sand. However, the molds were baked 15 hr at 450 F and then cooled to room temperature just prior to the time they were poured.

Natural Green Sand Mix: This sand was a natural, clay-bonded Tennessee sand tempered with 6 per cent water. It was mulled wet for 4 min and had A.F.S. green permeability of 46, and a green compressive strength of about 9 psi.

Test Castings: Only two different types of test castings were poured. One type was a test bar similar to Fig. 3 of ASTM B208-49T. The other was a set of four leak-test cylinders on a plate, as illustrated by Fig. 1. These cylinders were machined all over, inside and out, to the final dimensions shown by Fig. 2.

Metal Composition: As indicated previously, 85-5-5-5 and 81-3-7-9 alloys were investigated. Each heat of the alloys was analyzed and all were within the ASTM specification limits for these materials. Typical analysis of each alloy is as follows:

			Co	mposit	ion, pe	r cent		
Alloy	Cu	Sn	Pb	Zn	Fe	Sb	Ni	P
85-5-5-5	84.5	4.7	5.6	4.29	0.10	0.15	0.65	0.015
81-3-7-9	80.5	3.1	6.7	8.9	0.20	0.20	0.40	0.015

Melting Practice: Duplicate melts of each alloy were prepared by the following general procedures:

1. A good melting practice which involved the use of a high-frequency induction furnace equipped with a silica crucible. Ingot metal was charged as the previously charged ingots melted down. After heating to the desired temperature of about 125 F above the highest pouring temperature, the melt was skimmed prior to transferring it to a preheated pouring ladle. Zinc additions of approximately 11/4 per cent of the charge were made to compensate for zinc loss in either alloy, after which 2 oz of 15 per cent phosphor copper per 100 lb of melt was also added to the ladle during the transfer. No fluxes or glass covers were employed during the melting operation. The melting conditions were strongly oxidizing and the melts produced were of high quality, i.e., they produced highdensity castings and high test bar properties.

2. The same procedure as described in Method No. 1 was employed, except that the melt was kept covered with a mixture of glass with sufficient borax added to maintain proper fluidity. This glass cover was maintained on the melt during the entire melting operation, and was transferred to the pouring ladle to prevent oxidation prior to casting. Zinc and phosphorcopper additions were made as described for Method No. 1, except that the amount of zinc added was less. This melting procedure invariably produced melts of high gas content, because the melt did not contact the oxidizing atmosphere. The high-gas-content melts produced low density and low tensile properties.

3. A gas-fired crucible tilting furnace was used in a manner quite similar to that described for Method No. 1. An Orsat apparatus was used and the furnace was operated to maintain approximately 0.5 per cent oxygen in the flue gases. In one or two instances, a reducing flame was also used for making part or all of

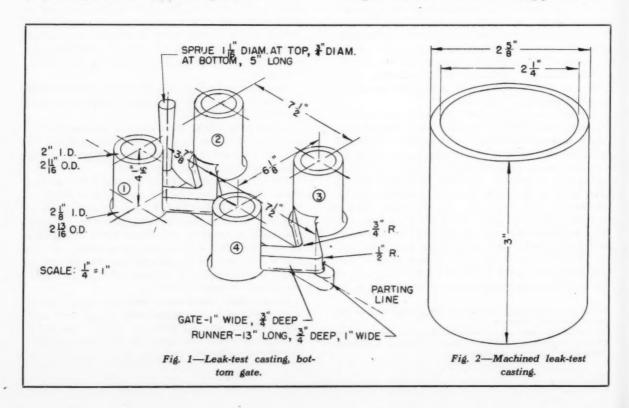


TABLE 1-AVERAGE TENSILE PROPERTIES AND DENSITY OF TENSILE BARS AND LEAK-TEST CYLINDERS OF 85-5-5-5 AL-10Y PREPARED IN HIGH-FREQUENCY INDUCTION FURNACE AND IN GAS-FIRED TILTING FURNACE AND CAST IN VARIOUS MOLD MATERIALS

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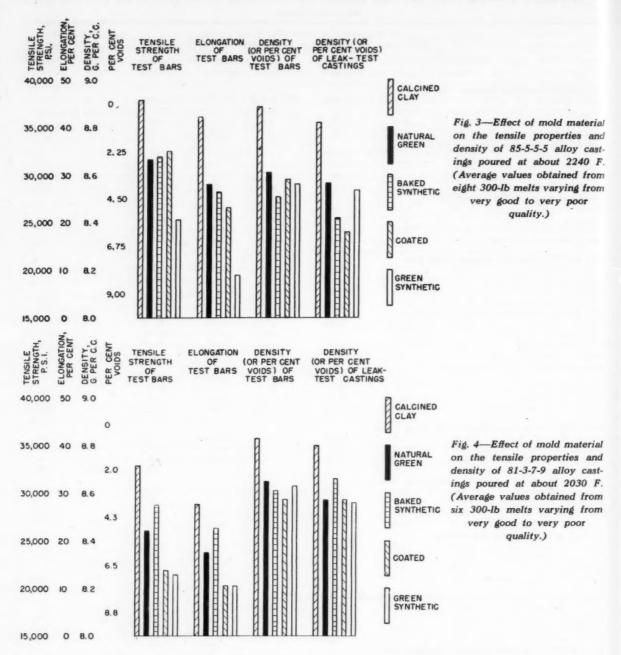
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			r poured at ab Casting poured a			25)		at about (+45,-15
	Calcined Clay	Natural Green	Baked Synthetic	Coated Sand	Green Synthetic	Avg. for 3 Molds*	Coated Sand	Natural Green
Heat A4726—Induction Fu	rnace—No	Cover				-		
Tensile strength, psi	_	35,600	35,100	34,600	24,400	35,100	_	
Flongation, % in 2 in.	-	38.0	32.4	28.4	9.3	32.8	-	_
Density of test bar	_	8.65	8.69	8.65	8.72	8.66	-	
Density of leak-test cylinder		8.64	8.72	8.48	8.60	8.61		8.88
Heat A4728—Induction Fu	37,700	Cover	9= 400	- *		95 400		
Tensile strength, psi Elongation, % in 2 in.	39.0		35,400 35.3		_	35,400 35.3	_	
Density of test bar	8.87	_	8.68		_	8.68	_	_
Density of leak-test cylinder	8.79	_	8.58	8.73	8.66	8.65	8.90	8.72
Average of Heats A4726 a								
Tensile -strength, psi	37,700	35,600	35,200	34,600	24,400	35,200	_	-
Elongation, % in 2 in.	39.0	38.0	33.8	28.4	9.3	33.5	-	_
Density of test bar	8.87	8.65	8.68	8.65	8.72	8.66	_	_
Density of leak-test cylinder	8.79	8.64	8.65	8.60	8.63	8.63	8.90	8.80
Heat A4727—Induction Fu	rnace—Glas	s Cover						
Tensile strength, psi	37,500	-	33,400	_	23,800	33,400	7	-
Elongation, % in 2 in.	41.3		31.8	_	9.0	31.8	-	
Density of test bar	8.90	_	8.62	_	8.69	8.62		
Density of leak-test cylinder	8.78		8.53	_	8.60	8.53	8.81	8.72
Heat A4729—Induction Fu			20 400	94 900	07 400	94.000		
Tensile strength, psi	37,200	35,400	32,400	34,300	27,400	34,000		-
Elongation, % in 2 in. Density of test bar	45.7 8.88	34.2 8.57	28.2 8.54	30.3 8.63	15.3 8.66	30.9 8.58		
Density of leak-test cylinder		8.72	8.37	8.32	8.58	8.47	8.86	8.80
verage of Heats A4727 ar		0.74	0.01	0.04	0.00			
Tensile strength, psi	37,400	35,400	32,900	34,300	26,200	33,800	-	-
longation, % in 2 in.	43.5	34.2	30.4	30.3	13.2	31.3	_	
Density of test bar	8.89	8.57	8.58	8.63	8.68	8.59	_	-
Density of leak-test cylinder	8.77	8.72	8.45	8.32	8.68	8.48	8.88	8.76
Heat A4950—Gas-Fired Fu	rnace—Oxi	dizing Flame						
Tensile strength, psi	38,600	32,400	31,200	31,900	17,600	31,800	-	-
longation, % in 2 in.	45.4	27.2	21.0	20.6	2.6	22.9	_	-
Density of test bar	8.90	8.69	8.39	8.72	8.45	8.60	0.00	0.05
Density of leak-test cylinder	8.87	8.64	8.09	8.48	-	8.40	8.82	8.85
leat A4952—Gas-Fired Fur Tensile strength, psi	38,800	33,000	32,500	31,600		32,400	_	
longation, % in 2 in.	41.4	21.1	23.8	21.0	_	22.3	-	_
Density of test bar	8.89	8.61	8.60	8.49	8.55	8.57	_	_
ensity of leak-test cylinder	8.82	8.49	8.62	8.45	8.55	8.52	8.77	8.76
verage of Heats A4950 an	d A4952							
ensile strength, psi	38,700	32,700	32,100	31,800	17,600	32,200	_	
longation, % in 2 in.	43.0	24.2	22.9	20.8	2.6	22.6	_	****
ensity of test bar	8.90	8.65	8.50	8.60	8.50	8.58		-
ensity of leak-test cylinder	8.84	8.56	8.36	8.46	8.55	8.46	8.80	8.80
leat A4951—Gas-Fired Fu		dixing Follows	ed by Reducing	Flame				
ensile strength, psi	37,800	32,500	31,500	31,000	20.800	31,700	_	_
longation, % in 2 in.	42.2	25.7	26.9	19.3	6.5	24.0	-	-
ensity of test bar	8.90	8.62	8.50	8.43	8.55	8.52	0.74	0.00
ensity of leak-test cylinder	8.90	8.54	8.43	8.02	8.37	8.33	8.74	8.39
eat A4933—Gas-Fired Fur	пасе—кеп		23,500		20,400	27,200		
ensile strength, psi longation, % in 2 in.	_	31,000 . 22.6	13.9		7.9	18.2	_	-
ensity of test bar	_	8.51	8.06	_	8.27	8.29	_	_
ensity of leak-test cylinder	_	8.38	8.01	8.01	8.45	8.14	_	8.87
verage of Heats A4951 and								
ensile strength, psi	37,800	31,750	27,500	31,000	20,700	29,900	-	-
longation, % in 2 in.	42.2	24.2	20.4	19.3	7.0	21.7	-	
ensity of test bar	8.90	8.56	8.28	8.43	8.46	8.42		
ensity of leak-test cylinder	8.90	8.46	8.23	8.02	8.41	8.23	8.74	8.58
verage of All Heats								
ensile strength, psi	37,800	31,600	31,900	32,500	25,200	32,400	_	_
ongation, % in 2 in.	42.3	28.1	26.4	23.2	9.1	26.5	-	_
ensity of test bar	8.89	8.61	8.51	8.58	8.56	8.56	0.00	-
ensity of leak-test cylinder	8.82	8.57	8.42	8.36	8.54	8.44	8.82	8.75

sand, synthetic baked sand, and coated sands only. Data on other two mold materials not considered to be representative of usual in many of the columns. foundry practice.



the melt, as indicated by the explanatory data in Table 1.

Pouring Temperatures: The 85-5-5-5 alloy and 81-3-7-9 alloy melts were poured into test-bar and into leak-test molds at the following pouring temperatures in degrees F:

					I	.eak-T	est Mo	olds	
	(A	-Bar M II 5 M Materia	old .	All 5 Mold Materials			Natural Green Sand & Coated Sand Only		
Alloy	Max.	Avg.	Min.	Max.	Avg.	Min.	Max.	Avg.	Min.
85-5-5-5		2245	2215	2270	2240	2205	2175	2130	2115
81-3-7-9	2060	2035	2020	2060	2030	2005	2150	2140	2110

Testing: The tensile-test bars were sandblasted and

machined to standard 0.505-in. threaded test bars with a 2-in. gage section, the densities determined, and then tensile strength and per cent elongation measured. The leak-test castings were sandblasted, and machined inside and out to the dimensions illustrated by Fig. 2 and the densities determined. Then the machined leak-test castings were leak tested up to 1300 psi. This was accomplished by connecting a tank of nitrogen to the interior of the casting to apply internal pressure while the casting was submerged in water. The pressure at which leaks were first noted was determined. After this operation, representative castings, 85-5-5-5 alloy only, were radiographed and examined metallographically to determine the cause of leakage.

Table 2—Average Tensile Properties and Density of Tensile Bars and Leak-Test Cylinders of 81-3-7-9 Altoy Melts Prepared in High-Frequency Induction Furnace and in Gas-Fired Tilting Furnace and Cast in Various Mold Materials

	Test bar poured at about 2035 F (+25, -15) Leak-Test Casting at about 2030 F (+30, -25)						Poured at about 2140 F (+10, -30	
	Calcined Clay	Natural Green	Baked Synthetic	Coated Sand	Green Synthetic	Avg. for 3 Molds*	Coated Sand	Natural Green
Heat A5893—Induction Fu	rnace—No	Cover						
Tensile strength, psi	34,800	30,100	33,800	27,600	27,600	31,100	_	
Elongation, % in 2 in.	33.4	21.5	32.5	16.0	16.0	24.8	_	_
Density of tensile bar	8.88	8.77	8.82	8.81	8.79	8.81		
Density of leak-test cylinder		8.73	8.81	8.73	8.79	8.77	8.52	8.54
Heat A5991—Induction Fu			0.01	0.10	0.75	0.17	0.02	0.00
Tensile strength, psi	34,400	32,600	34,000	25,600	28,800	30,700		
Elongation, % in 2 in.	31.9	25.8	30.0	12.5	16.5	22.8		
Density of tensile bar	8.88	8.87	8.82	8.84	8.84	8.84		_
Density of leak-test cylinder		8.78	8.84	8.76	8.78	8.79		_
Average of Heats A5893 a		0.70	0.01	0.70	0.70	0.75	_	
Tensile strength, psi	34,600	31,400	33,900	26,300	28,400	30,900	-	_
Elongation, % in 2 in.	32.6	23.6	31.2	13.7	16.3	23.7	-	
Density of tensile bar	8.88	8.82	8.82	8.82	8.82	8.82		_
Density of leak-test cylinder		8.76	8.82	8.74	8.78	8.78	8.49	8.53
						0.70	0.10	
Heat A5922**—Gas-Fired	33,900	deducing, Follo	owed by Oxidis	ring Flame				
Tensile strength, psi		_		_		_		
Elongation, % in 2 in.	29.5 8.82	_	_	_	_	-		_
Density of tensile bar		8.50		_	_	8.50	8.38	_
Density of leak-test cylinder			_	_	-	6.30	0.30	_
Heat A5923—Gas-Fired Fu			00 000	00.000	00 000	00 000		
Tensile strength, psi	33,800	31,500	29,200	23,900	20,900	29,060	-	-
Elongation, % in 2 in.	30.0	22.5	21.5	10.0	9.0	19.6	_	_
Density of tensile bar	8.84	8.66	8.59	8.45	8.54	8.57	P 90	8.43
Density of leak-test cylinder	8.85	8.62	8.72	8.66	8.56	8.67	8.32	0.43
Average of Heats A5922 and		01 500	00.000	00.000	00.000	00.000		
l'ensile strength, psi	33.800	31,500	29,200	23,900	20,900	29,060		-
Elongation, % in 2 in.	29.8	22.5	21.5	10.0	9.0	19.6	_	_
Density of tensile bar	8.83	8.66	8.59	8.45	8.54	8.57	0.95	0.49
Density of leak-test cylinder	8.82	8.56	8.72	8.66	8.56	8.58	8.35	8.43
deat A5892—Induction Fur	rnace—Glass							
l'ensile strength, psi	31,800	22,000	25,900	19,000	18,700	23,200	_	-
longation, % in 2 in.	24.4	12.0	16.5	8.0	7.8	13.2	_	_
Density of tensile bar	8.84	8.56	8.55	8.53	8.52	8.55	_	_
Density of leak-test cylinder		8.45	8.56	8.51	8.35	8.51	8.22	8.30
leat A5990—Induction Fu	rnace—Glas	s Cover						
Tensile strength, psi	28,800	14,400	21,000	14,400	11,900	15,700	_	
longation, % in 2 in.	18.4	5.0	13.0	6.5	3.8	7.2	_	_
Density of tensile bar	8.74	8.38	8.25	8.22	8.44	8.28	_	_
Density of leak-test cylinder	8.75	8.32	8.33	8.21	8.30	8.28	8.00	8.16
verage of Heats A5892 and	d A5990							
ensile strength, psi	30,300	16,900	24,300	15,900	15,300	19,000	_	_
longation, % in 2 in.	21.4	7.3	15.3	7.0	5.8	9.9	_	_
Density of tensile bar	8.79	8.47	8.40	8.38	8.48	8.41	_	-
Density of leak-test cylinder	8.77	8.38	8.44	8.36	8.32	8.39	8.11	8.23
verage of All Heats								
ensile strength, psi	32,900	26,600	29,600	21,500	20,900	26,300	_	-
longation, % in 2 in.	27.8	18.0	23.8	10.3	10.0	17.9	_	_
ensity of tensile bar	8.83	8.65	8.61	8.57	8.63	8.61	-	-
ensity of leak-test cylinder	8.80	8.57	8.66	8.57	8.56	8.59	8.32	8.39
* Average data obtained or	castings po	oured in natu	ral green	** Exclusion	n of castings	poured outside	of the indi	cated tem-

sand, synthetic baked sand, and commercial coated sands only.

Data on other mold materials not considered to be representative of usual foundry practices.

** Exclusion of castings poured outside of the indicated temperature ranges accounts for the apparent absence of data for this heat.

# Effects of Mold Materials on Tensile Properties and Density

Table I and Fig. 3 show the effects of mold material on the average tensile properties and the average density of tensile-test bars and leak-test cylinders of 85-5-5-5 alloy. The average data are for individual heats varying widely in quality. They were prepared the high-frequency induction furnace and in the in

gas-fired tilting furnace, as shown by the information in the table, and then cast into the various mold materials at the average temperatures indicated. Table 2 and Fig. 4 show similar data on 81-3-7-9 alloy.

The high tensile properties and densities obtained in the calcined-clay mold baked at 1600 F and cooled to about 300 F before pouring are outstanding. The reason, of course, is that the melts of either alloy

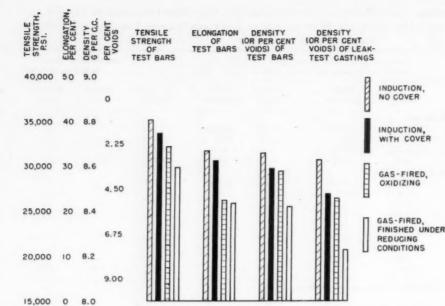


Fig. 5—Effect of melting practice on tensile properties and density of 85-5-5-5 alloy castings poured at about 2240 F. (Average results obtained in three different mold materials. See text.)

poured into this inert mold material absorb little or

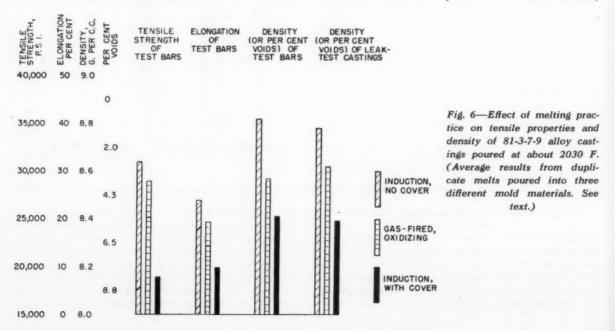
The outstanding feature of the data represented by Fig. 3 is the extremely low tensile properties obtained with 85-5-5-5 alloy cast in the green synthetic sand, despite the fact that fairly satisfactory density values were obtained. Test bars made in this mold material invariably fractured in the shoulder. Metallographic examination and density determinations made on sections along the length of the test bar showed that there was localized unsoundness in the shoulder where fracture occurred. Such a localized defect did not, apparently, reduce the density appreciably. Of the mold materials tested, the synthetic green sand also produced the lower tensile properties obtained with 81-3-7-9 alloy.

The occurrence of localized unsoundness in test

bars cast in synthetic green sand has been observed earlier.² The mechanism by which this mold material produces the defect is not known. Although it may be possible to produce satisfactory castings in synthetic green sand, the leak-test data discussed later indicate that, at least with the type of sand used in this investigation, leaky castings are usually obtained.

### Effects of Melting Practices on Tensile Property and Density

The effects of melt quality on the tensile properties have been evaluated by taking the average values obtained on castings poured into natural green sand, the coated sand, and the baked synthetic sand only. The use of the other two mold materials, calcined clay baked at 1600 F and green synthetic sand, was not considered to be representative of usual foundry



practice. Actually, the coated sand is not commonly used in brass and bronze foundries. Except for the tensile properties with 81-3-7-9 alloy, the results obtained with the coated sand were fairly close to those obtained in natural green sand. The data obtained with the coated sand were, therefore, included in

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The effects of melting practice on the tensile properties and density of 85-5-5-5 alloy are illustrated by Fig. 5. The tensile properties and density obtained by melts prepared in the high-frequency induction furnace under a glass cover are somewhat higher than normally obtained. In order to produce melts of high gas content by this method, it is necessary to keep the melts carefully covered during the melting operation, and then transfer the cover to the pouring ladle and pour from beneath it. Failure to do this probably accounts for the fact that the properties and densities of the test bars are higher than those normally obtained.

Figure 6 shows the effect of melting practice on the tensile strength, per cent elongation, per cent voids, and density of 81-3-7-9 alloy castings. Melting in the high-frequency induction furnace, employing a glass cover, produced the low-quality melts that are usually obtained by this melting method. For the reasons cited previously, the effect of melting practice is based on the average results obtained from the three

mold materials only.

The leak-test cylinders were tested at pressures up to 1300 psi. The testing pressure at which leakage

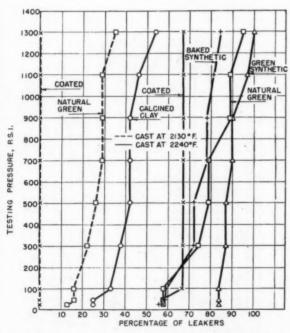


Fig. 7—Effect of mold material on the percentage of machined cylinders of 85-5-5-5 alloy which leaked at various testing pressures. Each curve for low pouring temperature represents the average of 19 to 31 cylinders; at high pouring temperature, 19 to 32 cylinders. Eight melts are represented, varying in quality from high to very poor.

was produced versus the percentage of leakers is plotted in Figs. 7, 8, 9, and 10. These charts show the effects of melting practice and mold material on the percentage of leakers obtained. It should be emphasized, of course, that the castings were machined inside and out and on both ends prior to testing. Because the exterior of the casting is generally sounder than the interior, the removal of the exterior of the casting will, of course, greatly accentuate the tendency to leak.

# Effects of Melting Practice and Mold Material on Leak Tightness

85-5-5 Alloy: Figure 7 shows the effects of mold material on the occurrence of leakage in 85-5-5-5 alloy cylinders poured at high and medium pouring temperatures. Data from eight melts varying in quality from very good to poor are illustrated. The occurrence of leakage is greatly reduced at the lower pouring temperature. It will also be noted that the coated sand molds produced zero per cent leakers at 1300 psi when the pouring temperature was approximately 2130 F, whereas natural green sand produced 35 per cent leakers at a testing pressure of 1300 psi under similar pouring conditions. At the higher pouring temperature, the coated sand is again slightly superior to natural green sand. In general, the mold materials for 85-5-5 alloy may be rated on the basis of tensile properties, density, and tendency to form leakers as follows:

Rating	On the Basis of Tensile Strength	On the Basis of Density of Leak-Test Cylinders	On the Basis of Per Cent Leakers Obtained at 1300 psi on Castings Poured at High Temperatures
Best	Calcined clay  Coated sand.	Calcined clay	Calcined clay
	baked synthetic sand, and natural green sand	Natural green sand and green synthetic sand	Coated sand, baked synthetic sand, and natural green sand
Poorest	Green synthetic sand	Baked synthetic sand and coated sand	Green synthetic sand

Figure 8 shows the effect of melting practice on the occurrence of leakers when the 85-5-5-5 alloy cylinders are poured into the various mold materials at a high and at a medium pouring temperature. As indicated by Table 1, the melts were prepared under the following conditions:

1. Duplicate heats in a high-frequency induction furnace without employing a liquid-glass cover.

Duplicate heats in a high-frequency induction furnace, employing a liquid-glass cover.

3. Duplicate heats in a gas-fired furnace, employing an oxidizing flame.

4. Duplicate heats in a gas-fired furnace, finished

under reducing conditions.

In general, the melt quality decreased and the gas content increased in the above order, as indicated by the tensile properties and density data presented in Fig. 5. It will be noted in Fig. 8 that the tendency to form leakers in castings poured at 2235 F is also in the above order, indicating that, as the gas content of

the melt increases, the tendency to form leakers also increases. It should be emphasized, however, that a reversal may be expected when melts of very low gas content are employed, if low pouring temperatures are used, and the castings are not well fed. Melts of low gas content poured at low temperature are more apt to form localized shrinkage which, in turn, will accentuate the occurrence of leakers.

It may be safely assumed that duplicate melts prepared in the high-frequency induction furnace without a cover closely approach the optimum gas content to provide a minimum of leakers. All the other melting methods, however, produce gas contents which are higher than the optimum for the produc-

tion of a minimum percentage of leakers.

81-3-7-9 Alloy: Figure 9 shows the effects of mold material on the occurrence of leakers in 81-3-7-9 alloy. It will again be noted that the castings poured in the higher temperature range were characterized by a very much higher percentage of leakers than those poured at the lower temperature.

The mold materials for 81-3-7-9 alloy may be rated according to the three criteria as follows:

Rating	On the Basis of Tensile Strength	On the Basis of Density of Leak-Tcst Cylinders	On the Basis of Per Cent Leakers Obtained at 1300 psi on Castings Poured at Low Temperatures		
Best	Calcined clay Baked synthetic sand and natural green sand	Calcined clay Baked synthetic sand	Calcined clay Baked synthetic sand, coated sand, and natural green sand		
Poorest	Coated and green synthetic sands	Coated sand, natural green sand, and green synthetic sand	Green synthetic sand		

It will be noted from the above that, with respect to leak tightness, coated sand again is slightly superior to the natural green sand. A fair, though not perfect, correlation is obtained with the tensile properties of separately cast test bars. The ratings of mold materials on the basis of density of the leak-test castings again correlate only in a general way with either of the other two criteria for rating the mold materials.

It will also be noted that the baked synthetic sand rates higher in respect to the attainment of leak tightness, as well as high tensile properties, when the alloy is 81-3-7-9 instead of 85-5-5-5. Such a difference in the results obtained with the two alloys in baked synthetic sand would hardly be expected. It is not known whether or not the difference in the results obtained with the two alloys in this mold material is real, or whether it is a result of an uncontrolled variation during the conduct of the experimental work.

Figure 10 illustrates the effects of melting practice on the occurrence of leakers. By comparing the trends shown in Fig. 10 with those shown in Fig. 6, it is noted that as the gas content of the melt increases the occurrence of leakers increases.

Causes of Leakage: Several representative 85-5-5-5 alloy cylinders were radiographed, and sections cut from the castings were examined under the micro-

scope to determine the cause of leakage. Metallographic specimens were cut from leaky castings and a section polished near the location of the leak. Figures 11 to 20 illustrate the results obtained by this study. Of the ten castings illustrated, only one shows some shrinkage, but this defect did not cause a leak in this instance. The leaks in all of the castings examined were caused by microporosity.

Microporosity and shrinkage cavities differ in degree. Microporosity is generally distributed and is initiated by shrinkage and accentuated by gas evolution. Shrinkage cavities are coarser defects than microporosity and are associated with lack of adequate feeding. Furthermore, shrinkage is accentuated by decreasing pouring temperature, whereas microporosity is decreased by decreasing pouring temperature.

# Pouring Temperature and Microporosity

All of the 85-5-5-5 alloy castings examined were poured at relatively high temperatures. As would be expected, therefore, microporosity was far more of a problem than shrinkage. Had the pouring temperature been low, on the order of 2050 F, much less microporosity would have been encountered but more shrinkage would have occurred. It has already been seen, however, that the low pouring temperatures favor the production of leak-tight castings.

It should also be noted that dross and trapped gas bubbles did not appear in any of the radiographs. The absence of these defects is a direct result of the bottom gating technique, which prevented excessive

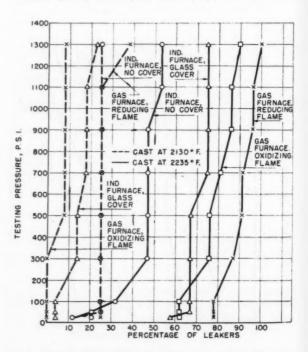


Fig. 8—Effect of melting practice on the percentage of machined cylinders of 85-5-5-5 alloy which leaked at various testing pressures. Each curve for low pouring temperature represents the average of eight to 22 cylinders cast in coated sand and natural green sand; at high pouring temperature 12 to 23 cylinders cast in coated sand, natural green, and baked synthetic sand.

turbulence during the pour.⁵ The cause of leakage in the 81-3-7-9 alloy castings was not determined.

#### Summary

Eight 300-lb heats of 85-5-5-5 alloy and six 300-lb heats of 81-3-7-9 alloy were prepared by various methods to produce melt qualities varying from very good to very poor. The 14 melts were cast into test bars and into leak-test cylinders, four per mold, employing the following mold materials:

- Calcined clay baked at 1600 F and cooled to 300 F prior to pouring.
  - 2. Natural green sand.
  - 3. A coated sand.

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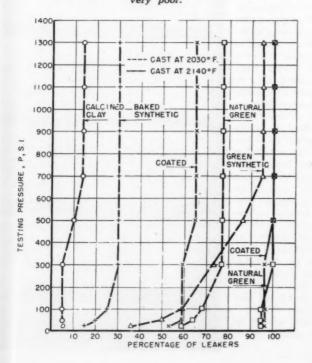
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- 4. Synthetic sand bonded with bentonite and baked at 450 F.
  - 5. Green synthetic sand bonded with bentonite.

Probably because of the nearly complete absence of gas absorption from the mold, the inert calcined-clay mold produced test specimens having the highest tensile properties, and leak-test cylinders having the highest density and the lowest percentage of leakers. The mold materials were rated on the basis of tensile properties, density of leak-test castings, and the percentage of leakers obtained in the various mold materials. The three criteria were in general agreement, but did not agree precisely. Based on the occurrence of leakers at a testing pressure of 1300 psi, the following is the order of decreasing merit of the five mold materials for the 85-5-5-5 and 81-3-7-9 alloys:

Fig. 9—Effect of mold material on the percentage of machined cylinders of 81-3-7-9 alloy which leaked at various testing pressures. Each curve for low pouring temperature represents the average of 17 to 22 cylinders; at high pouring temperature, 20 to 24 cylinders. Six melts are represented, varying in quality from high to very poor.



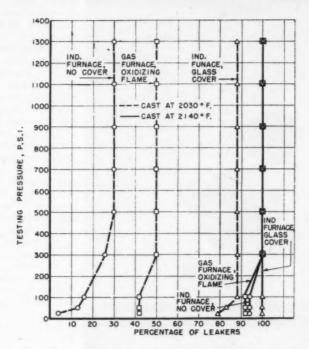


Fig. 10—Effect of melting practice on the percentage of machined cylinders of 81-3-7-9 alloy which leaked at various testing pressures. Each curve for low pouring temperature represents the average of 12 to 24 cylinders cast in coated sand, natural green sand, and baked synthetic sand; at high pouring temperature, 12 to 16 cylinders cast in coated sand and natural green sand.

(p	85-5-5-5 Alloy (poured at about 2240 F)  1. Calcined clay 2. Coated sand		81-3-7-9 Alloy (poured at about 2030		
1.	Calcined clay		1.	Calcined clay	
2.	Coated sand			Baked synthetic sand	
3.	Baked synthetic sand		3.	Coated sand	
4.	Natural green sand		4.	Natural green sand	
5.	Green synthetic sand	100	5.	Green synthetic sand	

X-ray and metallographic examinations have shown that the cause of leakage in the 85-5-5-5 alloy cylinder poured at a high temperature is primarily microporosity. Trapped gas holes and dross did not occur because the bottom gating technique was used. Shrinkage, as a defect, was not important because of the high pouring temperatures employed.

On the other hand, the occurrence of microporosity in the 85-5-5-5 alloy castings was accentuated considerably by high pouring temperatures and by the employment of melting practices which do not provide for strongly oxidizing conditions during the preparation of the melt. The cause of leakers in the 81-3-7-9 alloy castings poured at a low temperature was not determined. In all probability, however, the causes were shrinkage and microporosity.

The results show that melts prepared under reducing conditions produce fairly sound castings in the calcined clay, whereas the same metal produces extremely unsound castings in the other four mold materials. This may be rationalized by assuming that both oxygen and hydrogen are absorbed from the water vapor formed in all but the calcined-clay

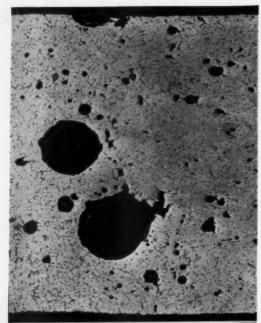
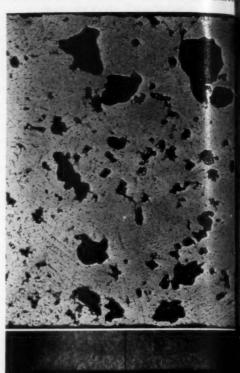


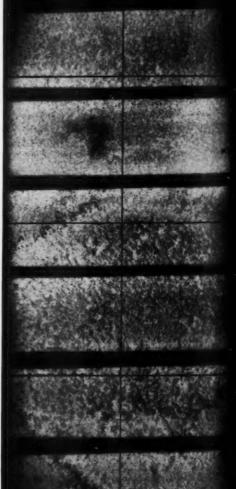


Fig. 11 (left)—Radiograph (1 X) and photomicrograph (20 X); alloy 85-5-5; cylinder No. A-4728D1*; melting method, high-frequency induction, no cover; mold material, coated sand; pouring temperature, 2240 F; density of cylinder: 8.72 grams per cc; leaks, none; section photomicrographed, top, Quadrant 3. In this and all of the following radiographs the top of the radiograph represents the top of the casting, and the four quadrants are numbered 1 to 4, clockwise, No. 1 being the quadrant to which the bottom gate is attached (Fig. 1). The radiograph shows unsoundness at the tops of the third and fourth quadrants. A stringer of shrinkage is also present, but this cylinder did not leak. Obviously, the defects did not pass all the way through the section. The rounded gasholes shown by the photomicrograph are fairly typical of castings made in the commercial coated sand, but such rounded gasholes are not likely to cause leakage. This may account for the relatively good showing of this mold material from the standpoint of the production of leak-tight castings, despite the low density and tensile properties produced as a result of the considerable gas absorbed from the mold.

Fig. 12 (right)-Radiograph (1 X) and photomicrograph (20X); alloy 85-5-5; cylinder No. A-4729 I4; melting method, high-frequency induction furnace, liquid-glass cover; mold material, coated sand; pouring temperature, 2210 F; density of cylinder, 8.06; leaks, 7 at 5 psi pressure, bottoms of Quadrants 1, 2, and 4, and top of Quadrant 2; section photomicrographed, bottom, Quadrant 1. The radiographs show extremely heavy, generally distributed microporosity which is the cause of the leaks. The microporosity is so coarse that it borders on pinhole type of porosity, as shown by the photomicrograph. The extreme unsoundness is a result of the high gas content of the melt in the pouring ladle and the considerable gas absorption from the coated sand mold.

* The last number indicates the position of the cylinder in the casting (Fig. 1).





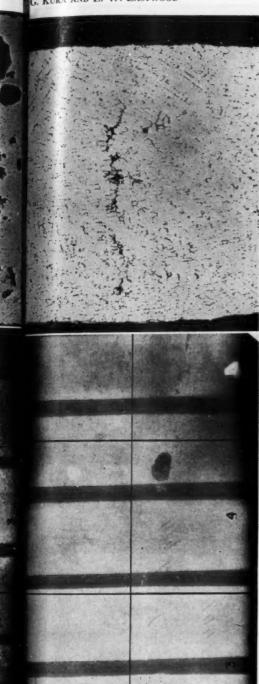
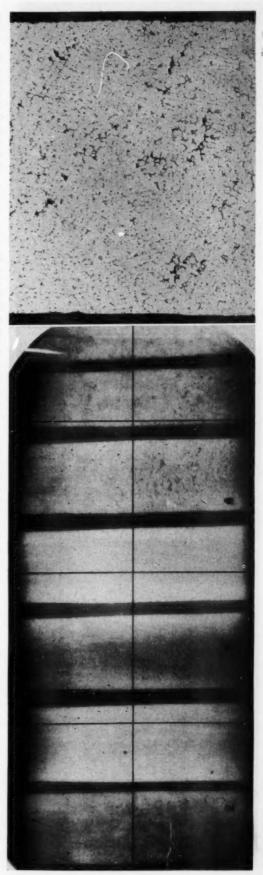
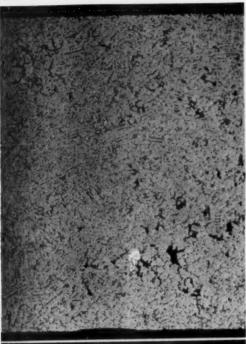


Fig. 13 (left)-Radiograph (1 X) and photomicrograph (20 X); alloy 85-5-5; cylinder No. A-4950 B1; melting method, gas-fired furnace, oxidizing flame; mold material, calcined clay baked at 1600 F and cooled to 300 F prior to pouring; pouring temperature 2270 F; density of cylinder, 8.86; leaks, 1 at 10 psi pressure, bottom of Quadrant 1; section photomicrographed, bottom Quadrant 1. The radiograph shows this to be a relatively sound casting with a small amount of microporosity in Quadrant 1 adjacent to the gate. This unsoundness caused the one leak. The elongated microporosity extending almost all the way across the section is shown by the photograph of a section adjacent to the gate. Despite the fact that this is an inert mold, a small amount of gas evolution or perhaps inadequate feeding caused the formation of microporosity adjacent to the gate, which, in turn, caused the single leak. The fine, dark particles shown in all of the photomicrographs are lead.

Fig. 14 (right)—Radiograph (1X) and photomicrograph (20X); alloy 85-5-5; cylinder No. A-4727 B3; melting method, high-frequency induction furnace, liquid-glass cover; mold material, calcined clay baked at 1600 F and cooled to 300 F prior to pouring; pouring temperature, 2240 F; density, 8.81; leaks, none; section photomicrographed, bottom Quadrant 1. As would be expected with this mold material, the radiograph again shows a fairly sound casting with some microporosity adjacent to the gate at the bottoms of Quadrants 1 and 4. The microporosity, however, did not cause any leakage, even at 1300 psi. As compared with Fig. 13, this melting method has produced slightly more microporisty, as would normally be expected. The microporosity observed in this cylinder and in Fig. 13 arises from the evolution of a small amount of gas during solidification. The gas, however, was in all probability in

(Fig. 14 caption continued on following page.)







(Fig. 14 caption continued from preceding page.)

the metal in the pouring ladle and probably was not absorbed from this inert mold. The photomicrograph of a section cut from the bottom of Quadrant 1 adjacent to the gate shows fine, angular, more or less connected microporosity. Leakage did not occur because the microporosity was not continuous across the section.

Fig. 15 (left)-Alloy 85-5-5-5; cylinder No. A-4951 K3; melting method, gas-fired furnace, oxidizing and then a reducing flame; mold material, natural green sand; pouring temperature, 2225 F, density, 8.59; leaks, 15 at 1300 psi pressure, bottom and top of Quadrant 3; section photomicrographed, top Quadrant 3. The radiograph (1X) shows some microporosity in the area adjacent to the gate (Quadrant 1), and more concentrated microporosity in the upper right portion of Quadrant 3 where the leaks occurred. The photomicrograph (20X) of this section shows the occurrence of the angular, more or less connected microporosity, some of which apparently extends all the way across the section of the casting, causing leakage.

Fig. 16 (right)—Alloy 85-5-5-5; cylinder No. A-4728-J1; melting method, high-frequency induction furnace, no cover; mold material, natural green sand; pouring temperature, 2165 F; density, 8.68; leaks, 3 at 25 psi pressure, bottoms of Quadrants 3 and 4; section photomicrographed, bottom Quadrant 3. The radiograph (1X) shows fairly heavy microporosity in the region to one side of the gate. The leaks occurred in this area. The nature of the microporosity in the leaky area is shown by the photomicrograph (20X).



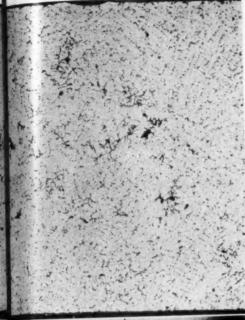
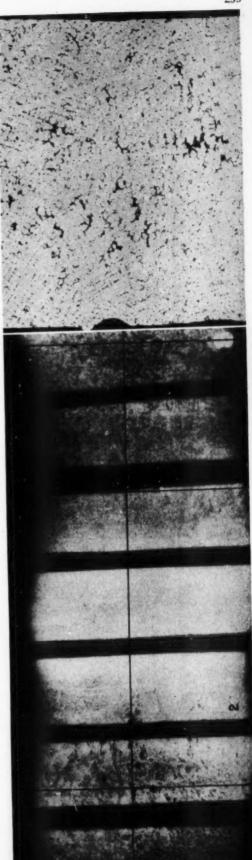


Fig. 17 (left)-Alloy 85-5-5; cylinder No. A-4728 J2; melting method, high-frequency induction furnace, no cover; mold material, natural green sand; pouring temperature, 2165 F; density 8.76; leaks, none; section photomicrographed, bottom Quadrant 3. This casting was made in the same mold as the one shown in Fig. 16. The microporosity is much finer, as shown by the radiograph (1X) and the photomicrograph (20X). This fine microporosity apparently does not extend all the way across the section of the casting and did not cause a leak.



Fig. 18 (right)-Alloy 85-5-5-5; cylinder No. A-4933 G2; melting method, gas-fired furnace, reducing flame; mold material: green synthetic sand; pouring temperature, 2250 F; density, 8.72; leaks, 1 at 5 psi pressure, top Quadrant 2; section photomicrographed, top Quadrant 2. This melt. made under reducing conditions in the gas furnace, has a high gas content which, in turn, has contributed to the large volume of microporosity shown in the radiograph (1X) and in the photomicrograph (20X). As expected, the microporosity is most pronounced in the region adjacent to the gate. The section represented by the photomicrograph was cut from a location adjacent to the leak which, however, was not in the region of heaviest microporosity.



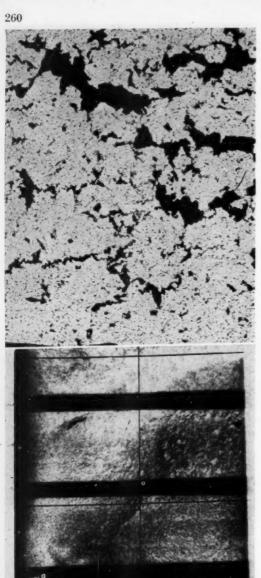
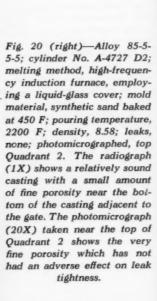


Fig. 19 (left)-Alloy 85-5-5; cylinder No. A-4933 G4; melting method, gas-fired furnace, reducing flame; mold material, green synthetic sand; pouring temperature, 2250 F; density, 8.18; leaks, 1 at 5 psi pressure, top Quadrant 3; section photomicrographed, top Quadrant 3. Castings made in the same mold may vary in soundness. The cause of this difference is not known. Cylinders shown in Figs. 18 and 19 were made in the same mold. The greater degree of unsoundness and lower density (Fig. 19) are quite apparent. Extremely heavy microporosity, distributed throughout the section and somewhat coarser in the area adjacent to the gate is evident in the radiograph (1X). The photomicrograph (20X) shows the extreme coarseness of the microporosity. This unsoundness is attributable primarily to the poor melting practice intentionally employed. Comparing with Fig. 14, it is seen that extreme unsoundness produced by a poor melting method occurs only when gas, presumably oxygen and hydrogen, is also absorbed from the mold.





ASTIN

waterless) molds according to the following reacion as it proceeds from left to right,

Metal  $+ H_2O \rightleftharpoons 2H + metal oxide$ .

If the melt being poured is low in oxygen and high n hydrogen, the absorption of oxygen from the water vapor (steam) in the molds provides sufficient oxygen (metal oxide) for the formation and evolution of water vapor as the above reaction reverses itself at lower temperatures or as the alloy solidifies. This accentuates microporosity. Apparently, the amount of oxygen absorbed from the air is insufficient to produce a damaging amount of water vapor when solidification takes place in a calcined-clay mold.

# Acknowledgments

The authors wish to express their appreciation to the Brass and Bronze Ingot Institute who sponsored the work described and authorized its publication, to G. P. Halliwell, Chairman, H. L. Smith, and A. H. Hesse, members of the Research Subcommittee who have contributed a great deal to the work, and to the Editorial Committee, W. Westphal, G. P. Halliwell, and A. H. Hesse who ably reviewed the paper. The authors also extend their thanks to their fellow staff members who assisted in this work, especially to Carl Owens who carried out the major portion of the laboratory work, and to the H. Kramer & Co. who conducted all of the analytical tests.

# References

1. W. T. Pell-Walpole, Foundry Trade Journal, January and February, 1946.

2. L. W. Eastwood and J. G. Kura, "Effect of Mold Material on Gas Absorption by 85-5-5-5 Alloy," The Foundry, October,

3. H. Lepp, International Foundry Congress, Prague, 1933, pp. 95-118.

4. L. W. Eastwood, Gas in Light Alloys, John Wiley & Sons, Inc., New York, 1946, pp. 48-51.

5. J. G. Kura and L. W. Eastwood, "Effect of Gating Practice on the Leak Tightness of 85-5-5-5 and 81-3-7-9 Alloy Castings, A.F.S. Transactions, vol. 60, pp. 287-295 (1952).

#### DISCUSSION

Chairman: C. A. ROBECK, Gibson & Kirk Co., Baltimore. Co-Chairman: B. F. Shepherd, Ingersoll-Rand Co., Phillipsburg, N. J.

Recorder: F. L. RIDDELL, H. Kramer & Co., Chicago.

S. A. KUNDRAT: 1 Was there any organic or carbonaceous material, such as wood flour or cereal, added to the synthetic sand? MR. KURA: No. The synthetic sand consisted of washed silica sand, bentonite and water.

R. H. GILBERT: 2 Was the gas pickup while melting the same for each heat? The melting practice should have more effect than the molding practice or type of investment used.

MR. KURA: Test bars poured in the calcined clay mold were used to classify the quality of the heat.

G. E. Dalbey: Were the "bubbles" or gas holes, apparent near the surface of the metal, caused by gas in the metal or by gas generated from the core or molding sand?

MR. KURA: There were no dry sand cores used. The subsurface defects were caused by the mold materials while pouring.

WERNER FINSTER: 4 Why did the metal cast in the green synthetic sand give low properties?

MR. KURA: In most cases there was a local shrink in the shoulder of the test bar.

H. H. FAIRFIELD: 5 Did the residual phosphorus vary from heat to heat? Would you expect to get a variation in the residual phospherus with the same amount added to each heat?

MR. KURA: There was little variation in the residual phosphorus in the various heats. The amount of phosphorus remaining in the alloy would depend on several factors, that is, the degree of oxidation of the liquid metal and the manner in which the phosphorus is added.

P. L. BUTLER: 6 Did you investigate the variation in the thermal conductivity of the various mold materials?

MR. KURA: We did some work on this but there were no definite results.

MR. FAIRFIELD: When making large castings, we have used sand mixed with 40 per cent graphite to get lower thermal conductivity which would help feeding.

MR. GILBERT: Is the test bar illustrated in Fig. 3 of the ASTM specification on test bars a good indication of melt quality? MR. KURA: By using this test bar, we were able to differentiate

between the effects of molding materials and melting conditions. Mr. Shepherd: It has often been thought that if we could obtain the ideal density metal, that is, a cast metal of the lowest possible gas content, our foundry problems would be over. In some experiments run in our foundry on some pump impellors made from Navy "M" metal, we used lithium for deoxidation after melting in a strongly oxidizing atmosphere. The density obtained was exceptionally high, however there was a marked local shrink evident in one section of the impellor. Under normal melting practices, a lower density was obtained but there was no evidence of a shrink in the castings. This would indicate that the high density metal, that is, one with low gas content, does not always produce the soundest metal.

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² Metallurgist, H. B. Saiter Mfg. Co., Marysville, Ohio.

³ Metallurgist, Mare Island Naval Shipyard, Vallejo, Calif.

⁴ Metallurgist, Reading Steel Casting Div., American Chain & Cable Co., adding. Pg. * Metallugist, Reading, Pa.

5 Chief Metallurgist, Wm. Kennedy & Sons Ltd., Owen Sound, Canada.

6 Chief Metallurgist, Arwood Precision Casting Corp., Brooklyn.

# REFRACTORIES FOR THE BASIC CUPOLA

By

H. M. Kraner*

The primary purpose of lining a cupola with basic refractories is to make it possible to maintain certain slag compositions which will yield desired metallurgical results. Good slag control is possible only through the use of a refractory lining which is more or less permanent in character and dissolves in the slag in lowest degree during the run. One might attempt to carry a basic slag in an acid lining but the lining would not survive very long in this environment. Likewise it is not practicable to carry a slag which is acid in character (siliceous) in a basic-lined cupola.

Everything which enters the cupola, as in most other metallurgical furnaces, must leave as a gas or a liquid. The slag must be of a composition at normal furnace operating temperature that will flow freely. The use of lime or other fluxes in an acid furnace is merely to liquefy the refuse material, such as coke ash, etc. In the basic cupola the added reason is to produce a slag which will make it possible to desulphurize the metal being melted.

The blast furnace does a good job of carrying away refuse material and preparing its slag for the desulphurizing action which it has to perform. The cupola, because of its shorter height and smaller working volume, is not as apt to perform as smoothly or as efficiently as does the blast furnace, and it is therefore not unlikely that there will be considerable variation in slag composition from run to run, and perhaps even from hour to hour during a run.

Slags involved in reported papers vary considerably but in general they tend toward compositions of blast furnace slags.

	Renshaw 1943	Renshaw 1949	Carter 1950	Renshaw 1951
SiO,	14.44	37-48	31.6-39.2	30-35
Al,O,	1.50	4.8-14		8-10
Fe,O,	3.44			<1.5
CaO	40.4	19-31	25.8-31.5	42-45
MgO	40.3		8.7-18	9-11
MnO				< 1.5
Loss on				
Ignition	0.25			

^{*}Ceramic Engineer, Research Department, Bethlehem Steel Co., Bethlehem, Pa.

In the blast furnace the slags are maintained at a high degree of basicity in order to desulphurize the metal as thoroughly as possible in the time the metal is in contact with the slag. Much of the desulphurization takes place as the metal trickles through the slag layer. The slag must be liquid not only to be tapped from the furnace, but to do the job of desulphurizing efficiently. A high degree of basicity and ample fluidity are necessary for satisfactory operation. If the slag beccmes more basic than normal it is more viscous due to the presence of some crystal phases.

The blast furnace does not use a basic refractory but the refractories used are extensively water-cooled. The problem of survival of refractories involves a compromise between viscous slags which neither dissolve the refractory nor carry out the desulphurizing process, and a slag of proper composition and viscosity which does a good job of desulphurizing but which might also dissolve considerable refractory.

A slag which contains crystals is said to be saturated with regard to these particular crystals at the temperature at which they coexist with the liquid portion of the slag. Such a slag in a furnace operation, even though it is saturated with crystals of a particular mineral might still have a capacity for the solution of the constituents of a refractory.

# Potential Solubility of Slag

If the temperature of this slag is raised above the point at which all of its crystals are dissolved, it has a potential solubility for almost any other mineral constituent with which it comes in contact. By raising the temperature of the slag, therefore, its dissolving action is automatically increased. Carbon as a refractory is a possible exception to this and, except for certain practical operating reasons, carbon might have very satisfactory properties of interest to the operator of the basic cupola. As hot liquid slags dissolve almost anything which comes in contact with them, the reason for being interested in slags in connection with refractories problems is quite evident.

The terms "acid" and "basic" often erroneously imply that all one needs to do to provide resistance to a basic slag is to use a basic refractory. While this is true in a practical sense, it is not scientifically true. A good example of the fallacy of this lies in a record of basic refractories in the basic cupola.

In 1943 Renshaw reported his basic lining, consisting of rammed "basic,"—a hard-calcined dolomite

made in England.

At that time his slag contained 40.3 per cent MgO and 40.4 per cent CaO, and this quite evidently eroded from the lining. This slag, containing 40.3 per cent MgO, was probably not entirely liquid or fluid. His 1949¹⁴ slag, and that of Carter (1950), ¹⁶ and Renshaw's (1951)¹⁰ slags are more nearly normal American blast furnace slags. Even these, however, have potential dissolving power for magnesia. It is well known that in the system CaO-MgO-Al₂O₃-SiO₂ included in blast furnace slag compositions as much as 20 per cent MgO is found dissolved in fluid slags at 2700 F.

# Consumption of Refractory Lining

It is not surprising therefore to find that Renshaw* in his basic cupola operation used as many pounds of lining as he had in his acid lining practice. The basic brick, however, cost several times more per pound than the acid lining materials. Although the solution of extra magnesia or lime in a slag does not adversely affect the slag metallurgically, this would not be the case had the slag been contaminated by "acid" lining dissolving in it, unless the extra lime or magnesia "saturates" the slag, and crystalline materials appear. This more or less primes the operator not to expect either a panacea or a consolation, but to expect a higher refractory cost in the operation of the basic cupola.

The blast furnace is one of the most efficient but the most extensively water-cooled furnaces we have. Only clay refractories or carbon are now used. Although carbon is insoluble in blast furnace slags and holds up well in the presence of slags corrosive to most other refractories, water-cooling is essential to both carbon and clay refractories and is, to a large extent, responsible for it being possible to use clay refractories in the blast furnace. Water-cooling of carbon refractories causes metal and slag to be congealed on the surface and in block joints, with a lifesaving result. The cost of maintaining the furnace may in such cases be found in the slight extra fuel cost due to the heat lost through such water-cooled walls. Water-cooling in cupolas is not new and it is frequently advantageous from a cost standpoint, continuity of operation, and because of likely better slag control.

In a small furnace, the ratio of cooled wall surface to cross-sectional area is large. In a larger furnace this ratio is smaller and the heat loss would be less of a factor than would be the case in a small blast furnace or a small cupola. The operation of the larger water-cooled furnace or cupola would probably be more readily controlled. If one were to use carbon

in a cupola lining the occasional mistake of insulating it, should be avoided. Insulating, or in effect so doing, by backing-up carbon brick with clay brick builds up temperature throughout the carbon wall, allows slag and metal to penetrate through the entire joint and prevents the accumulation of a protective slag layer on its surface. If ceramic brick of any kind are to be used in connection with carbon brick or blocks, the carbon should be placed near the shell to assist in the cooling of the inner wall. It has been found that brick backed-up by carbon and the carbon cooled with water-cooling on the shell forms an effective combination providing long life and minimum contamination to the slag. The benefit is obviously derived directly from the water-cooling and would be effective in some degree through any kind of brick, even without the carbon in the wall.

It is our understanding that one basic cupola operator in this country has lined his cupola with carbon and backed it up with magnesite brick extending to the water-cooled shell. His tap hole is also carbon. He is able to run an entire 5-day period at a refractory cost less than his daily cost had been with acid refractories on acid operation.

While this combination appears to be effective, it would seem that the positions of the magnesite brick and the carbon should be reversed or that the magnesite brick should be eliminated.

It has been the intent here to avoid a discussion of proprietary materials, ramming, spraying, or the other methods described in the very helpful papers already printed on this subject and included in the appended bibliography. It should suffice to say that we, in the United States, have an abundant supply of magnesite brick ranging from 75 to 92 per cent MgO, and magnesite grain having a range of composition from 65 to 96 per cent MgO. The accessory materials in these magnesites, such as CaO, SiO₂, and Fe₂O₃ vary inversely with the magnesia contents.

# Ingredients in Refractory Mixtures

In making ramming or spraying mixtures manufacturers have added to these granular magnesite materials various proportions of magnesium chloride, magnesium sulphate, sodium silicate, clay, etc., and have provided ramming mixtures which can be used as monolithic linings rammed, daubed, or sprayed into place. We do not make the material referred to by the British as "basic," a hard calcined dolomite, nor do we make a stabilized dolomite brick. Both of these would probably be cheaper than magnesite or magnesite brick. While magnesite brick are probably the most expensive brick to be considered for a cupola and require bricklayer labor to install, they are fired and have a degree of density, uniformity of character, and reliability, sometimes not found in a rammed wall. Magnesite and dolomite products will probably not perform as well in contact with blast furnace type slags as they do in the open hearth where oxidizing slags are used. Water-cooling, and the use of carbon, should therefore offer considerable hope in critical areas of the basic cupola. Water-cooling

^{*}Foundry Trade Journal, Aug. 23, 1951.

105-118 (1941).

properly applied might even make it practicable to use high-grade clay refractories.

**Bibliography** 

1. J. T. MacKenzie, "Carbon and Sulphur in the Cupola," A.F.A. Transactions, vol. 38, pp. 385-432 (1930).

2. I. Trifonov, "Sulfur in Cupola Practice," Die Giesserei, vol. 20, Nov. 10, pp. 497-500 (1933).

3. Colin D. Abell, "Sulphur in the Cupola," Foundry Trade Journal, vol. 63, Oct. 3, p. 219 (1940).

4. T. Swinden and J. H. Chesters, "Dolomite Bricks for Use in Steel Works," Procedures, Iron & Steel Institute, vol. 144, pp.

5. E. S. Renshaw, "Basic Cupola Process for Desulphurisation," Foundry Trade Journal, vol. 70, June 24, 149-153 (1943)

6. P. C. Fassotte, "Developments in the Design and Use of Side-Blown Converter Plants," Journal Iron & Steel Institute, vol. 150, no. 2, pp. 339-358 (1944).

7. C. Heiken, "Production Results with a Basic-Lined Cupola," Die Giesserei (Oct. 28, 1934) and Foundry Trade Journal,

vol. 72, Mar. 2, pp. 185-187, 190 (1944).

8. First Report on the Basic Cupola, Melting Furnaces Subcommittee (British Committee), Procedures, I. B. F., May (1944).

9. Second Report on the Basic Cupola (British Committee) Procedures, I. B. F., vol. 39, April (1946).

10. Second Report, Subcommittee T.S. 10, Institute of British Foundrymen, Foundry Trade Journal, vol. 79, July 4, pp. 239-242 (1946).

11. E. S. Renshaw and T. Foley, "Duplexing Low Carbon Al-

loys," The Iron Age, vol. 159, May 22, pp. 56-59 (1947). 12. G. R. Rigby and A. T. Green, "Refractory Materials in the Foundry Industry," Procedures, I. B. F., vol. 41, pp. A144-51 (Paper no. 906) (1948).

13. W. H. Bamford (B.I.S.R.C. Conference on Cupola and Converter Refractories): "Water-Cooling of Cupola and Tuyeres," Foundry Trade Journal, vol. 86, no. 1707, p. 473 (1949).

14. E. S. Renshaw and S. J. Sargood, "Some Modifications in Cupola Design," Foundry Trade Journal, vol. 87, no. 1728, pp. 449-456 (1949).

15. F. H. Norton, Refractories, McGraw-Hill (1949).

16. S. F. Carter, "Basic-Lined Cupola for Iron Melting," Trans-ACTIONS, A.F.S.; vol. 58, pp. 376-392 (1950).

17. W. Levi, "Operation of the Cupola," Transactions, A.F.S., vol. 58, pp. 1-19 (1950).

18. Robert Doat, "The Metallurgical Blast Cupola," Trans-ACTIONS, A.F.S., vol. 60 (1952).

19. E. S. Renshaw, "Basic Cupola Melting and Its Possibili-

ties," Transactions, A.F.S., vol. 59, pp. 20-27 (1951).

20. Correspondence of F. R. H. Allon, Foundry Trade Jour-

nal, vol. 91, Aug. 30, p. 232 (1951).
21. J. P. Holt, "Basic Cupola Operation," American Foundryman, vol. 21, no. 1, pp. 39-43 (1952).

22. W. W. Levi, "Melting Iron in a Basic-Lined Water-Cooled Cupola," Transactions, A.F.S., vol. 60 (1952).

#### DISCUSSION

Chairman: R. A. WITSCHEY, A. P. Green Fire Brick Co., Chicago.

Co-Chairman: S. F. CARTER, American Cast Iron Pipe Co., Birmingham, Ala.

Recorders: R. A. WITSCHEY and S. F. CARTER.

W. W. LEVI (Written Discussion)1: The author is to be congratulated on his presentation of concise data which should be of considerable value to those interested in operating basiclined cupolas. For example, Mr. Kraner's comments on the importance of slag fluidity for good sulphur reduction is in line with our experience. We have found that slags of a given basicity and relatively poor fluidity may be less effective in removing sulphur than a less basic, highly fluid slag. As pointed out by Mr. Kraner, we have found that "basic" refractories may be severely attacked by "basic" slags and this applies to the refractories in the well, breast, and trough of our front slagging cupola. Mr. Kraner mentioned the effectiveness of water-cooling in blast furnace practice and this has proved

to be most effective in basic cupola operation as well.

Mr. Kraner refers to slags reported by Renshaw in 1949, Carter in 1950, and Renshaw again in 1951 as "more nearly normal American blast furnace slags" and that "blast furnace slags are maintained at a high degree of basicity-contact with the slag." While this writer does not profess to be familiar with blast furnace operation, we have never, in connection with our basic cupola practice, considered slags containing more SiO2 than CaO as very strongly basic, though, this is the case in connection with Renshaw's 1949 slag and Carter's 1950 slag. In order to produce irons from our basic cupola with an average 0.016 per cent sulphur or less, we have operated with slags of the following average composition:

CaO ..... 52.7 per cent MgO ..... 8.2 per cent SiO₂ ..... 27.6 per cent Al₂O₃ ..... 5.0 per cent FeO ..... 0.63 per cent S ...... 1.35 per cent

The above slag is considerably more basic than any of the three slags previously mentioned as being more or less typical blast furnace slags and the sulphur content of the iron produced is lower than in any blast furnace pig ever received by our Company to date. Experience indicates that attack on refractories by this slag is more severe than by less strongly basic

slags, as would be expected.

In his paper, Mr. Kraner refers to a basic cupola in this country lined with carbon and backed up with magnesite brick extending to the water-cooled shell. His description makes it seem more than likely that the cupola in question is the one operated by our Company since our cupola is lined and operated in the manner mentioned. While we do not take a defensive attitude in connection with our method of lining, especially since we have not as yet tried backing the magnesite brick with carbon, we have a number of reasons for lining as we do, and some of these are as follows:

1. After several months of development work we were able to operate our basic cupola continuously for periods of 65 to 85 hr. The front-slagging spout was lined with magnesite brick between the tap hole and the dam, and these failed during the 65 to 85-hr periods due to the severity of the slag attack. These failures were eliminated by using rammed carbon backed with magnesite brick. This same combination of refractories (rammed carbon, backed with magnesite brick) was then tried in the cupola well and results obtained were very

2. The logic in backing magnesite brick with carbon against the water-cooled shell is apparent, because carbon has a higher heat conductivity than magnesite, and consequently there will be a tendency toward lower temperatures at the working face of the magnesite brick and this in turn will tend to reduce the rate of attack. However, it is the writer's understanding that baked carbon blocks are used for blast furnace linings and at the present time baked carbon shapes (of the blast furnace type) which will insure a good tight lining in the well of our cupola are not available. Consequently, we have used a rammed carbon plastic, but the manufacturers of this material state that under certain conditions fairly wide shrinkage cracks may develop in the carbon plastic and this, of course, would constitute a lining failure. Wide shrinkage cracks do not seem to develop in baked carbon blocks nor in magnesite brick and until such time as baked carbon cupola blocks are available we feel that it is safer to back the rammed carbon with magnesite brick.

3. It is known that magnesite brick spall quite seriously when subjected to temperature changes and such is the case when the cupola bottom doors are dropped. Under these conditions the protective layer of rammed carbon reduces the rate of cooling to which the magnesite brick are subjected and

thus materially reduces their tendency to spall.

4. Operation of our basic cupola with the combination of refractories described above has been fairly successful and on several occasions we have operated continuously for 168 hr. At present we melt iron for 16 to 19 hr per day for 5 or 6 days per week and bank the cupola during the hours when there is no demand for iron. The bottom is dropped only once a week. In spite of what has just been said the writer feels that there is still room for improvement in our refractory practice.

¹Chief Metallurgist, Lynchburg Foundry Co., Radford, Va.

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MR. KRANER: Mr. Levi's discussion leaves little on which to comment since we are in accord with his observations.

CO-CHAIRMAN CARTER (Written Discussion): The author has emphasized several important points we are in agreement with. Limited experimentation at our plant with carbon and graphite linings in combination with water cooling has suggested some possibilities for such a combination. About a year ago we lined a small cupola with graphite brick in the melting zone. This cupola is regularly water cooled with external sprays cooling a thin acid lining. The graphite brick were laid with no insulation and against the shell as closely as possible, as suggested by Mr. Kraner, in order to cool the graphite. This was not accomplished completely where irregular spots and riveted joints interfered. Where the lining was made up with tight joints and good contact with the cooled shell, the lining showed essentially no attack. Where it was not so tight some attack was evident at the joints. Lining failure was caused by the falling out of some of the brick that were not properly put in. The cupola was lined too hastily and brought into service too fast. Another trial with carbon blocks was planned but the opportunity has not been found.

During this operation both acid and basic slags were obtained by adjustment of the flux. Melting was fast, hot and clean, and satisfactory in every respect. The economics of such a water-cooled carbon lining would appear to depend greatly on the length of heat and the value of continuous operation.

The author has wisely suggested several refractory approaches toward maintaining a basic slag. Slag basicity does the work. The refractories simply contribute to or subtract from efforts toward that basicity. The most direct and conventional approach is the substitution of basic refractories. The use of neutral refractories like carbon is another approach. If acid refractory consumption can be sufficiently reduced with extensive water cooling, basic slags can be maintained in an acid-lined furnace. We have produced fairly basic slags in an acid lining without water cooling, but not for long.

As the author suggested, many blast furnace techniques suggest applications in basic cupolas. However, in the blast furnace the more reducing conditions and the larger diameter are more favorable than in the cupola. Carbon refractories withstand hypereutectic iron and reducing conditions better than lower carbon equivalent irons and oxidizing zones.

Overall costs will probably find one refractory combination best for one set of circumstance and another for other applications. The author is to be commended for his suggestions of several refractory approaches to this newly acquired ability to utilize cupola slag chemistry for iron refinement.

CO-CHAIRMAN CARTER (Oral Discussion): Our basic cupola experience verified Mr. Kraner's discussion of basic slag compositions. Decreased fluidity and desulphurizing efficiency has been experienced when basicity became very high. For that reason we have shown in our papers the use of small fluorspar additions to maintain slag fluidity and assist the desulphurizing reactions, whenever a high basicity is expected. Good fluidity is obtained from limestone without fluorspar when only moderately basic slags are obtained.

Our experience also verified Mr. Kraner's discussion of the relative effects of CaO and MgO. As long as the MgO re-

mained a relatively minor basic constituent, MgO seemed equally effective. However, as MgO becomes equal or greater than CaO the MgO appears to give lower fluidity and less efficient desulphurization. For that reason we try to select a flux and refractory combination that will give us as much CaO as possible.

MR. KRANER: In connection with Mr. Carter's remarks regarding basicity, desulphurization and viscosity, I would like to call attention to the work of Holbrook and Joseph of the University of Minnesota relative to blast furnace slags. This as well as some work recently done for Bethlehem Steel Co. which has not as yet been published but which soon will be indicates that if we maintain a ratio of lime to magnesia in the slag of approximately 3 to 1 we can vary the silica rather more widely than if we use only lime as a flux.

W. R. JAESCHKE: What is the proper formula for basicity ratio? Some formulae include Al₂O₈ along with the silica in CaO + MgO

the denominator as follows Al₂O₃ + SiO₂

while other basicity formulae omit the Al₂O₃. Which is correct? What is the effect of Al₂O₃?

MR. Kraner: I personally prefer to consider Al₂O₈ as playing a role similar to SiO₈ and include it in the denominator.

CO-CHAIRMAN CARTER: I am afraid I will have to disagree with Mr. Kraner on the effect of Al2O2 on basicity ratio as it effects sulphur in the cupola. From other standpoints it seems logical to consider Al2O3 as an acid and include it in the denominator of basicity ratio calculations. But from the standpoint of sulphur in the basic cupola we have not found Al₂O₂ nearly as detrimental to desulphurization as SiO2. In both basic cupola and basic electric operation we have accomplished very efficient desulphurization with slags rather high in Al2O3. Al₂O₃ seems to improve desulphurization if substituted for SiO₂. Desulphurization has been accomplished by the Perrin ladle slag process using lime-alumina slags. For these reasons we have preferred to leave Al2O3 out of the denominator for our basicity ratio until relationships can be more accurately determined. In our operation we think a ratio without AlgOg appears to correlate more accurately with desulphurizing effi-CaO + MgO

ciency than the formula Al₂O₈ + SiO₂

MR. Kraner: This is only for the purpose of maintaining a constant ratio which is desirable from an operating control point of view. The question of Al₂O₈ as a desulphurizer is a more complicated one, it is serving amphoterically and does not perform exactly as silica, nor does it function as effectively as lime or magnesia in solution in a liquid slag. Its effectiveness in a lime-alumina slag can not be compared directly with its performance in a blast furnace slag. In the latter type slag there fortunately is little variation in Al₂O₈ content. If there were a wide variation, then when the alumina is very low, the silica might need to be increased in order to maintain a fluid slag.

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# EFFECT OF BINDERS AND ADDITIVES ON THE HIGH TEMPERATURE PROPERTIES OF FOUNDRY SANDS

By

R. E. Morey,* C. G. Ackerlind* and W. S. Pellini**

#### ABSTRACT

In order to study sand deformation under load at elevated temperatures automatic stress-strain recording apparatus was adapted to a conventional sand testing dilatometer furnace. The previous work of the authors has been extended to include load-deformation measurements of three representative clays, and, to show the comparative effect to supplementary binders and additives, one clay was tested with six different supplementa y materials. Compacted foundry sand was found to follow Hooke's law and behave as an elastic solid at loads below the elastic limit. At higher loads plastic deformation or permanent set occurs. The general pattern of strength, modulus and deformation as a function of temperature is established by the nature of the clay binder used. Supplementary organic binders or other additives serve to modify the properties of the clay, some causing increases in properties while others tend to decrease the test results.

#### Introduction

The "engineering" of sand mixtures to provide for the elimination of sand defects is possible only to the extent of our knowledge of the properties of the materials and requirements of service. Unfortunately, our present knowledge is inadequate as indicated by the frequency of casting defects attributable to the sand. Notable among these defects are veins, penetration, scabs, buckles, rat-tails and erosion. These defects take a heavy penalty in expense, effort, and manpower.

In service, sands are subjected to the full range of temperature from ambient to approximately 3000 F. Information on high temperature properties is needed, particularly in respect to strength and ductility. Of these two only compressive strength has been investigated to any appreciable extent. Hot ductility has not been investigated inasmuch as the necessary equipment has not been available. The present investigation which represents an extension of the previously reported studies of hot strength was aimed at establishing the hot ductility characteristics of various sands. As a further contribution a general analysis was made of the nature of the temperature-induced physical changes which are responsible for

the specific levels of the strength and ductility properties.

As a first step in this investigation a method was evolved for obtaining automatically recorded loaddeformation curves by application of conventional O. S. Peters stress-strain recording equipment used in tension and compression testing of metals. A standard Dietert Dilatometer furnace was modified to permit the use of the recording equipment. Figure 1A and B shows a schematic diagram and photograph of the equipment. The measurements are recorded on a strip chart; typical stress-strain curves for sand specimens measuring 11/8 in. in diameter and 2 in. high are shown in Fig. 2. The equipment provides six ratios of strain magnification-200:1, 100:1, 50:1, 20: 1, 10:1 and 5:1. Strain magnification of 20:1 is suitable for most sand mixtures. The value of the load at full scale may be 2000, 1000, 200 or 100 lb. Since the specimen has an area very close to 1 sq in. (0.994) the load values may be taken as unit loads in pounds per square inch. The load scale is changed as necessary to suit the material and test temperature.

The typical load-deformation records of Fig. 2 show that the forms of the curves vary widely but that certain features are always present; Fig. 2D clearly illustrates the principal features. From A to B on the curve, the loading ram is rising but the top of the specimen has not made contact with the top compression head. A movement in the strain direction is therefore shown with no change in load. At point B the top of the specimen touches the upper compression head. From B to C "seating" occurs as the ends of the specimen, carborundum loading discs and compression heads are forced into intimate contact. When "seating" has been accomplished, as shown at point C, load and strain increase proportionally producing a straight line on the chart C to D indicating that sand mixtures behave as elastic solids and follow Hookes Law. At point D the proportional limit is reached and from D to E strain increases more rapidly than load, indicating plastic flow. At point E the ultimate compressive load is reached and from E to F instability is developed as indicated by a falling load. At point F a sudden failure occurs usually

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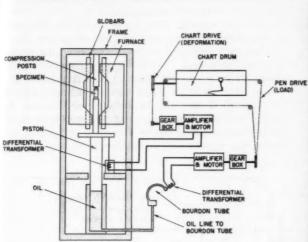
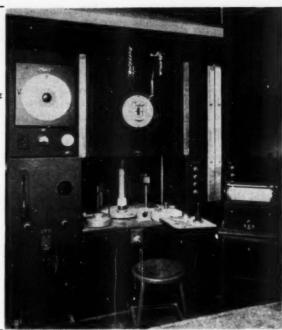


Fig. 1A (Left)—Schematic diagram of testing equipment. Fig. 1B (Right)—Photograph of testing equipment.



along 45-degree shear planes. In the case of very high ductilities the specimen may continue to deform after maximum load without shear failure until the limit of travel of the loading ram is reached. In these cases the specimen becomes barrel-shaped and continues to support a load. If the load is gradually released at a point such as D in the proportional range the unloading trace follows the line from D to C as should be expected of an elastic material. If the load is slowly released at some point beyond the

proportional limit such as point E, the unloading trace will follow a line parallel with CD but through point E. The horizontal distance between the two lines represents the permanent set or plastic deformation. If the specimen is tested to failure, as at point F, the plastic deformation is indicated by the distance L-M while the total deformation necessary to cause failure is indicated by the distance G-H. The strain required to seat the discs firmly (B-G) is not included in this measurement. The ultimate compressive load

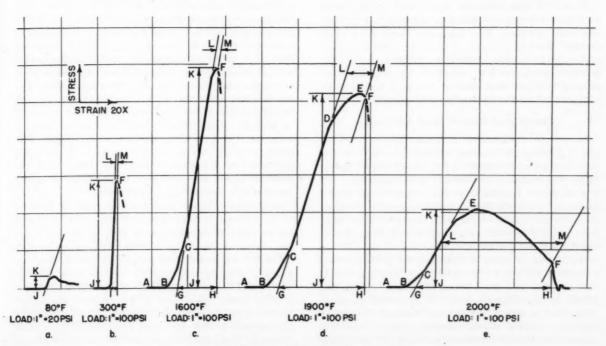


Fig. 2-Representative stress-strain curves.

is read as the distance from J to K. The elastic modulus is computed by drawing a straight line through the proportional section of the curve, C to D, selecting two points on this line and dividing the load difference between the two points by the strain difference

To assure reproducibility it is essential that the rate of loading be held within reasonable limits. In these tests the loading rate was such that, with a steel specimen between the compression posts, the load increased from 30 to 730 psi in approximately 15 sec. In all cases tests were made in duplicate and the resulting curves were found to be almost identical. In a few cases where results did not check satisfactorily, several check tests were made. The test values used are averages of two or more tests.

#### Materials and Methods

The test mixtures were made in 3000-gram batches using A.F.S. Secondary Standard Test Sand, binders and additives as indicated, and water to produce 3.8 per cent to 4.0 per cent moisture content. The materials were mixed 2 min dry and 8 min wet in a Simpson 18-in, laboratory muller, then passed thru a 1/4-in. screen and placed in sealed glass jars. Cylindrical specimens 11/8-in. in diameter by 2 in. long were prepared using 3 blows of the standard double-end rammer. Each specimen was placed in the furnace immediately following preparation and heated at the test temperature for 11 min, 45 sec. The loading mechanism was then started and the specimen usually contacted the top compression head in approximately 15 sec thus ensuring the standard total heating time of 12 min. No direct readings were made as all information necessary was automatically recorded on the stress-strain record.

A total of nine mixtures were tested to determine maximum compressive strength, deformation to rupture and modulus. For discussion these have been divided into two groups; (1) a group containing clay as the only binder, including Western bentonite, Southern bentonite and fire clay; (2) A group containing Western bentonite as the clay binder with additions of secondary binders and additives.

#### Properties of Clays Without Additives

In order to understand the nature of the reactions which produce the marked changes in the strength and ductility of the sands at the various test temperatures, it is necessary to consider the fundamental nature of clays. Inasmuch as the properties of quartz grains are relatively fixed or unchangeable,² the variations in properties of the sand aggregate must arise from physical changes occurring in thin films of clay which bind the grains together.

In the case of Western bentonite the principal constituent is sodium montmorillonite. This mineral is characterized by a plate-like crystalline structure, the plates being relatively large in area but small in thickness. As shown by Grim and Cuthbert³ this mineral has a strong affinity for water which is held as oriented molecular layers surrounding the individual plates of the crystal. As the water content increases

the number of layers increases. This assemblage of a clay crystal with surrounding layers of water is called a clay-water micelle. The water molecules adjacent to the plates are very strongly held while the bond holding each successive layer outward from the plates is increasingly weak. The binding force between clay plates decreases as the number of water layers increase. The strength of the clay accordingly should be expected to vary with the initial water content. It should also be expected that the elimination of water on heating should result in a decrease in spacing between clay crystals and an increase in strength.

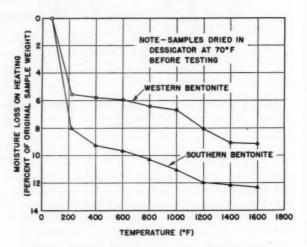


Fig. 3—Moisture loss as a function of temperature.

The effect of heat in removing water from clay is shown in Fig. 3. Samples of "as received" Western and Southern bentonite clay (no sand) were dried in a desiccator at room temperature to remove all water not strongly absorbed to the clay crystals. The samples were then placed in crucibles and held an adequate time at a series of successively higher temperatures with re-weighings following each temperature step to determine the loss of water with temperature. For Western bentonite there was a large moisture loss on heating to 220 F and a relatively small loss from 220 to 1000 F. From 1000 to 1400 F there is again a large loss. Above 1400 F the loss is very slight. The effect of water content on the strength of the sandclay mixture is consistent with the theory that the strength varies inversely with the number of water layers which separate the clay crystals.

It has been shown⁴ that green strength is lowered by an increase in water content above that required for an effective distribution of the clay during mulling. Figure 4 shows that strength of Western bentonite sand is increased consequent to a loss of water on heating. The marked strength increase from 6 psi at 100 F to 110 psi at 400 F is in agreement with the marked water loss shown in Fig. 3 over this range. On this same basis the small water loss observed from 220 to 1000 F and the larger loss from 1000 to 1400 F should indicate that there will not be much change in strength from 220 F to 1000 F and that there will

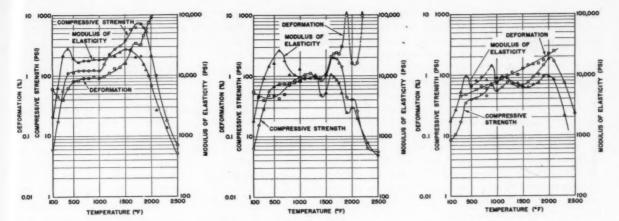


Fig. 4-Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite. Water content 3.8-4.0%.

Fig. 5-Effect of temperature on properties of a sand mixture bonded with 4% Southern bentonite. Water content 3.8-4.0%.

Fig. 6-Effect of temperature on properties of a sand mixture bonded with 8% fireclay. Water content 3.8-4.0%.

be a large increase in strength from 1000 to 1400 F. Examination of the strength curve in Fig. 4 shows that both of these expectations are fulfilled. From 1400 to 1750 F there is a strength increase due to the formation of viscous glass resulting from the melting of the clay crystal. From 1750 to 2500 F strength decreases rapidly as the viscosity of the glass is lowered by the increasing temperature. The final strength at 2500 F is less than 7 psi or about the same as the green strength and only 1 per cent of the strength at 1750 F. The modulus of elasticity, as seen in Fig. 4 roughly parallels the trends for compressive strength.

The total deformation before failure drop from 0.6 per cent in the green state at 100 F to 0.4 per cent at 300 F. From 300 to 500 F there is a sharp increase in deformation followed by a gradual increase to 0.9 per cent at 1000 F and then a rapid increase to 3.4 per cent at 1700 F. From 1700 to 1800 F there is a slight drop in deformation which will be seen to be characteristic of several mixtures, and from 1800 to 2000 F deformation increases rapidly to over 9 per cent. At higher temperatures the actual value of deformation could not be measured as the specimen continued to deform without an actual shear failure and there was no end point from which to make a measurement.

The similarity shown by the strength and ductility trends in the range of 300 F to 1700 F suggests that the improvement in ductility is related to the development of the more effective bond between sand grains. The extremely high ductility developed above 1700 F reflects the change in the clay from a crystalline material to a viscous glass.

Southern bentonite is quite similar in crystallographic structure to Western bentonite3 except that the sodium ions of the montmorillonite are replaced with calcium and hydrogen. It also adsorbs water which may be removed by high temperatures as shown by the dehydration curve for Southern bentonite in Fig. 3. The moisture loss is rapid up to 400 F, then gradual to completion at 1200 F.

The properties of a mixture containing 4 per cent Southern bentonite are shown in Fig. 5, from which it may be observed that the compressive strength and modulus of the Southern bentonite mixtures are low when compared with Western bentonite. The strength changes essentially parallel those of the Western bentonite except in the range from 400 to 1200 F where a gradual increase in strength may be seen as should be expected from the moisture loss in this temperature range. The deformation of the Southern bentonite mixture is roughly parallel to but slightly below that of the Western bentonite mixture. A marked difference between the clays is observed in the range of 1900 F to 2200 F. In this range Southern bentonite shows a minimum in ductility resulting in shear fractures which were not obtained in the case of Western bentonite. Corresponding with the minimum in the deformation curve at 2050 F, peaks are developed in the modulus and strength curves. The maximum in the strength curve at this point was more clearly shown by a larger number of tests in the previous investigation1 where the tests were made at much closer intervals of temperature. It is believed to be due to the presence of kaolin as an impurity.

The properties of an 8 per cent fire clay mixture are shown in Fig. 6. The principal constituent of fire clay is kaolin which is known to have a considerably lower capacity for adsorbing water than either Western or Southern bentonite. In general the properties of the 8 per cent fire clay mixture are lower than those of the 4 per cent Western bentonite mixture and the changes in properties with temperature are less marked. In the range from 2000 to 2500 F the compressive strength and modulus of the fire clay mixture exceeds that of the bentonite due to the refractory nature of kaolin.

The properties of the three clays may be compared directly in the summary graphs of Figs. 7 to 9. Figure 7 which summarizes the compressive strength curves shows that the maximum strength for Southern bentonite is reached at 1600 F, for Western ben-

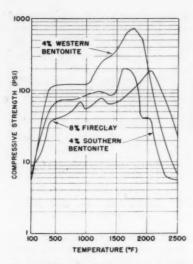


Fig. 7—Effect of the nature of the clay binder compressive strength at elevated temperatures.

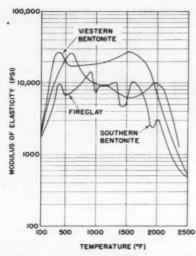


Fig. 8—Effect of the nature of the clay binder on modulus of elasticity at elevated temperatures.

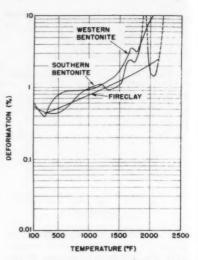


Fig. 9—Effect of the nature of the clay binder on deformation at elevated temperatures.

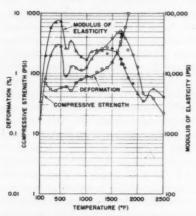


Fig. 10—Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite and 2% gelatinized corn flour. H₂O, 3.8-4.0.

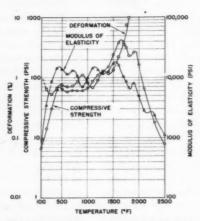


Fig. 11—Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite and 2% wood flour. Water content 3.8-4.0%.

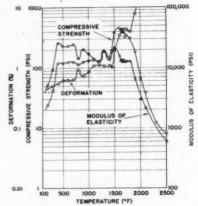


Fig. 12—Effect of temperature on properties of a sand mixture bonded wth 4% Western bentonite and 2% asbestos. Water content 3.8-4.0%.

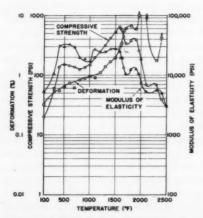


Fig. 13—Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite and 2% seacoal. Water content 3.8-4.0%.

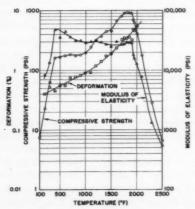


Fig. 14—Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite and 2% iron oxide. Water content 3.8-4.0%.

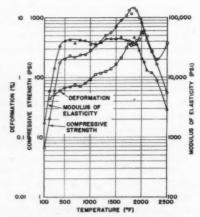


Fig. 15—Effect of temperature on properties of a sand mixture bonded with 4% Western bentonite and 10% silica flour. H₂O, 3.8-4.0.

tonite at 1750 F and for fire clay at 2050 F. In all cases the initial rise in strength from 100 to 400 F is due to elimination of water from the clay crystals. The drop in strength at temperatures above that producing maximum strength reflects the decreasing viscosity of the molten glass formed from the clay plus such silica as may have dissolved in the melt from the adjacent grains. No explanation has been found for the peaks in the fire clay curve at 900 and 1250 F.

The modulus curves for the three clays are shown in Fig. 8. As in the case of strength there is a rise in modulus during drying of the clay and a decrease in modulus as the clay melts and the glass becomes less viscous. Western bentonite and fire clay reach their initial maxima at 400 F, while 600 F was required to produce the initial maximum in Southern bentonite.

The deformation curves are summarized in Fig. 9. In general, Western bentonite had high deformation at failure while fire clay had low deformation. Western bentonite sands are highly plastic above 1900 F while Southern bentonite sands which are also highly plastic at 1900 F revert to low ductility at 2000 to 2100 F, then become plastic again at 2200 F. In this range the fire clay mixtures were brittle by comparison.

## Supplementary Binders and Additives

The effect of six supplementary binders or additives on the properties of the Western bentonite reference mixture are shown in Figs. 10 through 15. The reference mixture contained 4 per cent Western bentonite and 3.8 to 4.0 per cent water. The modified mixtures contained 2 per cent of the supplemental material except for silica flour in which case the addition was 10 per cent.

Figure 10 shows the properties of a mixture containing 2 per cent gelatinized corn flour. This mixture develops high strength in the range of 360 to 500 F due to the formation of a paste or glue which becomes increasingly rigid as drying progresses. At temperatures above 500 F, the starch molecules decompose to form sugar and liberate water with consequent loss of the glue effect and the strength of the mixture drops at 600 F. In the range of 600 to 700 F the sugar derived from the starch, is meited or caramelized. This material, when first melted, is viscous and serves to increase strength, producing the peaks shown at 700 to 750 F. Above 750 F the viscosity decreases causing a reduction in strength. The sugar further breaks down producing carbon and water vapor. The strength of the mixture from about 900 F upward, therefore, is essentially that of the clay. At 2300 F the possibility of a strength peak is seen which does not appear in Fig. 4. This effect is believed to be due to a fluxing action of the ash left from the combustion of corn flour,

The curve for modulus parallels the strength curve and corresponding peaks may be seen. Deformation is quite similar to that of the 4 per cent bentonite mixture except in the range from 1700 to 1800 F where deformation increases for the corn flour mixture but decreases in the 4 per cent bentonite mixture.

Figure 11 shows the properties of a mixture containing 2 per cent wood flour. Compressive strength and modulus are considerably lower than in the case of gelatinized corn flour. Peaks at 450 F and 750 F may be seen particularly in the modulus curve. This is to be expected because the wood flour contains a large percentage of cellulose which is chemically very similar to starch. The sugars produced from the break-down of cellulose, carametize to produ e the strength maximum at 750 F. From 1000 to 1600 F the shape of the strength curve is essentially that of Western bentonite but it is lower in magnitude possibly due to voids left in the clay bond when the particles of wood flour burn out. There are well defined peaks in strength and modulus at 1960 F and in modulus at 2250 F. As in the case of the gelatinized corn flour, these peaks are probably due to a fluxing action of the ash left by the wood flour. Deformation is similar to that of the 4 per cent Western bentonite mixture except at temperatures of 1700 F and higher where deformation is much higher in the wood flour mixture. This high deformation in wood flour mixtures at temperatures near 1700 F which the reference mixture does not have may explain why wood flour mixtures do not crack and therefore prevent "vein" defects in phosphor bronze castings.5

Short fiber asbestos has recently been recommended as a "cushioning ingredient" in sand mixtures to prevent scabbing and buckling on the theory that the soft inter-granular material permits the expansion of quartz occurring at the alpha to beta inversion to take place without cracking. The properties of a mixture containing 2 per cent asbestos are shown in Fig. 12. By comparing Figs. 4 and 12 it may be seen that the deformation for the two mixtures are similar. The asbestos caused both compressive strength and modulus to be lower at temperatures above 1100 F but did not have any significant effect at lower temperatures.

The properties of a sand mixture containing 2 per cent seacoal are shown in Fig. 13. Three high temperature strength peaks were observed at 1575, 1900 and 2300 F with corresponding peaks in the modulus. As in the case of Southern bentonite, the strength maximum at 2300 F was accompanied by low deformation and shear failure while specimens at both lower and higher temperatures deformed plastically with deformation so high that it could not be measured. When loading was stopped, in the tests at 2100 F, the specimen still supported a small load even though it was reduced to 75 per cent of it original length.

Additions of 2 per cent of iron oxide developed the properties shown in Fig. 14. Compressive strength and modulus were increased considerably through most of the temperature range. Deformation, however, was changed only slightly by the addition of the iron oxide. In this maxture there is only one strength peak and the decline in strength from 1900 to 2500 F is continuous as in the case of the reference mixture.

The properties of a typical foundry mixture of 10

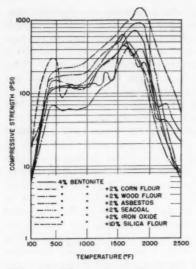


Fig. 16—Effect of supplementary binders or additives on strength of the reference mixture bonded with 4% Western bentonite.

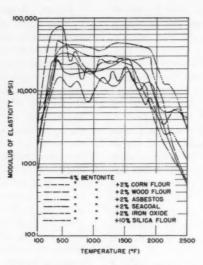


Fig. 17—Effect of supplementary binders or additives on modulus of elasticity of the reference mixture bonded with 4% Western bentonte.

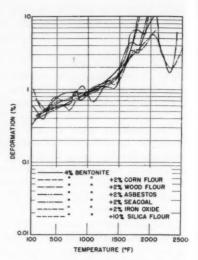


Fig. 18—Effect of supplementary binders or additives on deformation of the reference mixture bonded with 4% Western bentonite.

per cent 325-mesh silica flour are shown in Fig. 15. The strength and modulus of this mixture are extremely high at all temperatures. The deformation curve follows that of the reference mixture quite closely up to 1800 F but from 1900 to 2500 F is much lower. In general, the silica flour-Western bentonite mixtures have a high strength over a wide temperature range and also good deformation characteristics.

In order to facilitate comparison of the various supplementary binders and additives, the strength curves for the six materials studied are summarized in Fig. 16 along with the strength curve for the reference mixture. It may be seen that the general shape of the curve is established by the Western bentonite and that the added materials merely serve to modify the strength level. For increasing strength over the whole temperature range, iron oxide and silica flour are effective. Silica flour is especially effective above 2200 F and in the range from 100 to 600 F gelatinized corn flour is very effective. Two materials, corn flour and seacoal, produce a high temperature peak at 2250 F. Corn flour and asbestos lower the temperature at which the strength maximum ocurs. Wood flour lowers strength by a large factor.

Figure 17 summarizes the curves for modulus of elasticity. The general pattern is set by the Western bentonite in the mixture and, as in the case of strength, the additives merely serve to modify the modulus established by the bentonite. In general silica flour and iron oxide increase modulus while wood flour decreases it. Gelatinized corn flour is effective in raising modulus in the range from 100 to 500 F and asbestos lowers it from 1000 to 2500 F.

Figure 18 summarizes the deformation curves. Below 1500 F the spread between curves for the various materials is narrow compared with the spread in strength and modulus. This indicates that supplementary binders have a minor effect on deformation at temperatures below 1500 F. At 1800 F wood flour and corn flour produce relatively high deformation. Seacoal and silica flour reduce deformation in the range from 2200 to 2400 F.

Sosman² reports a temperature of 1063 F for the alpha to beta inversion of quartz and indicates that the total linear expansion of a quartz aggregate from 32 F to a temperature above the inversion point is 1.32 per cent. It is interesting to note that at 1070 F (above the inversion) in Fig. 18, the deformation measured at failure ranged from 0.67 to 1.06 per cent. Inasmuch as a rapid expansion occurs at the inversion point and the total possible deformation is roughly of the same magnitude as the expansion value it seems likely that mold conditions may easily become critical for the development of various forms of sand failures. The sudden application of heat during pouring, mold wall restraint and other similar factors may be expected to provide the unfavorable conditions leading to crushing or buckling.

#### Conclusions

1. Foundry sands behave as common elastic-plastic solids to the highest temperatures of service. On loading the response is initially elastic with stress being proportional to strain within the elastic limit. At loads above the proportional limit permanent plastic deformation is developed. As loading is continued a limit is reached at which point instability occurs with consequent decrease in load. Depending on the inherent ductility, shear failures or "upsetting" without failure follows the development of instability.

2. Ultimate compressive strengths varied from 5.5 to 1430 psi, low values being developed at low and at very high temperatures and high values at intermediate temperatures. Maximum strengths occurred

at temperatures in the range of 1600 to 2050 F depending on the binders used.

3. Variations in modulus of elasticity essentially parallel strength variations. Modulus values ranged from 480 to 74,500.

4. Ductilities measured as compressive deformation varied from 0.4 to over 10 per cent. In general ductility increased with temperature.

5. The nature of the clay used as a primary sand binder establishes the general shape or pattern of the curves of strength, modulus and deformation as a function of temperature.

6. Secondary binders or additives control the level of strength and modulus over the entire temperature range. The effect on the level of deformation is not marked except at temperatures above 1500 F.

7. Silica flour and iron oxide are the most effective materials for augmenting strength and modulus over a wide temperature range. Gelatinized corn flour additions give high strength and modulus at temperatures below 600 F.

8. Wood flour additions reduce strength and modulus but increase deformation at temperatures above

9. Explanations based on the known chemical characteristics of the sands or additives have been advanced to explain major changes in properties as a function of temperature. The factors considered included dehydration of the clay structure which re-

sults in bringing the crystal plates of the clay closer together thus producing stronger bonds, melting of binder constituents such as sugars derived from starch or cellulose, fluxing due to the ash of organic materials, and glass formation due to melting of the clay and silica grains. In the case of melting or fluxing there is usually an initial increase in strength as a viscous glass is formed followed by loss in strength on heating to higher temperatures as viscosity decreases.

# Acknowledgment

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# Bibliography

- 1. R. E. Morey and C. G. Ackerlind, "A Study of the Effect of Varicus Binders and Additives on the Hot Strength of Molding Sands," Transactions, A.F.S., vol. 58, pp. 411-422 (1950).
- ing Sands," Transactions, A.F.S., vol. 58, pp. 411-422 (1950).

  2. R. B. Sosman, "The Properties of Silica," Chemical Catalog Company, 1927.
- 3. R. E. Grim and F. L. Cuthbert, "The Bonding Action of Clays, Part 1 Clays in Green Molding Sand," University of Illinois.
- 4. R. E. Morey and H. F. Taylor, "Some Properties of Synthetically Bonded Steel Molding Sands," Transactions, A.F.S., vol. 49, pp. 388-426 (1941).
- 5. R. E. Morey and J. R. Kattus, "Preventing Veining and Penetration on Castings Made in Synthetic Sands," Transactions, A.F.A., vol. 54, pp. 129-140 (1946).

# RESPONSIBILITY OF TOP MANAGEMENT IN TRAINING

By J. D. Judge*

There is inherent in human nature a desire for expression. We like to advance our ideas. We feel the urge to make things, to perform some service, to fulfill some need, to accomplish some objective. We feel strongly, and at times wrongly, that our way is the best way; we often think, and frequently rightly, that we could do a given task better than it is being done. This tendency to manifest our own will is a basic motivation. Uncontrolled, such a trait leads to despotism. Intelligently directed, however, it is the basis of all freedom, it is the motivating force responsible for all progress in the world.

These natural creative sentiments which we possess, coupled with the expectation of material reward, are responsible for the foundation of all business ventures. Most such enterprises are initially small. Time and talent, determine their growth and fix their span of years. If they have been wrongly conceived and poorly planned productivewise, marketwise, or profitwise, they soon wither away. If organizationally they are unsound they will eventually succumb. If, however, they are sound of product, efficient of manufacturing, astute of marketing, and enlightened of management, they will not only enjoy longevity but frequently will expand many fold.

# Leadership Plays Decisive Role

In any business, assuming of course a need for the product or service, leadership plays the decisive role in its success. This leadership may be embodied in an individual, who himself is one of the half-dozen workers in his plant, or it may be vested in a group which directs the activities of one of our modern industrial giants. In either event, and through all of the gradations of size and type, there is in every business some individual or some group, shouldering the responsibility of leadership; someone or some group whose duty it is to give final answers to questions of major importance.

We loosely classify such individuals under the gen-

eral term management, although actually the formulation of matters of policy is more accurately an administrative duty, and the execution and direction of such policies a management function. Because of the characteristics of the businesses comprising our industry, particularly as to size, we shall consider top management as that individual or group actively engaged in the business and charged with the responsibility of formulating and directing the execution of matters of major policy.

In most every business today, with an enlightened top management—and remember again that that term includes the owner-worker in the small establishment—one of the matters of major policy is training! Just what do we mean by training? What is its place in industry and particularly in the foundry business?

#### Training Defined

According to Webster, to train means to educate, to form by instruction and practice, to impart proficiency, to prepare. Training for all of us begins long before we reach the age of reason. Formal training is initiated, on a general plane, in the grade school. Directionally, it is channeled somewhat in high school, particularly in the vocational and trade schools. In college, training is usually directed toward some specific objective. When this academic phase of training has been completed as a major projectand this may occur at some period during the grade school term, or after years of college and post-graduate work—the second phase of training begins. This phase which constitutes the continuous application of accumulated knowledge and experience, and the acquisition of skills in a particular field, is a lifetime undertaking. There are many fields of application, but we are concerned with the industrial area and specifically with the foundry business.

As we said, businesses usually start in a small way and then develop in accordance with a variety of circumstances. In the simplest type of organization there are two classifications, the owner and the workers. With some expansion, a new classification is added—supervision. Increased activity calls for another classification—clerical help; and large scale operations

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Conference between top management and younger supervisors, whether the plant is large or small, are part of the training which is management's responsibility.



necessitates technical specialists. These five classifications, management, supervision, technical specialists, office or clerical, and shop workers, represent a skeletonization of the organizational constitution of most businesses. Within each of these classifications there must be planned and continuous training if success is to be achieved.

# Important Training Objectives

There are at least three important objectives for training within these classifications. The first is concerned generally with those individuals in the semi-skilled and skilled group. The aim is to help them to become more dextrous—more skillful—and to increase their overall efficiency. These ends are achieved through a variety of on-and-off-the-job formal training programs or through informal in-plant instruction.

A second objective of industrial training is directed toward those individuals who form part of the managerial hierarchy—the foreman and supervisors, the technical staff, and the junior executives. Here again the purpose is to develop individuals so as to enable them to become more proficient and to prepare them for the assumption of greater responsibilities. The training field to be covered for this group is necessarily diverse, including as it does such subjects as labor relations and regulations, plant policies, social legislation, time and motion study, waste prevention, industrial economics and administrative procedures. The usual aids to these ends are extension courses, executive conferences, discussion groups, plant visitations, technical society meetings, trade magazines and so on.

The two training objectives just outlined are concerned with the individual, principally because of his actual and potential contribution to the success of the business. The third objective of training differs somewhat from these, in that it is concerned in large measure with the individual because of himself and society. It is universal in its purpose, applying to the laborer as well as to the executive. The aim of such training is to make the individual a better man—

physically and mentally, materially and spiritually, and as a corollary to make him a better member of his community and a better citizen of his country. Since these matters lie almost exclusively in the field of human relations, the means to their accomplishments are varied. Plant letters, bulletins and house organs, counseling service, health and hygiene programs, and company sponsored activities, are a few of the more common means to this end.

Training then has for its purpose not only the development of abilities and skills, but it has a wider implication in that it is instrumental in stimulating and arousing desires for self improvement not only in the manual field but the intellectual as well, not only in the realm of performance but that of planning also, not only in the role of craftsmanship but in that of leadership.

# Training Insures Continuation of Business

All of this leads naturally to the question, why should any business be interested in promoting the all around development of its employees? Fundamentally and basically, why should business be interested in training? The answer is not simply to make more efficient employees, so that the more economical operation resulting therefrom, will be reflected in greater profit. Neither is the answer in the social benefits which accrue to the individual through training. Hypothetically, if the business were to stop operating at the end of the week, would any training be necessary? If the operating period were projected to a month, would a training program be required? If, however, the business is to continue for a year, for ten years, for one-hundred years, does training then become a part of the business? The answer is obviously in the affirmative. Basically then, the reason for training, is to insure the continuation of the business.

Why should the continuance of any business be important? The answer for some companies, because of their nature, is clear. They have a weighty responsibility for perpetuity toward the general public. Utilities, financial institutions and insurance companies

are in this class. Many other corporations, whose stock is widely held, have a very real responsibility toward their investors. Even in the smaller companies, the need for a continuance of family income can be the important factor.

#### Reasons for Continuance of Business

In addition to these primary reasons, there are also secondary reasons for the continuance of business enterprises. As we know, most businesses start on a small scale. They embody the ideals, the hopes, the desires, and the faith of an individual or a small group. Such a personal undertaking is a creative accomplishment, and as such it evokes those natural instincts of preservation. The father wants to see his sons carry on in the business to which he has given so much of himself. He wants them to taste the fruit of accomplishment; to reap the reward of hard effort. He has a justifiable pride in this business which his management has developed.

Although admittedly on a much less personal basis, the pride of accomplishment, also exists in the larger organizations, even among those individuals who have come into established businesses. It is natural for a person to want to see the continuation of a concern which has incorporated in it many of his ideas and plans, and to which he has given so much of his

time and effort.

There is also a third consideration for the continuation of any established business, and that is the personnel of the plant. There are many employees who build their lives around their jobs. They establish their homes, raise their families, predicate their hopes, their aspirations, on the security of their employment. These human considerations should not and cannot be ignored.

We have reviewed some of the reasons prompting the initiation of a business; we have defined top management; we have sketched the organizational makeup of business; we have discussed the objectives and the reason for training, and now we shall briefly consider the responsibilities of top management in this

matter of training.

We said that training begins in the grade school, is given direction in high school and objectivity in college. It follows from this that top management should evidence a real and an active interest in the curriculum of the grade schools, through teacher contacts and through the boards of education. Top management should be even more interested in the high schools, particularly in the trade and vocational schools. Supplying part-time instructors and equipment for shop courses; promoting plant visitations; providing after school and summer work, and encouraging personal consultations, are but a few of the means of evidencing this interest. There is a profound conviction on the part of many foundrymen that the high school level provides the most promising area for stirring the imagination, awakening an interest, providing a challenge, and stimulating a desire, in young men in the foundry business. Most of these students are undecided as to their future plans. Far too many of these young men leave high school with only the haziest kind of a notion of the foundry business, with no idea as to it operation, its vital place in industry, or its opportunities. At the college level, top management is provided with a unique and well organized program in the Foundry Educational Foundation, which needs only its continued material and active support, to successfully carry on this outstanding work. The Foundation as you know, has been responsible for a reactivation, a reorientation, on the part of many universities and colleges of their class and laboratory courses, in foundry and allied subjects. This together with the cooperative and summer work provided by the member foundries, has resulted and will continue to result in a pool of young engineers with both academic training and practical experience in the foundry business-young men from whose ranks should come the managers of tomorrow.

Practically all knowledge that we possess has come to us from others, consequently in simple justice we all have a responsibility to pass on, that which we have inherited. This is not only ethically sound, it is



Plant visitations, technical society meetings, extension courses, along with executive conferences are training aids management should use and encourage. VIVG

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also good business. Efficient and profitable operation, depends upon the maintenance of an organization skilled not only in general fundamentals but also trained in details peculiar to a specific organization. This applies not only to manual operations, but also to all other functions of the business. Top management, accordingly, has the definite responsibility of formulating policies and delegating or directing their execution so that this basic objective of instructing others in all aspects of the business will be accomplished.

# Reduction of Labor Turnover

Labor turnover occurs in every business. Top management is familiar with this fact, and again, it is its duty to formulate general policies relative to this problem, particularly as it applies to the semi-skilled and skilled employees, to the end that through proper training, with all its implications, turnover will be reduced, or its effect upon the efficiency of operation of the business when it does occur will be a minimum. Top management should be much more directly concerned with those positions in the organization of greater responsibility—the foremen, the supervisors, the technical and sales personnel, and the executive positions. Management has the responsibility of facing the facts of life, of realizing that key personnel are constantly attracted by better opportunities, real or fancied, in other organizations; that accidents, illhealth, retirement, and death pay their inexorable visits with impartial regularity to all. Available statistics provide management with a guide to intelligent planning in these matters. There is no greater responsibility for management than to plan for upgrading and replacements in the managerial hierarchy. Training should be fashioned not only to care for the unexpected contingiency, but to prepare for the one, the five, the ten-year future. To this latter end recruitment of young men is essential. Men of vigor and enthusiasm, men with fresh and modern conceptions of materials, of methods and of men. Men who can be trained in the ways of the organization; who can be a force in overcoming any tendency toward stagnation; in short men who can keep the business young mentally.

Management, therefore, in order to insure the continued efficient operation of the business must formulate a policy of planned development of skills and abilities at all levels within the organization, and must intelligently approach the problem of personnel upgrading and replacement. Essential though they are, these policies to be fully effective must be augmented—they must be coordinated in the field of human relations.

#### Social and Economic Phenomenon

It has been said that industry is a social as well as an economic phenomenon. It is made up of individuals who truly must earn their daily bread, but who in addition have ideas, beliefs, and convictions; desires, hope, and plans. These aspirations, these cravings, these needs must be recognized and must be satisfied to some extent, if an approach to personal

peace and contentment is to be attained and reflected in organizational cooperation, efficiency and success. Enlightened management can go far toward the discharge of this responsibility through sincerity and understanding. Frank, unbiased explanations and discussions of union negotiations, of the pension and retirement fund, of the health and accident insurance, of profit sharing, of incentive plans, of safety and hygiene, of community projects, of national issues, in short, of any major plant policy or outside occurrence which intimately affects the employees or the company, is educational, it is instructive, and it is practical training of the mind.

Now, let us briefly evaluate these responsibilities of top-management in this matter of training, as they apply to our own industry, the foundry business. As of October, 1951* there were in the United States 5,299 foundries of all kinds, captive and jobbing, ferrous and non-ferrous. Fifty-three, that is only 1 per cent of the total foundries employ 1,000 or more persons; 21 per cent employ 100 or more; 38 per cent employ 20 or more; and 41 per cent employ less than 20. In other words, 80 per cent of the foundries in this country employ less than 100 persons.

# Justification for Training

It is evident that ours is an industry with a large percentage of small businesses. Naturally, the question arises, is training necessary in these small foundries? How can it be justified? We have indicated that training is essential in any business, regardless of size, if it is to continue for any length of time as a profitable concern. Consider the small foundry with a single supervisor. Such an individual has a diversity of duties. He must oversee the melting and molding, the coremaking and cleaning, the shipping and storage. He is practically indispensible for the continued operation of the business. Training of a replacement for such a man is imperative.

Most larger foundries recognize the need for training and have such programs. There exists, however, in some instances a complacency in regard to supervisory and managerial replacements which is hazardous for the organization. As an example, top management of many foundries in its endeavor to further the educational advantages of its supervisory force, frequently sends groups to neighboring cities for various types of meetings. Does top management ever consider the status of the business, if group disaster should overtake such a trip? Does the training program provide for understudies suitably prepared for advancement? These are real managerial responsibilities!

There is just one other aspect of this training problem which should be briefly considered, and that is the matter of obtaining replacements from the field rather than from within the organization. There is no company which has not at times engaged in such practice, either purposely or through necessity. Such procedure is justified at times on the basis of limited

^{*} Marketing Guide, Penton Publishing Co., Cleveland.

capabilities within the organization, of outmoded outlook due to organizational inbreeding, of inordinate depletion of personnel, of unforseen expansion,

or major procedural changes.

Determining the extent to which such practice should be followed, in order to maintain the proper balance of morale within the organization, is one of the more delicate responsibilities of judgment for top management. However, regardless of the extent to which this policy is pursued, management cannot by this means honestly escape the responsibility of training. We are dependent beings. We need one another. Inordinate selfishness should have no place in our organizational thinking. Management should feel that if trained personnel are obtained from the industry, that it—management—has, in turn, the obligation to train others, primarily, of course, for its own organization, but in the event that circumstances direct future changes, for the industry.

This axiom that anything which benefits the industry as a whole will also result in a benefit to the individual members, is a truth which needs constant emphasis. It is a firm belief in this truth which should prompt top management to look beyond the immediate and specific return for its own business and to support for example such programs as the Foundry Educational Program, technical and trade societies, research projects, and other undertakings of equally general benefit to the industry. Such an attitude is in truth real enlightened management!

This industry of ours, even with its preponderance of small units, employes roughly one-half million people who produce annually more than 22 million tons of castings. These castings are the very foundation of our industrial structure. Upon them is built in large measure all other industry. Truly we can be proud of our contribution to the progress of our country and the world. Equally true is the realization that we must recognize the tremendous responsibility which rests upon the industry for the continuance of such progress and the responsibility which rests upon top management within the industry for such continued contribution through training.

# A.F.S. SAFETY, HYGIENE, AND AIR POLLUTION PROGRAM

By

James R. Allan*

We shall review what the foundry industry has done in the past two decades in trying to live up to the slogan that "the Foundry is a Good Place to Work." We shall also discuss briefly the problems that lie ahead if the industry is to accept the definite trends, and the urging of an enlightened public toward better public relations, freedom from severe regulatory control of working and operating conditions, and the need to keep abreast of the advancement of the knowledge developed by industrial medicine.

## Silicosis Problem of Early 1930's

Many of you remember the silicosis problem of the early 1930's during the depression period. Thousands of workers were laid off because of no work. Many workers, out of work, decided that they could prove occupational disease compensation cases, because of long exposure to silica dust. Law suits were numerous and the total of all cases reached a staggering sum of money.

Most foundries were particularly vulnerable and few managements knew the fundamentals of the cause and effect of pneumonosis and specifically sili-

#### A.F.S. Early Efforts to Help Industry

To bring light on the silicosis situation in those days, and to help the industry with recommendations to improve their working conditions and to work with regulatory authorities, the American Foundrymen's Society organized a Safety and Hygiene Committee comprised of a number of engineers and others from industry who had a very good understanding of the hazards and the measures necessary for improving working conditions.

Among the important accomplishments of this Committee were the drafting and publication of a number of recommended good practice codes, as follows:

Code of Recommended Good Practices for Grinding, Polishing, and Bussing Equipment Sanitation Code of Recommended Practices for Testing and Measuring Air Flow in Exhaust Systems

Fundamentals of Design, Construction, Operation, and Maintenance of Exhaust Systems

Code of Recommended Good Practices for Metal Cleaning Sanitation

Recommended Good Safety Practices for the Protection of Workers in Foundries

Code of Recommended Practices for Industrial Housekeeping and Sanitation

During World War II, the Committee prepared a code of Recommended Practices for the Protection of Life, Property, and Production in Industry during the War.

This latter code was adopted by the Chicago metropolitan area and many other urban areas as rules to safeguard all industry during the war.

The Foundry Industry benefited quite generally from these recommended practice codes as it gave the engineers and management some very definite engineering rules to apply on many of the problems of health and safety. The work was helpful to the foundry equipment industry as many of the recommendations were incorporated into equipment design or were otherwise provided for.

All of the code work having to do with foundry operations has been pretty generally accepted by the major industrial states as a basis for regulatory code rules. This is particularly true in Illinois. These codes have received world-wide recognition as a guide in foundry and allied operations.

All of this was accomplished by the volunteer services of technicians from the foundry and allied industries. The code work cost the Foundry Industry nothing except a tremendous amount of time of some few who served on the Committee. The benefits that the industry has derived from this early beginning, trying to set up reasonable working standards, have been great.

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# Emphasis on Safety and Hygiene for Future

The health and safety work of the Committee began to lag because the various members who had contributed so unstintingly of their time and experience could no longer spare the time because of involvement in their respective businesses. The health and safety activity of the American Foundrymen's Society ceased, but the need for aggressive work and leadership in the field of industrial safety continued to grow because of newer problems and a better understanding of the old problems, and also because of greater knowledge gained from experience in the field of industrial medicine.

Many of us have realized that some agency should be set up to carry on where we left off in the middle 1940's. There has been a "let George do it" attitude for some time, until the National Castings Council made a careful study of the whole situation and reached the conclusion that the American Foundrymen's Society as the technical society for the castings industry was the one and only agency that could reactivate the safety and hygiene program, based on their past successful efforts, and carry on for the future.

The safety and health program for the future must be on a planned basis and must be administered and steered for the benefit of the whole industry. It cannot be a hit and miss proposition. It has to be a continuing active program. This is going to take the full-time efforts of qualified individuals as members of the American Foundrymen's Society technical staff to supervise the program.

The American Foundrymen's Society Board of Directors approved a resolution in order to prescribe the functions of the activity as follows:

"The Safety and Hygiene and Air Pollution Program of A.F.S. shall be devoted to research and the accumulation of engineering data for the development of recommended practices in the field of Foundry Safety, Hygiene and Air Pollution."

I further quote from the prospectus of the Society's published brochure as follows:

"The American Foundrymen's Society does not intend to undertake the writing of regulatory codes nor to endeavor to influence legislation favorable to the industry; but rather, it will assemble authoritative data intended to assist foundries, lawmakers and others in developing reasonable standards of safe, practical working conditions. It is a governing principle of A.F.S. to 'advocate nothing but analyze all, selecting for dissemination only that which meets accepted high standards'."

Part of the program will be the compiling of a reference library including regulatory code publications and miscellaneous data. Available data will be correlated and, where lacking, limited research will be undertaken under definite contracts with patent protection for the membership. Existing research facilities will be employed where practicable. It will be a program policy to consider aspects of industrial medicine as well as engineering standards.

The program has been definitely put in motion even before the full amount of the funds necessary for its continuance have been pledged because of the urgency and needs now confronting the Foundry Industry.

Work is already under way in the American Foundrymen's Society in reviewing and to determine the need for any revisions of the past recommended good practice codes. All of this activity will be guided within the limits set by the American Foundrymen's Society Board of Directors and an appointed Safety and Hygiene and Air Pollution Steering Committee.

Some of the projects and good practice codes already listed for attention are:

- (a) Woodworking
- (b) Exhaust Systems for Electric Melting Furnaces
- (c) Welding code, to include welding and generating plants for acetylene and oxygen
- (d) Acoustical treating of work areas to minimize noise
- (e) Control of external air pollution
- (f) Ventilation requirements for sand handling systems and shake-outs
- (g) Ventilation of non-ferrous melting furnaces
- (h) Ventilation of the general working area

In addition to this, there is need for careful study to be given to the question as to the type and magnitude of foundry stack emissions, efficient dust and fume collectors for furnace emissions, and the influence of the height and aerodynamics of stack effluents.

The so-called Welding Code is one of the most important items because in Illinois we are looking for some guidance in a welding code that must be adopted in that state, having the force and effect of law. This will have to be presented to the Illinois Industrial Commission this Summer (1952) for review. The basic work on this code is being done by others outside of the Foundry Industry, but we are looking to the Foundry Industry for helpful suggestions because they are greatly involved. Unfortunately the time is too short for the Foundry Industry to really render all of the help that they could if the new technical setup had been created a little sooner.

Other similar problems will come up faster than we think. Industry needs guidance but that guidance should come from our industry—not from other agencies that may not understand our problems too

The American Foundrymen's Society last Fall (1951) started a drive with its membership, to raise funds to make this Health and Safety program an actual fact, and to be sustained over a long period of time—not less than ten years. It is conservatively estimated by those having to do with setting up this program, that a minimum of \$350,000 is needed for this long-range program. It needs and must have your financial support because the American Foundrymen's Society is the one agency that has done most in the field of health and safety to date for the Foundry Industry that has been of real practical benefit.

The American Foundrymen's Society will receive

much counsel and guidance in their program from those who have been long associated with it in the past. The need is real and urgent to reactivate and carry on in the interest of health and safety and air pollution control work. We can only point out that if the American Foundrymen's Society does not carry on, some other agencies will do the job for us, possibly in a way that the industry will not like—if we do not all cooperate and assume our leadership in this important work.

# New Problems Confronting Industry

Until recently, you have been concerned in your business with safety and health problems inside your buildings. Some of you have done a splendid job in improving conditions and trying to prove that "the Foundry is a Good Place to Work." Some others unfortunately have not done as much to improve conditions as they should have, but in the meantime an enlightened public is insisting that attention be paid to industry's effect on the surrounding neighborhoods in which they are located.

Much is being said in the public press today about unhealthy and dirty atmospheric conditions created by all types of industry in various localities and neighborhoods. Legal action is being taken to force industries to clean up the outside air so that the public can enjoy the surrounding area, free of obnoxious gases, odors and dust.

The various city so-called "smoke ordinances" are being expanded to include all types of contaminants, hence the foundry is particularly vulnerable.

Do not forget this, that any owner of a piece of property adjacent to an industrial plant can, by means of law, stop by injunction any plant from covering his property with dust or dirt, fumes, gases, and even stop excessive noise, like the unloading of pig iron and scrap, if it interferes with his normal rest. It makes no difference if the plant was there 50 years and the complaining neighbor just moved in next door; the burden of cleaning up the outside situation is up to the plant.

I am informed that in certain cities some foundries already are defending damage suits because of core room odors, and dust, dirt, smoke and fumes from cupolas and effluent from poorly designed dust arrester systems, on the basis that it is damaging to homes and property.

There is pending in the Senate of the United States Joint Resolution 110:

"To provide for intensified research into the causes, hazards, and effects of air pollution, into methods for its prevention and control and for recovery of critical materials from atmospheric contaminants, and for other purposes. "Whereas certain industrial processes, the use of certain fuels, and other activities have caused serious air pollution in many areas of the United States; and

"Whereas deaths, serious illness and suffering from air pollution strike not only industrial workers but their families as well; and

"Whereas air pollution is causing many deaths and illnesses among the people of the United States and extensive damage to public, industrial, and residential structures, to clothing and personal property, agriculture, forests, livestock and to other property; etc."

Recently certain state legislatures have been giving consideration to creating commissions to study the problem of air pollution.

The rather recent Denora, Pa., incident and the Los Angeles radical approach to cleaning the air of contaminants have spark plugged the drive throughout the country for relief from air pollution.

Many of us have been giving a lot of attention to the problem of the abatement of air pollution, but I can tell you that we do not have too many answers yet. Research is going on in some of the equipment companies in an attempt to find a practical and not too expensive way of capturing cupola dust, smoke and fume. At the present time a large foundry operating a battery of cupolas could easily spend \$350,000 on an installation to do this. This is the total amount of money the American Foundrymen's Society needs for its ten-year program.

We again emphasize the need for your companies' contributions to support the Safety and Hygiene and Air Pollution Program because the program is going to help find the answers to our foundry problems and endeavor to create a sane approach to some problems that we do not yet know too much about.

Already quite a few compensation cases are appearing around the country because of effects of noise on the human body. Acoustical treatment of work areas or the adoption of operations that will reduce noise levels is not too far off. Authorities are now studying the noise problem. In some states, committees are already working with regulatory bodies to find some answers to the problem. Here, again, the American Foundrymen's Society Safety and Hygiene and Air Pollution Program can be of inestimable value to all its members if the industry will make it possible financially.

# SINGLE OBJECTIVE SAFETY

By

Dan Farrell*

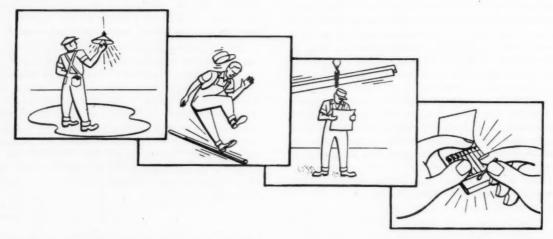
In the early days gold was discovered in the West. Vast amounts were mined, using hand methods and crude implements of the times. These methods, however, were inadequate for further mining. The mines were abandoned. All that remained were waste dumps.

But a later generation with increased know-how and new processes took from the abandoned mines more gold than had come from them originally—proof that progress and discoveries are not limited to new frontiers. And so it has been with safety. Where have the greatest improvements been made? The obvious starting point was the working environment and facilities and we have seen such developments as improved lighting, traffic lanes, guards and many others. The protection of the individual came in for its share of consideration in the form of safety equipment such as goggles, helmets and welding gloves, to name but a few.

1940 to 1948 shows little, if any, additional progress. But the mines have not been abandoned.

Future progress lies within the gap which remains from where we are now to where we want to be. In that area hitherto untapped lies additional pay dirt. We know that where proper design, maintenance, and use of our facilities are observed, relatively few accidents are traceable to these facilities or the lack of personal protective equipment. It seems only reasonable to believe that since individuals continue to be injured, the source of injury must be traceable largely to an act or series of acts or even the failure to act on the part of individuals.

These acts may be as familiar as touching a light fixture while standing on a wet floor, leaving objects as tripping hazards, walking under suspended material, or the simple act of striking a paper match before closing the cover. In searching for clues to



In motivating safety-consciousness we have safety rallies, huddles, personal safety contacts on the job, posters, and slogans like "Safety-first." That all of these approaches produced pay dirt is substantiated by the accident frequency record of the American steel industry. The record, however, from the year these unsafe acts, it is impossible to digest such a mass of information unless it is first put into meaningful form. This can be done by the way of the occupation since most of these acts leading to injury are performed in connection with one's assigned duries

Therefore, the first step in approaching the problem in any department or unit is to get the facts and

^{*} Supervisor of Safety, United States Steel Co., Pittsburgh.

make a simple tabulation of the frequency of acts responsible for injury in each occupation involved. From this tabulation we can then spot the critical situations. Those occupations which show the greatest frequency of accidents and unsafe practices will come into the spotlight as will the act or acts responsible.

# Take Corrective Action

Having determined the critical situations by statistical analysis, it is now possible to single out objectives. With the objective determined, we can now take action. Corrective action begins with a study on the job which may reveal that additional guards on facilities may be required. It may also disclose the lack of personal protective equipment. But more important, as experience has determined, the employee may have developed unsafe work habits which only a study on the job can detect.

The single objective approach does not attack safety problems on a broad, general basis. The single objective approach does establish a series of individual

targets which can be hit one by one.

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So much for the theory itself. Now let us see how it works in practice. To get at the facts, suppose we read a typical description from a Foreman's Accident Report: "The injured man touched hot bar while picking it up to place it on pile and burned his hand." In this description, as in all other accident descriptions, we look for three facts. First, what is the occupation—Piler. Second, what act is involved—touched hot bar. Third, what part of the body was injured—hand.

# AWARENESS CHART

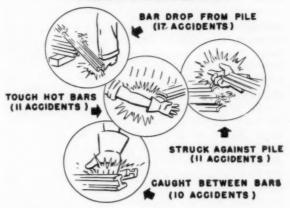
	SPECIFIC	ACTS	INVOLVED		CODE
2 AC PLINS					6 614 9 800' 1 800' 1 100 6 800'
8					TOTALS
H-1					
					-
		FOUCH HOT BAR WHILE FILMS	TOUCH HOT BAR WINE FLINE	SPECIFIC ACTS INVOLVED	Surray James

In order to correlate the many thousands of Foreman's Accident Reports and record the occupation, act involved, and the part of the body injured, it is necessary to use what we call an Awareness Chart. In tabulating the case just mentioned, the piler occupation is shown as one heading and the act involved, "touching hot bar while piling," as the other

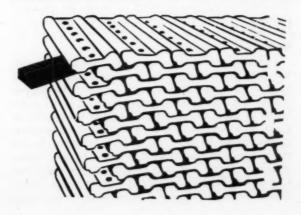
heading. Then, for simplification, the parts of the body injured are designated by seven code letters. Here, the code letter "H" is used for hand. Against it is placed one tally to record the accident.

From an Awareness Chart completed for the Splice Bar Department from a review of various Foreman's Accident Reports of minor and lost time accidents we found that we can spot the critical situations. They center around four occupations of the 19 in the department and concern only eight acts. Let us consider more closely the facts concerning the piler occupation. Of these acts, all concerned with piling, four occur most frequently namely, bars rolling off pile, striking against pile, touching hot bars and catching hand between bars.

# PILER ACCIDENTS



In the first case there were 17 injuries to the employees' feet or legs because the top overhanging bar had been falling off the pile before it was keyed in by the next layer. The supervisors, wishing to eliminate these 17 accidents, went out and studied the job on the job. Their study revealed that the simple act of inserting a small wooden block to support the overhanging bar until it was keyed in accomplished their objective. The other three acts, the difference between 17 and 54 accidents, involved the establishment of but one safe method of performance namely, the placing of a bar on the pile. This method included the wearing of hand leathers and the raising of one end of the bar well above the pile. In completing the method, the middle of the bar is lowered onto





the edge of the pile. The upper hand is released and the bar is then pushed onto the pile with the other hand. Thus, by establishing but two safe methods, four unsafe acts were eliminated.

Through the Awareness Charts, the supervisors of this department became aware that a critical situation existed in this hooking operation. The single objective took the form of the following method. The hands are placed to the outside of the chains, well above the lift to eliminate pinch points and so the chains draped over the end of the lift. The feet are not used to position the chains and are kept in the clear to prevent foot or leg injuries. In lifting with a spreader bar, the Awareness Chart revealed many injuries to the employees' hands and body in performing this operation. A study on the job indicated a hand hook should be used instead of the hands to pull the chain under the lift. This method eliminates arm and body contacts with the load.

Digging out the tap hole is an operation of much concern in the open hearth. In the development of safe performance, personal protective equipment has been given very positive consideration. The employee is furnished asbestos clothing for adequate body protection. While statistical analysis is important, it is definitely not a substitute for on-the-job observation.

In perpetuating the Single Objective Safety program, the use of a Control Chart has proved invaluable. This chart provides a running inventory of the frequency of the acts involved in injury for each occupation. When no tallies appear, the objective is achieved. If injuries are tallied, there is a call for more study out on the job.

The same experience was encountered in the Splice Bar Department where the Awareness Chart showed four occupations out of the 19 accounted for 75 per cent of the accidents, and typifies the experience we found in other departments. The chart prepared by Plant A which shows that 8 of the 64 occupations accounted for 50 per cent of their injuries. In the maintenance unit in Plant B, 5 of the 13 occupations accounted for 77 per cent of the injuries. At Plant C, the Cold Strip Finishing Department had a similar experience in that 6 of the 41 occupations accounted

for 71 per cent of the injuries. As a final example, the experience of the foundry at Plant D was that 6 of the 27 occupations accounted for 63 per cent of the injuries. We believe that this is sufficient proof that our efforts need not be aimless or without guideposts to show us the way.

In the total accident picture, the portion of the picture which has received the most emphasis has been lost time accidents and well it should, for the uselessness of human suffering colors a problem which none of us should lose sight of. Yet for every lost time accident there are scores of minor accidents stemming from the same type of act. Industrial experience shows that these minor accidents constitute as much as 99 per cent of the picture—accidents that could have been lost time accidents.

The total accident picture can be improved when we get the facts—spot critical situations—single out objectives—establish safe methods of performance. No blueprint in itself ever put up a building. No safety plan in itself ever eliminated accidents. All planning can do is point out the way. Single Objective Safety is a blueprint to simplify the task. That is all than can be claimed. But add to this initiative, perseverance and devotion to a cause and the objective can be achieved.

### The Three "C's" of Safety

"Single Objective Safety," deals primarily with the necessity of establishing safe methods of performance. Awareness Charts show the areas on which we should concentrate attention. Above everything else, "Single Objective" stresses an awareness or comprehension of all aspects of the job, with equal emphasis on the human element, the mechanical element, protective clothing and other related factors. To assist our supervisors in the continued use of "Single Objective" and to stimulate further the all-important factor of awareness, we are giving them an additional tool, the Three "C's" of Safety, which are Comprehension, Contact and Control.

The Three "C's" were developed from an objective analysis of 2,000 heavy-industry accident reports. In our analysis, report after report revealed that while injuries were summed up by such generalized statements as, the employee "assumed a hazardous position," "he erred in judgment," "there was lack of interest," or, "he was careless." The descriptions of the accidents at the same time indicated that an employee was injured through contact with some object. For example, he caught some part of his body between one object and other object, or conversely, that his body was in motion and struck against an object not in motion, or some portion of the employee's body was struck by a moving object.

As the study proceeded, these three factors—caught between, struck against and struck by—appeared with startling frequency. Eventually, they began to form a pattern. The study also disclosed that the only objects an employee could have contact with were solids, liquids or gases. The recognition of these limited possibilities of contact simplified the task of classifying accidents.

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# Typical Accident Cores

Suppose we look at a few typical cases. First, caught between. Here a temporary prop was used to support a bin. Where such temporary supports are used, the possibilities of being caught between are always present. The positioning of heavy objects by hand—such as unloading heavy material from trucks—presents many possibilities of injurious contact. Perhaps, in instances, the use of magnet cranes or dump tractors can eliminate these sources of injury. In this instance, the employee is shearing material with a mechanically operated shear knife. While many operations require precise positioning of objects, every possibility, where contact is involved, should be comprehended or visualized and steps taken to correct the situation.

When moving material by cranes the lift itself represents one of the greatest possibilities of injurious contact. The employee's hands may be caught between the lift and the pile. He could also be struck by the chain or the lift. Yet, these situations arise, despite the fact that all of us in heavy industry know about safety handles, hand hooks, and other tools to prevent employees from coming into direct contact with injurious objects.

It is necessary, in many instances, that employees be in the immediate area of moving equipment and material. Suppose a worker is placing a block between stacks of sheets which move along a conveyor. The stacks must be prevented from striking one another. The shortness of the block places his hand in direct contact so that it can be caught between the block and the stationary stack on the right.

# Striking Against Objects

In another case a worker's hand was caught between a load and the block, when the bracing was forced into an upright position, by the weight of steel rods.

Now let us consider the second group—striking against objects. On many occasions, employees of their own volition use improvised methods. Imagine an employee using a ladle stand as a ladder. Also imagine a workman using the back plate of a Bessemer vessel as a platform. Improvising takes many forms. In one case a wooden column was supported on the small base of a jack. In this particular case, the jack collapsed and the workman fell, striking against the planks and the floor.

In another case a walkway was inadequate. The possibility exists for the employee to "strike against" hot water piping or some object equally hazardous. To prevent situations such as this, a permanent walkway and an adequate drainage system could be provided.

In maintenance jobs, such as removing a bearing cap from a vertical turning lathe in a wheel plant, injuries occurred due to improvised methods.

Situations were discussed where employees were caught between and struck against objects. The third group will be concerned with cases of struck by objects.

On a shear machine it is possible for the man,

known as a "catcher," to be struck by a sheet, should one pass through the rolls while he is straightening the stacks. Certainly, a better method could be devised to eliminate the need for him to place his hands in the area of contact. A repairman, in trying to pass the exhaust valve crosshead through the bonnet of an engine, used the wrong tools to cut off the bushing. Result was a fractured hand.

From these examples, it must be obvious to you that one has to have contact with some object before injury can occur. This is true whether it is in the home or in industry. In controlling contacts, the first step is to anticipate or be able to comprehend the possibilities of injury—the ability to "see" while

we are looking.

The most phenomenal strides in this direction have occurred in the home appliance field. Typical of these is the abandonment of the old-fashioned wringer for the spin dryer; the departure from iron handles and pan holders to nonconductors of heat, such as plastic and glass handles.

Likewise in the automotive field, jacks are now made with extensions, improved vision is provided along with hundreds of other developments designed to eliminate the possibilities of injurious contact.

In the steel industry, we have followed the same line of thinking. A Betts boring machine operator, in endeavoring to perform two operations at the same time, reached for the mechanism to adjust the machine while trying to concentrate on the boring operation. His finger was caught between the teeth of the meshing gears. After engineering principles were applied, an improved method for indexing the feeding mechanism and a shield were installed, thereby eliminating the possibility of contact.

We believe there is still considerable distance to go in eliminating injurious contact. But to do this it is necessary to study every job to comprehend all the possibilities of injurious contact. An analysis sheet enables us to spot those parts of the job where employees might experience injurious contacts and be caught between, struck against or struck by objects.

Suppose we analyze a job in a rolling mill. The occupation is "spot scarfer." He is "turning slabs on edge for strapping into bundles." The steel slabs are placed flat upon skids to facilitate cutting them into proper lengths. It is necessary, next, to set them up on edge and bundle them with wire. First, are determined the objects capable of inflicting injury during this operation. They are the slabs, the skids and the tongs. Now, let us look at the possibilities of injurious contact with these objects. An employee can be caught between the "slabs and skids," either while "turning slabs or after turning" them. He can be caught between the "arms of the tongs," also, while "turning slabs." Another possibility is that he can strike against the skids during the walking he must do on the job; or in losing balance through undue exertion, he can strike against the tongs. Finally, he can be struck by the slabs, also, during or after the turning operations, and because of the manner in which he must stand to use the tongs, he faces the possibility of being struck by them.

Knowing these possibilities, the next step is to control the situation by developing a safe procedure. These hazards can be minimized—in fact, virtually eliminated—by the following method: "provide holes in skids so that steel pins can be placed in holes to prevent slabs from falling after they have been turned. Tongs should be eliminated and a turning wrench employed to eliminate the possibility of being caught between the arms of the tongs and to provide proper leverage for turning."

Having analyzed the job and found a safe method of performance, let us check ourselves—"Does the recommended procedure eliminate contact?" With the use of steel pins to hold the slabs in place and a turning wrench which provides proper leverage, we can say yes with respect to the possibilities of being caught or being struck by objects. However, as long as the employee must remain in close proximity to the slabs, skids and tongs or turning wrench, we cannot eliminate the possibility of his striking against these objects.

Let us consider another job-positioning a crane magnet over wheel blanks at a water pit where they have been cooled. Wheel blanks, as you know, are the sections of steel from which railroad car wheels are forged. Under objects capable of inflicting injury are listed the magnet, the wheel blanks, hot water, hand hook, the column behind the man and the rail on the ground. The possibility of the employee being caught between the wheel blank and the ground exists any time the magnet is handling the blanks and the employee is in the area. He can strike against the hot water, the magnet or the column while positioning the magnet, and he can strike against the rail either in walking or, again, while positioning the magnet over the wheel blank. He can be struck by the blank, magnet or splashing of the hot water

during the magnet's handling of the blank. The hand hook can strike him, while he is positioning the magnet.

To control the possibilities of contact, the pit should be reconstructed so that it is no more than 6 in. wider than the magnet. A trough with wooden sides should be built the entire length of the crane's path on each end of the pit so that there is no chance of the blanks rolling beyond the reach of the magnet. This arrangement, or a conveyor, would eliminate the necessity for the employee to be in the area where the above contacts are possible. If this procedure is followed, all of the listed possibilities for injurious contact would be eliminated and controlled as far as this employee is concerned in his job situation. As one can see from these analyses, the problem facing industry can be solved if our plans avoid the maze of abstract thinking that has tended to inhibit our actions in the past. We believe our next major gain toward the ultimate solution of the problem lies in the principles of the Three "C's" of Safety-Comprehension, Contact and Control.

#### DISCUSSION

Chairman: J. M. Kane, American Air Filter Co., Louisville. Kv.

Recorder: H. F. Scobie, American Foundrymen's Society, Chi-

F. B. SKEATES: 1 Does the Safety Division of your Company have final authority in correcting unsafe conditions it discovers?

MR. FARRELL: In our Company safety is a staff responsibility, so when an unsafe condition is discovered, it is promptly brought to the attention of the operating department involved. Under normal conditions, the operating department immediately corrects the unsafe condition referred to it. Should it be necessary, there are operating channels to use as high as the Vice President-Operations should that step be required to correct any unsafe condition found.

¹ Plant Supt., Link Belt Co., Chicago.

# EFFECTS OF GATING PRACTICE ON LEAK TIGHTNESS OF 85-5-5 AND 81-3-7-9 ALLOY CASTINGS

J. G. Kura* and L. W. Eastwood**

#### Introduction

Foundrymen will universally agree that gating practice is an important aspect of foundry operation. There is, however, no universal agreement regarding the best method which should be employed to gate the castings. Recent studies 1, 4 sponsored by the American Foundrymen's Society have shown that turbulence and air aspiration must be carefully avoided if sound castings of reactive metals are to

be produced.

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On the basis of this work, it appeared that it would be more desirable to gate experimental leak-test castings at the bottom in order to avoid the turbulence which would occur if the melt was introduced at the top of the mold cavity. This, however, differed from some commercial practices. Accordingly, an investigation was undertaken to compare the effects of gating practice on the leak tightness of 85-5-5-5 and 81-3-7-9 alloys. Incidentally, the effects of pouring temperature and melting practice were also determined, and a comparison was made of the relative susceptibilities of the two alloys, 85-5-5-5 and 81-3-7-9, to leakage. This paper describes the results obtained in this investigation.

#### **Experimental Procedure**

Duplicate 300-lb heats of 85-5-5-5 and 81-3-7-9 alloys were prepared by a good melting practice in a highfrequency induction furnace equipped with a silica crucible, and in a gas-fired tilting furnace equipped with a clay-graphite crucible. A total of eight heats. four of each alloy, were prepared with charges consisting of ingot metal. The maximum temperature of the metal in the furnace was approximately 150 F higher than the highest pouring temperature employed. Strongly oxidizing conditions were maintained throughout all melting operations.

When the desired maximum temperature was reached, the melt was skimmed and transferred to

a thoroughly preheated pouring ladle. Ladle additions of 11/4 lb of zinc to every 100 lb of 85-5-5-5 alloy and 13/4 lb of zinc to every 100 lb of 81-3-7-9 alloy were made to compensate for the zinc lost during melting. Approximately 2 oz of 15 per cent phosphorcopper per 100 lb of melt were added to both alloys after the zinc addition had been made to the ladle during the transfer.

When the melts in the pouring ladle had cooled to the proper pouring temperature, as indicated on the accompanying charts, they were poured into green sand molds. The sand had an A.F.S. permeability of about 51 and a green compressive strength of 15 psi. The molds were rammed to a hardness of approximately 65. One test-bar casting, similar to the ASTM Tentative Standards B208-49T, Fig. 3, was poured from each heat in order to evaluate the quality of the melt obtained. Leak-test castings were gated in three different ways.

# Pouring Temperatures and Gating Designs

Each type was poured at three different temperatures. The three different gating designs are illustrated by Figs. 1, 2, and 3. Further details of the sprue-base design employed for the two top-gated patterns are illustrated by Fig. 4. Although the gating designs illustrated are not necessarily representative of current commercial practice, they represent what might be expected with castings (1) gated at the bottom, (2) gated at the top without a riser, and (3) gated at the top with a riser.

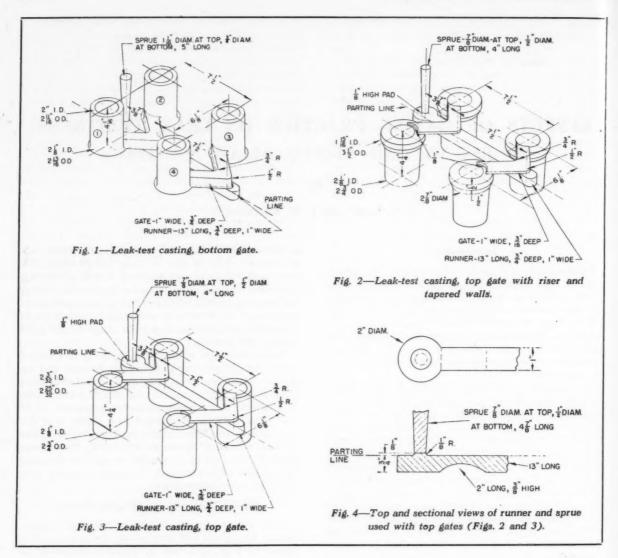
After the castings had cooled, they were shaken out of the mold and sandblasted. The test bars were then machined to standard 0.505-in. diameter threaded bars with a 2-in. gage section. Their densities were then measured before obtaining the tensile properties. The leak-test castings were faced off at each end,

their densities measured, and leak tested.

After the leak tests had been completed, the interior and exterior of the cylinders were machined to the dimensions shown in Fig. 5. Densities were again measured, after which the cylinders were again subjected to leak tests. All of the leak testing was done by using a nitrogen atmosphere up to a pressure of 1300 psi. The pressure at which leakage first occurred

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^{**} Formerly Supervisor at Battelle Memorial Institute, now Assistant Director of Research, Kaiser Aluminum & Chemical Corporation, Spokane, Washington.



was determined. If the cylinders leaked very badly, the testing was discontinued at a pressure which produced a large volume of leakage.

All of the melts were analyzed and found to be well within the ASTM specification limits. Typical analyses were as follows:

	Composition, Per Cent							
Alloy	Cu	Sn	Pb	Zn	Fe	Sb	Ni	P
85-5-5-5	84.75	4.89	4.98	4.50	0.06	0.19	0.61	0.013
81-3-7-9	80.82	3.09	6.77	8.55	0.17	0.18	0.41	0.014

#### **Experimental Results**

The tensile properties and densities obtained on the test bars are listed in Table 1. The generally high tensile properties and high densities obtained on the test bars indicate that most of the melts were of excellent quality. The last heat of 85-5-5-5 alloy had intermediate tensile properties, indicating a melt of only fair quality.

The leak-test data were analyzed in various ways to show the effects of gating practice, melting conditions, alloy composition, and pouring temperature on the susceptibility of the castings to leakage before and after machining. It should be emphasized that the castings, machined inside and out, are much more likely to leak than castings with the as-cast skin in place. The outer skin of the casting tends to be of a greater degree of soundness than the interior of the casting. This fact should be borne in mind when noting the relatively high percentage of leakers obtained after the castings had been machined.

Effects of Melting Practice: As noted previously, good melting practice was employed to produce melts in both the high-frequency induction furnace and the gas-fired tilting furnace. The induction furnace was equipped with a silica crucible, the gas-fired furnace with a clay-graphite crucible. Figures 6 and 7 show the effects of furnace type on the occurrence of leakers in leak-test cylinders of both alloys before and after machining. It will be noted that the percentage of leakers is much higher after the cylinders were machined. In this instance, the gas-fired furnace produced melts which most closely approached the op-

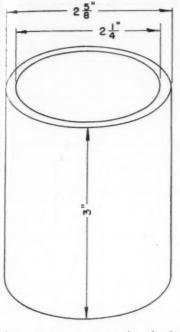


Fig. 5—Machined leak-test casting.

timum gas content to give the lowest percentage of leakers.

It will be shown later that the melts produced in the induction furnace have a higher density, indicating that melts made in this furnace had lower gas content than those made in the gas-fired furnace. In this instance, however, the gas content of the melts prepared in the induction furnace is lower than the optimum value for maximum leak tightness of the castings poured.

It does not follow, however, that the lowest percentage of leakers will always be obtained by melts in a gas-fired furnace. Experimental work described in another paper⁵ showed that the optimum gas content of the melt was more nearly approached in the high-frequency induction furnace than in the gas-fired furnace, a condition which was the reverse of that shown by Figs. 6 and 7.

#### Reduce Gas Contents

It is well to emphasize that the optimum gas content which will produce the lowest percentage of leakers is somewhat critical. Better control can usually be obtained in the foundry if melts of very low gas content can be consistently produced. If difficulties with leakage are encountered because of excessively low gas content of the melt, the pouring temperature can be increased slightly to increase the gas absorption from the mold to offset the low gas content of the melt. On the other hand, if the gas content of the melt is too high to produce the maximum degree of leak tightness, there is not much the foundryman can do except to melt the alloy under better conditions.

The effects of furnace design on the relative gas contents of the melts are indicated by Fig. 8, which shows the effect of pouring temperature on the density of the castings poured from melts made in the two furnaces. It will be noted that the induction furnace

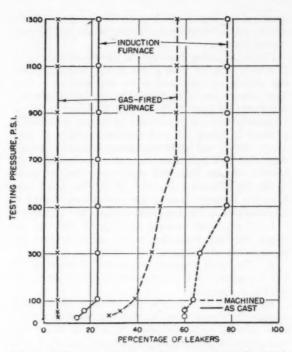


Fig. 6—Effect of melting method on leak tightness of cylinders of 85-5-5-5 alloy at various testing pressures. Each curve represents the average value for 36-72 cylinders gated by 3 different methods and poured over the temperature range 2000-2135 F from duplicate melts prepared by a good practice.

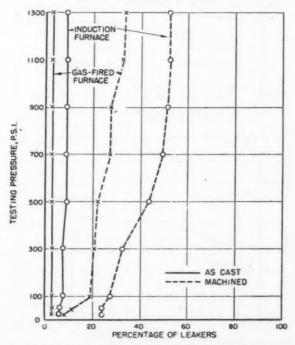


Fig. 7—Effect of melting method on the leak tightness of cylinders of 81-3-7-9 alloy at various testing pressures. Each curve represents the average value for 64-68 cylinders gated by 3 different methods and poured over the temperature range 2005-2140 F from duplicate melts prepared by a good practice.

produced castings of the highest density, and that the density was reduced by machining off the outer sound casting surface.

Effects of Gating Practice: As illustrated by Figs. 1, 2, and 3, castings were gated at the bottom, at the top with a top riser, and at the top without a riser. The relative merits of these three gating systems are illustrated by Figs. 9, 10, 11, and 12, wherein the relationship between percentage of leakers and testing pressure is shown. In each instance, castings of both alloys, before and after machining, have the greatest degree of leak tightness when they are gated at the bottom, and are least leak tight when they are gated at the top without a top riser. It will be shown later that the primary reasons for the greater degree of unsoundness in the top-gated castings are as follows:

1. The greater amount of trapped gasholes and dross resulting from the metal cascading into the mold cavity of the top-gated castings.

2. The more frequent occurrence of shrinkage in the top-gated castings, particularly in the ones not having a top riser.

Figures 13 and 14 show the effects of gating practice on the densities of the leak-test cylinders poured over a range in temperature. It will be noted that the cylinders poured with a bottom gate have high densities, while those poured with the top gate and no riser have low densities. Those poured at the top with a riser have densities in the unmachined condi-

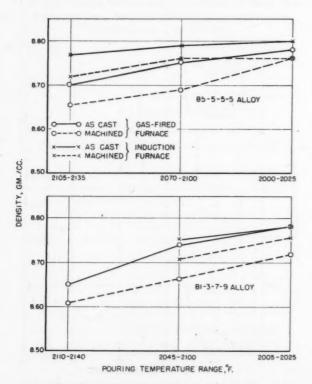


Fig. 8—Effect of melting turnace and pouring temperature on the density of leak-test cylinders of 85-5-5-5 and 81-3-7-9 alloys. Each point represents the average value for 10 to 20 cylinders from duplicate melts prepared by a good practice and poured into molds gated by three different methods.

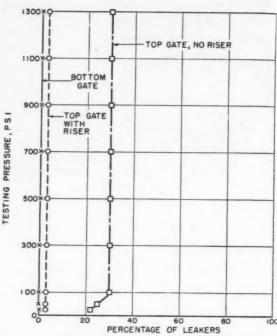


Fig. 9—Effect of gating practice on the leak tightness of as-cast cylinders of 85-5-5-5 alloy at various testing pressures. Each curve represents the average value for 36 cylinders poured over the temperature range 2000-2135 F from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

Table 1—Tensile Properties* of Melts Prepared for the Study of the Effect of Gating Practice on the Pressure Tightness of 85-5-5-5 and 81-3-7-9 Alloys

		Pouring	Tensile	Elong.	1	Density
Heat	Melting	Temp.,	Strength	in	Fracture	gr/
No.	Furnace	F	psi	2 in., %	Rating**	cu cm
		85-5	-5-5 All	ov		
6911A	Induc-	2150	36,600	34.0	Slightly	8.75
AL	tion		36,600	46.5	coppery	-
Average			36,600	40.2		
6912A	Induc-	2130	37,900	43	Slightly	-
AL	tion		37,900	37	coppery	8.83
Average			37,900	40		
6916A	Gas	2150	37,200	39	Slightly	8.74
AL	Fired		37,600	40	coppery	_
Average			37,400	39.5		
6982A	Gas	2150	35,000	28.5	Slightly	8.71
AL	Fired		35,400	32	coppery	_
Average			35,200	30.2		
		81-3	-7-9 Alle	DY		
6915J	Induc-	2000	33,500	25.5	Slightly	8.87
JL	tion		33,100	25.0	coppery	-
Average			33,300	25.2		
6983J	Induc-	2000	31,500	23	Gray	8.86
JL	tion		33,500	27	Slightly	
Average			32,500	25	coppery	
6917J	Gas	2000	33,300	29	Slightly	8.88
JL	Fired		34,600	33	coppery	_
Average			34,000	31		
6984J	Gas	2000	31,700	23	Slightly	8.83
JL	Fired		33,200	26	coppery	-
Average			32,500	24.5		

*Double horizontal full-web-type bar gated according to Tentative ASTM Standards, B208-49T, Fig. 3.

**Gray indicates sound metal. Slightly coppery indicates presence of a small quantity of microporosity.

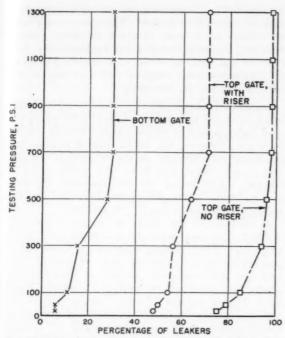


Fig. 10—Effect of gating practice on the leak tightness of machined cylinders of 85-5-5-5 alloy at various testing pressures. Each curve represents the average value for 46 to 48 cylinders poured over the temperature range 2000-2135 F from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

tion comparable to the bottom-gated cylinders but, in the machined state, they are lower and intermediate between the bottom-gated cylinders and cylinders gated at the top without a riser.

Effects of Pouring Temperature: As indicated previously, the leak-test cylinders were poured within three fairly narrow ranges of pouring temperature. The effects of the pouring temperature on the occurrance of leakers in both alloys before and after machining the cylinders are illustrated by Figs. 15 and 16.

It has been shown in a previous paper⁵ that high pouring temperatures of the order of 2250 F, in 85-5-5-5 alloy, greatly increase the percentage of leakers obtained, because at these high pouring temperatures large volumes of gas are absorbed from the mold and the gas is reprecipitated again during solidification of the casting. This causes the occurrence of a large volume of microporosity, which is one of the principal causes of leakers. Specifically, the lowest percentage of leakers was obtained at the following pouring temperatures:

Alloy	Condition	Best Pouring Temperature, F
85-5-5-5	As-cast	2070 to 2100 (medium)
85-5-5-5	Machined	2070 to 2100 (medium)
81-3-7-9	As-cast	2110 to 2140 (high)
81-3-7-9	Machined	2045 to 2140 (high and medium)

The significance of these specific temperature ranges may not be too real. It is quite likely, however, and there is considerable evidence to support the conclusion, that the very low-pouring-temperature ranges

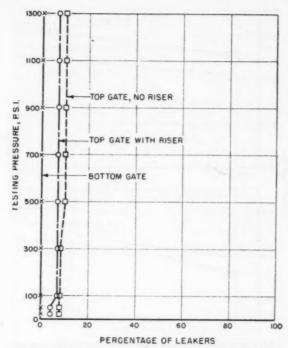


Fig. 11—Effect of casting practice on the leak tightness of as-cast cylinders of 81-3-7-9 alloy at various testing pressures. Each curve represents the average value for 40 to 48 cylinders poured over the temperature range 2005-2140 F from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

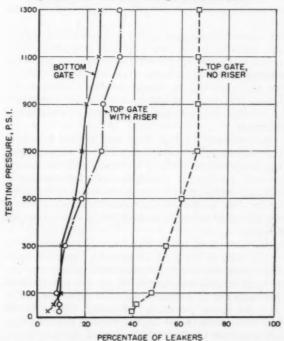


Fig. 12—Effect of gating practice on the leak tightness of machined cylinders of 81-3-7-9 alloy at various testing pressures. Each curve represents the average value for 40 to 48 cylinders poured over the temperature range 2005-2140 F from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

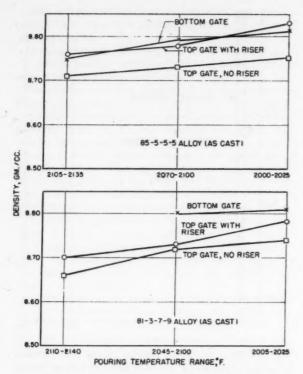


Fig. 13—Effect of gating method and pouring temperature on the density of as-cast leak-test cylinders of 85-5-5-5 and 81-3-7-9 alloys. Each point represents the average value for 2 to 12 cylinders from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

are undesirable because of the accentuation of shrinkage. In other words, better feeding is required when very low pouring temperatures are used. If improved feeding could be utilized, then the lowest possible pouring temperature would probably produce the lowest percentage of leakers because the occurrence of microporosity decreases as the pouring temperature decreases.

As illustrated by Figs. 8, 13, and 14, the density of the leak-test cylinders increases with decreasing pouring temperature. This is a direct result of the decreased amount of microporosity with decreasing pouring temperature. The occurrence of shrinkage at the low pouring temperatures does not have a marked effect upon the density, but it does accentuate leakage.

In these tests, then, high densities produced by using melts of low gas content or by using low pouring temperatures do not necessarily reduce the percentage of leakers obtained. If the castings could be perfectly fed to compensate for solidification shrinkage, the lowest percentage of leakers should be obtained by melts of the highest quality, i.e., lowest gas content, and by the employment of the lowest possible pouring temperatures. If, however, they are not perfectly fed, as is usually the case, it seems desirable to use slightly higher pouring temperatures or prepare melts having slightly higher gas content to avoid or reduce the occurrence of shrinkage.

Both of these factors will increase the amount of

microporosity. If overdone, an excessive amount of microporisity will be formed. Therefore, a very careful balance is necessary to avoid shrinkage without obtaining an excessive amount of microporosity. The superior leak tightness of the 81-3-7-9 alloy poured in the high-temperature range, as shown by Fig. 16, appears to bear out this reasoning.

Causes of Leakage: All of the machined cylinders which leaked at nitrogen pressures of 1300 psi or less were radiographed to determine the cause of leakers. In a few instances, the defect causing leakage was too small to be revealed on the radiograph. In general, it was found that there were three causes of leakage: (1) shrinkage; (2) microporosity; (3) dross and gasholes. These defects are illustrated by the radiographs shown in Figs. 17, 18, and 19.

A summary of the causes of leakers is given in Table 2. The summary is arranged to show the effects of gating practice, pouring temperature, furnace type, and alloy composition on the causes of leakers. This summary also shows the number of cylinders cast under each condition, the number of castings which were rejected because of other casting defects not pertinent to the investigation, the total number of cylinders leak tested, and the number of castings radiographed, which is the same as the number of castings which leaked. In a number of instances, leakage was caused by more than one type of defect. Therefore, the total number of leakers listed by the type of defect may appear to exceed the actual number of leaky cylinders radiographed.

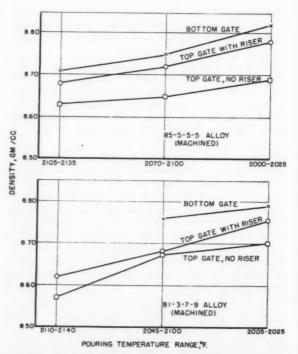


Fig. 14—Effect of gating method and pouring temperature on the density of machined leak-test cylinders of 85-5-5 and 81-3-7-9 alloys. Each point represents the average value for 4 to 12 cylinders from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

It is quite evident from the data in Table 2 that the superiority of the bottom-gated castings is achieved largely by the almost complete absence of dross and gasholes and the much less frequent occurrence of shrinkage. Shrinkage, dross, and gasholes are most predominant in the top-gated castings not provided with a top riser. The dross and gasholes result from the cascading liquid falling into the mold cavity from the top. If a riser is employed, some of the trapped gas bubbles probably have an opportunity to escape.

The effects of pouring temperature on the cause of defects are evident in Table 2. Pouring temperature has little or no effect on the occurrence of dross and gasholes. Shrinkage, however, is much more predominant at the low pouring temperature, as would be expected. On the other hand, the occurrence of microporosity is more common at the high pouring

temperatures.

Table 2 also shows the effects of melting method or furnace type on the probable cause of leakers. The data show that gasholes occur to the same degree in castings poured from melts made in the two types of furnaces. Likewise, the occurrence of microporosity is similar for castings poured from melts of either type of furnace. However, probably because of the lower gas content of the melts made in the high-frequency induction furnace and the consequent, more frequent occurrence of shrinkage, castings poured from these lower-gas-content melts are more likely to leak.

The 85-5-5-5 alloy, in this work at least, tends to

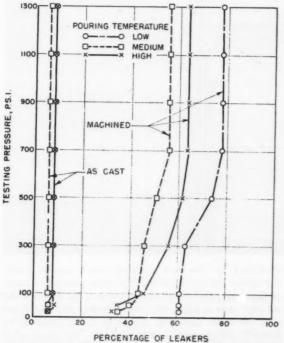


Fig. 15—Effect of pouring temperature on the leak tightness of cylinders of 85-5-5-5 alloy at various testing pressures. Each curve represents the average value for 34 to 51 cylinders gated by 3 different methods and poured in the temperature ranges of 2105-2135 F, 2070-2100 F, and 2000-2025 F from duplicate melts prepared by a good practice in induction and gas-fired furnaces.

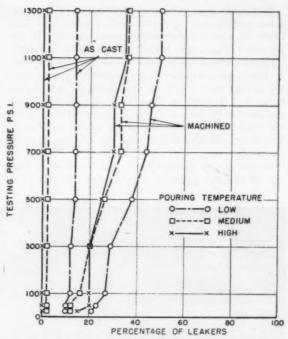


Fig. 16—Effect of pouring temperature on the leak tightness of cylinders of 81-3-7-9 alloy at various testing pressures. Each curve represents the average value for 20 to 64 cylinders gated by 3 different methods and poured in the temperature ranges of 2110-2140 F, 2040-2100 F, and 2005-2025 F from duplicate melts prepared by a good good practice in induction and gas-fired furnaces.

produce more leakers because of the more frequent occurrence of shrinkage and microporosity.

#### Summary

The effects of gating practice, melting furnace, and pouring temperature on the leak tightness of as-cast and machined leak-test cylinders of 85-5-5-5 and 81-3-7-9 alloys were determined. The castings were poured in natural green sand. Duplicate 300-lb melts of each alloy were made by a good melting practice in the high-frequency induction furnace equipped with a silica crucible and in the gas-fired furnace equipped with a clay-graphite crucible.

Three gating techniques were investigated. By one method, the gates entered the bottom of the casting. By the other two methods, the gates entered the top of the casting, one of which was equipped with a small top riser. Four cylindrical leak-test castings

were made in each mold.

Radiographic examination of all the leaky castings showed that the defects causing leaks were identified as follows: (1) shrinkage; (2) microporosity: (3) dross and gasholes.

Dross and gasholes caused leaks in the two topgated castings because of the turbulence during the pour. Probably because some of the trapped gas bubbles escaped, the cylinder equipped with a small riser on top was less subject to the occurrence of dross and gasholes. Melting methods, pouring temperature, or alloy composition had relatively little effect on the frequency of occurrence of this defect. Shrinkage was most likely to occur in the topgated casting not equipped with a riser, and least likely to occur in the bottom-gated casting. In the present work, it was found that melts in the highfrequency induction furnace contained less than the optimum gas content, and, consequently, greater leakage occurs as a result of the shrinkage which was formed. Likewise, in these tests, the 85-5-5-5 alloy appeared to be more prone to the formation of shrinkage than 81-8-7-9 alloy cast under similar conditions. Shrinkage was also accentuated by very low pouring temperatures.

Microporosity was less likely to be a cause of leakage in castings gated at the bottom. As would be expected, microporosity also occurred more frequently at high pouring temperatures. Because all of the melts were prepared by a good practice, the furnace type apparently had no appreciable effect on the occurrence of microporosity in the castings made from melts prepared in them. In this work, however, microporosity caused leakage more frequently in 85-5-5-5 alloy than in 81-3-7-9 alloy castings.

Because shrinkage is more likely to occur with decreasing pouring temperatures, and microporosity is accentuated by increasing pouring temperatures, the lowest percentage of leakers was generally obtained by an intermediate pouring temperature of approximately 2075 F for both alloys.



Fig. 17 (left)—Top-gated cylinder without a riser, poured from 85-5-5-5 alloy at 2020 F. This radiograph shows the typical appearance of heavy shrinkage (dark areas) in regions also occupied by light microporosity (dark mottling). Leaks occurred in the areas marked with a circle. The top of the casting is at the left of the radiograph and the four quadrants are marked off with vertical lines. The gate was attached to the quadrant at the bottom in this illustration and in the ones shown in Figs. 18 and 19.

Fig. 18(center)—Top-gated cylinder with a riser, poured from 85-5-5-5 alloy at 2100 F. This radiograph shows leakage caused by microporosity. The microporosity is illustrated by the dark mottling which occurs almost throughout the entire casting. The microporosity is small, irregular voids, more or less connected. In many of the castings radiographed, the microporosity did not occur over the entire casting as it does in this illustration. In general, microporosity is accentuated in the quadrant of the casting nearest the gate.

Fig. 19 (right)—Top-gated cylinder without a riser, poured from 85-5-5-5 alloy at 2075 F. This radiograph shows heavy shrinkage in the upper portion of the casting adjacent to the gate, light microporosity mainly below the shrinkage area, several large gasholes, and presumably some entrapped dross.

TABLE 2-PROBABLE CAUSES OF LEAKERS IN MACHINED CYLINDERS

		No. Cylinder Rejected	Total No	-	nd % tings			No.	and % o	f Leakers			
	Total No. Cylinders	Because of Casting	Cylinder	s Ra	dio- ohed		cro- osity	Shri	nkage	Dross Gasl	and noles	Other	
	Cast	Defects	Tested	No.	%	No.	%	No.	%	No.	%	No.	%
		Effec	t of Gatin	g Pract	ice at All	Pouring T	emperatu	res-Both	Alloys				
Gating													
Bottom	96	10	86	24	28	13	15	4	5	1	1	6	7
Top, with riser	96	4	92	49	53	18	20	19	21	12	13	2	2
Top, no riser	96	0	96	79	82	21	22	47	49	23	24	1	1
		Effect	of Pouring	Temp	erature W	ith All G	ting Met	hods-Bot	h Alloys				
Pouring Temp.,	F												
2105-2140 (high)	80	12	68	38	56	25	37	8	12	10	15	0	0
2045-2100 (med.)	108	1	107	47	44	20	19	14	13	14	13	3	3
2000-2025 (low)	100	1	99	67	68	7	7	48	48	12	12	6	6
	- 1	Effect of Fur	nace Type	With	All Pourin	g Temps.	and Gatin	ng Method	-Both	Alloys			
Furnace								-					
Induction	144	10	134	89	66	25	19	46	34	19	14	9	7
Gas Fired	144	4	140	63	45	27	19	22	16	19	14	0	0
		Ef	fect of Al	lov Wi	th All Pou	ring Tem	s. and G	ating Meth	ods				
Alloy .		-		,									
85-5-5-5	144	2	142	95	67	39	27	45	32	18	13	4	3
81-3-7-9	144	-	132	57	43	13	10	23	17	20	15	5	4

It is evident from the preceding statements that, for the specific casting investigated, the lowest percentage of leakers was obtained by employing the bottom-gating technique. The indications are also that fewer leakers were obtained by the use of intermediate pouring temperatures of approximately 2075 F, melts of optimum gas content, and 81-3-7-9 instead of 85-5-5-5 alloy. It should be emphasized, however, that leakage depends upon several factors. Therefore, it is difficult to make a reliable comparison of the susceptibilities of various alloys to leakage.

#### Acknowledgments

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#### **Bibliography**

1. R. E. Swift, J. H. Jackson, and L. W. Eastwood, "A Study of the Principles of Gating," American Foundrymen's Society, Transactions, vol. 57, pp. 76-88 (1949).

2. "A Study of the Principles of Gating," a black and white motion picture issued in 1949, and obtainable through the American Foundrymen's Society, 616 S. Michigan Ave., Chicago.

3. K. Grube and L. W. Eastwood, "A Study of the Principles of Gating," American Foundrymen's Society, Transactions, vol. 58, pp. 76-107 (1950).

4. "Gating for Improved Casting Quality," a motion picture

4. "Gating for Improved Casting Quality," a motion picture in full color issued in 1950, and obtainable from the American Foundrymen's Society, 616 S. Michigan Ave., Chicago.

5. J. G. Kura and L. W. Eastwood, "The Effects of Mold

Materials on Leak Tightness and Mechanical Properties of 85-5-5-5 and 81-3-7-9 Alloy Castings," A.F.S. Transactions, vol. 60, pp. 247-261 (1952).

#### DISCUSSION

Chairman: G. P. Halliwell, H. Kramer & Co., Chicago. Co-Chairman: H. J. Roast, London, Ont., Canada. Recorder: R. A. Colton, American Smelting & Refining Co., Barber, N. J.

R. A. COLTON (Written Discussion): There have been a number of comments about the fact that the authors used a bottom gating technique in these experiments to obtain their best results. It is certainly true that, generally speaking, bottom gating is not considered most advantageous for the manufacture of sound, porosity-free castings. I am certain that the authors were aware of this fact and their use of bottom gating in this particular instance probably can be considered as part of a group of experiments.

It is apparent that even though a bottom gating technique gave the most satisfactory results in the pressure test with the castings used in this work, it is a function of the particular casting and gating system used. That is to say, with the dimensions both of the casting and gating system involved, particularly with the thin walls of the casting, a bottom gating technique gave good results whereas if a heavier casting had been used less sound metal would have been available in the casting.

The evil of bottom gating is that it upsets the desired directional solidification pattern and establishes the temperature gradient in the wrong direction from that needed to properly feed castings. I believe the explanation of why the bottom gating technique worked in these experiments is that the wall of the casting was so thin that once the cavity was filled solidification took place more or less uniformly throughout the casting so that the defects normally found from a reversed temperature gradient were not evident here. It would be a serious mistake to use the results of these experiments as an endorsement of bottom gating since generally speaking a top gating, directional solidification pattern will give good results far more often than will bottom gating with most castings.

W. B. Scott: A comparison of top gating vs bottom gating could be without significance if no consideration was given to the velocity of the metal flowing into the mold cavity. High velocities as distinguished from high rates of volume flow could discount the value of either type of gating. This factor could account for the wide variations in the results reported for each type of gating.

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# RECLAIMING USED MOLDING SAND BY AIR SCRUBBING

By

H. H. Fairfield, James McConachie, and H. F. Graham*

#### ABSTRACT

A device for reclaiming used steel foundry sand is described. Dry sand is blown through a turbulent air stream. The effect on sand properties is recorded. Sand so treated is used in molding and core sand mixtures.

It is well known that discarded excess used molding sand can be processed into a product identical in properties to new washed and dried core sand. Jeter^{1,2} described sand reclamation methods. Cummings and Armstrong³ applied standard mining equipment to foundry sand reclamation. Bennett⁴ outlined the features of a modern sand reclaiming apparatus.

Since the equipment used to reclaim steel molding sand at the foundry with which the authors are associated differs from conventional designs, a description of the methods used might be of general interest

#### Difference Between New and Used Sand

In the tests on core sands described in this paper the following mixture was used:

																Grams
Sand		 				,										1000
Corn Starc	h .	 					 							 		10
Linseed Oi																
Water																

The base sand used in the foundry was a washed and dried silica sand averaging A.F.S. Grain Fineness No. 57. Samples of used sand were taken from the shakeout and from the cleaning room floor. Test cores 2 in. in diameter and 2 in. high were made from each type of sand. The baked compression strength of the cores was measured.

New sand cores ranged from 800 to 1000 psi. Shakeout sand cores had 120 psi compression strength. Cleaning room sand (also called burnt sand) cores had only 50 psi compression strength after baking.

It was apparent that the used sand contained impurities which made it unsuitable for use as a core sand. Since the impurities were mostly in the form of fines, it was concluded that the sand might be cleaned by sandblasting and blowing away the fines.

#### Trial Model Sand Cleaner

Figure 1 shows the first device used to reclaim the steel foundry sand. Its essential features are:

- (1) A steel pipe, 6 ft high, 4 in. 1.D.
- (2) An air syphon which sucks up dry sand and fires it into the pipe.
- (3) Air and sand stream enters the pipe tangentially at an angle of 20° from horizontal.
- (4) Syphon connected to 60-80 psi air pressure line.
- (5) Dust and fines float out the top of the pipe.(6) Cleaned sand drops out the bottom of the

#### Properties of Reclaimed Core and Molding Sands

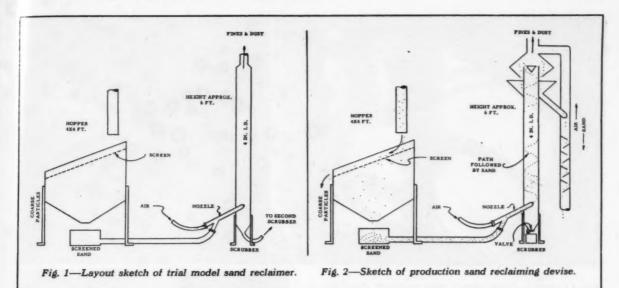
Cores were made from sand treated in the device shown in Fig. 1. Baked compression tests results on

TABLE 1—BAKED COMPRESSION STRENGTH OF CORES MADE FROM SAND RECLAIMED IN DEVICE SHOWN IN FIG. 1

Type of Sand	Number of Passes Through Reclaimer	Baked Compression Strength, psi (2 x 2-in. specimen)				
new washed and		,				
dried core sand	none	900				
shakeout	1	234				
shakeout	2	755				
shakeout	4	958				
cleaning room	1	188				
cleaning room	2	508				
cleaning room	3	684				
cleaning room	6 4	652				
	xes was tested using an the following results:	additional 1 per cent				
cleaning room	1	604				
cleaning room	2	859				
cleaning room	3	954				
cleaning room	4	954				

Note: Except as noted, above mixtures contained 1000 grams of sand, 10 grams of cornstarch, 30 grams of linseed oil, and 40 grams of water.

Chief Metallurgist, Sand Supervisor, and Foundry Engineer, respectively, William Kennedy & Sons Ltd., Owen Sound, Ontario, Canada.



these cores are shown in Table 1. Shakeout sand, after two passes through the reclaimer, apparently was a satisfactory core sand. Burnt sand from the cleaning room would not make a strong core unless an extra 1 per cent of oil was added to the mixture.

Batches of molding sand were prepared using reclaimed sand with 4 per cent fireclay and 4 per cent bentonite additions. The properties of these sands were measured by A. E. Murton at the Canadian Bureau of Mines, Ottawa, who reported the green, dry, and high temperature properties (Table 2).

#### Production Model Sand Cleaner

Since the results on the trial model were encouraging, it was decided that a production model should be built into the sand system. Figures 2, 3, 4, and 5 show the equipment now in use.

With the trial model it was found that changes in air pressure would result in all of the sand being

TABLE 2-PROPERTIES OF NEW AND USED MOLDING SANDS, AND SAND FROM RECLAMATION DEVICE (Fig. 1).

	New	Used		Reclaime Trea		
Properties	Sand	Sand	Once	Twice 3	Times 4	Times
Moisture, %	8.7	9.0	9.2	9.1	9.1	8.9
Green permeability	147	94	94	107	118	126
Dry permeability	184	152	126	167	190	195
Green compressive strength, psi Dry compressive	9.4	12.3	11.4	10.1	9.9	10.4
strength, psi Confined expansion	155	167	232	254	237	186
1500 F, in./in. Hot strength,	.0152	.0157	.0157	.0152	.0162	.0150
1000 F, psi Hot strength,	147	130	115	113	108	94
2500 F, psi	305	162	168	195	209	280

NOTE: Molding sand mixtures contained sand, 4 per cent bentonite, 4 per cent fireclay. Above tests conducted at Canadian Bureau of Mines by A. E. Murton. blown out the top of the pipe. In the production model a conical hood was added for the purpose of collecting the cleaned sand as it blows out of the top of the cleaning chamber. Since the sand is following a circular path, it is obvious that it will be thrown outward into the conical chamber. Fines and dust rise into the dust collecting pipe.

Wear-resistant iron castings were used in the cleaning chamber. Standard sandblasting nozzles were used in place of the air syphons. The equipment as now used consists of two identical units arranged so that the sand passes from one to the other. It was found necessary to screen the shakeout sand before putting it in the hopper which feeds the reclaimer.

With the production model, all of the sand moves upward and comes out of the top of the cleaning chamber. Slag, iron, and other large foreign objects drop to the bottom of the cleaning chamber and are removed periodically.

One unit as described above, will produce from 150 to 200 lb of reclaimed sand per hour. The unit will operate continuously, and require very little attention. Table 3 shows test results from cores made with sand reclaimed in the production model.

#### Castings Made from Reclaimed Sand

The reclaimed sand has been used regularly in molding and core sand mixtures. The surface of these castings was identical to surfaces of castings made in new sand mixtures. The reclaimed sand has proved to be a satisfactory material.

Uniformity of Reclaimed Sand: In a small jobbing foundry the sand coming from the shakeout varies widely in properties. At one time it may contain a burnt core sand with very little clay. Another time it may consist of returns from high silica flour mixtures. Because of the varying nature of the used sand, it can only be used sparingly in molding and core sand mixtures. Therefore it is customary to have an excess of used molding sand which must be disposed of, and new sand brought into the foundry.

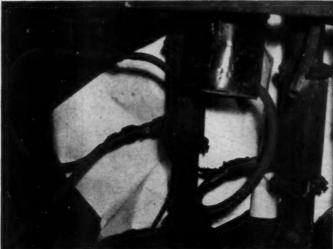




Fig. 3 (above)—Lower part of sand reclaimer. Sandblast nozzles force sand and air into the chamber.

Fig. 4 (above, right)—Intake end of reclaimer. Suction feeds sand through rubber hose.

Fig. 5 (right)—Exhaust end of the sand reclaimer. Dust particles and fines travel upward into dust collector system.

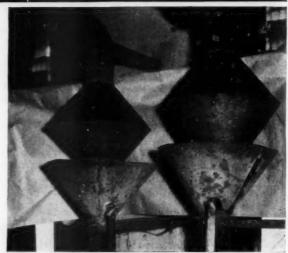
Reclaimed sand drops into the lower cone.

The sand reclaimer produces a sand with fairly consistent properties. During a 48-hr run tests were taken hourly from the production model sand reclaimer. Each test consisted of mixing the reclaimed sand to 4 per cent moisture content in a laboratory sand mixer. Green compression tests ranged from 0.9 to 1.9 psi, and permeability was within the range 100 to 130. During this period all types of molds used for 5- to 4000-lb castings were shaken out. It was concluded that the reclaimed sand was consistent enough in properties so that it could be incorporated in standard mixtures for molding and core sands.

TABLE 3-New, Used, and Treated Core Sands (Figs. 2, 3, 4 and 5).

	New	Used Sand	Used Sand Treated	Used Sand Treated
Properties	Sand	Untreated	Twice	3 Times
Moisture, %	3.9	5.0	5.0	4.8
Green deformation,				
in./in.	_	0.029	0.034	0.050
Green compression				
strength, psi	0.5	5.6	2.5	2.0
Density, lb/cu ft	106	104.5	104.5	105
Green mold hardne	ess			
after 3 rams	20	78	60	53
Green permeability	59	36	60	77
Baked permeability	90	56	100	140
Baked compressive				
strength, psi	1464	127	175	923
Baked tensile				
strength, psi	262	9	33	122.5

Note: Treated sands were put through production model reclaimer. Mixtures contained 1000 grams sand, 10 grams cornstarch, 30 grams linseed oil, 40 grams water.



#### Discussion

Permeability: Used sand varied in permeability depending upon the amount of silica flour contained. Reclaimed sand had a permeability approximately the same as the base sand used in the sand system.

Green Deformation: As Table 3 shows, reclaiming increased the green deformation of sand mixtures. This was due to the removal of bonding agents.

Green Compression Strength: A simple method of evaluating the reclaimed sand is to mix it with water and measure the green compression strength. When the green strength was under 2 psi the process was considered to be operating satisfactorily for producing a material to be used in molding sand mixtures.

Green Mold Hardness was lower when reclaimed sand was used. This is in line with the reduction in green strength.

Dry Compression Strength: As Table 2 shows, molding sand made with reclaimed sand had higher dry strength than similar mixtures using new sand. From this it was inferred that some bonding material remained in the reclaimed sand.

Expansion of Molding Sand: Tests conducted at the Canadian Bureau of Mines indicated that expansion

characteristics of the sand were not altered by the reclaiming operation.

Hot Strength at 1000 F: As Table 2 shows, molding sand made with reclaimed sand had a lower 1000 F hot strength than was found in all-new molding sand

Hot Strength at 2500 F: All-new molding sand had a 2500 F hot strength of 305 psi (Table 2). Molding sand made with shakeout sand had a hot strength of 162 psi, or half the strength at 2500 F. Reclaimed sand also had lower hot strength at 2500 F.

As far as high temperature properties are concerned, it is evident that the air-scrubbing treatment as used at this plant does not produce a sand identical in refractory properties with new silica sand. However, the reclaimed sand apparently is satisfactory for

Screen Analysis: Table 4 shows the effect of air scrubbing on the particle size distribution of sand. The reclaimer removes coarse particles (12 and 20 mesh). Pan-sized material was reduced from 2.0 per cent to 0.2 per cent by reclaiming. Three quarters of the clay in the shakeout sand was removed by air scrubbing it twice.

Baked Strength of Cores: The foregoing remarks have been mainly concerned with molding sand mixtures. When reclaimed sand was to be used in core sand mixtures, its suitability was evaluated by mixing it into a standard core test mixture, baking sample cores, and measuring their tensile or compression

It was evident from the work done in the plant that a material for use in molding sands could be produced by passing shakeout sand twice through the reclaiming chamber. In order to produce a suitable core sand, however, the used sand had to be treated to twice the amount of scrubbing. Since there was a great demand for molding sand, the final production reclaimer was set up to produce a partially reclaimed sand for this purpose.

Table 1 indicates that it is possible to produce a core sand by air-scrubbing operations on shakeout

#### Conclusions

(1) The non-uniformity of sand coming from the shakeout in a jobbing steel foundry made it necessary to use this material sparingly in sand mixtures.

(2) By passing the shakeout sand through an airscrubbing device and blowing out most of the fine material, a product of consistent properties was ob-

(3) The reclaimed sand proved to be satisfactory as a substitute for new sand in molding sand mix-

(4) The reclaimed sand produced in the plant did not have properties identical with new sand. Some fines and bonding agents remained in the sand. These impurities did not appear to have any deleterious results upon molding sand mixtures, or upon steel castings made in these molding sands.

(5) In order to reclaim a sand for use as a core sand, the amount of work performed upon the material must be two or three times that required for molding sand.

TABLE 4-SCREEN ANALYSIS OF USED AND RECLAIMED SANDS

	R	etained on Screen	1, %
Screen Size	Used Sand, Untreated	Used Sand, Treated Twice	Used Sand, Treated 3 Times
12	0.8	0.2	0.2
20	0.6	0.4	0.4
30	0.8	0.6	1.0
40	3.6	4.6	8.4
50	10.6	13.4	22.6
70	37.0	38.8	39.0
100	29.0	32.6	22.8
140	5.6	5.8	3.6
200	1.8	1.2	0.6
270	0.2	0.4	0.2
pan	2.0	0.2	0.2
A.F.S. Clay	8.0	1.8	1.0

(6) The low initial cost, and the low labor cost of operation, make air scrubbing equipment of interest to the small foundry.

(7) Further investigation on the air scrubbing treatment of used sand might prove of interest to the foundry industry.

#### Bibliography

- 1. E. C. Jeter, "Reclaims Ferrous Foundry Sand," AMERICAN FOUNDRYMAN, Feb., 1949, pp. 40-44.
- 2. E. C. Jeter, "Sand Reclamation for the Ferrous Foundry," The Foundry, Aug., 1946.
- 3. J. M. Cummings and W. M. Armstrong, "Foundry Sand Reclamation," Canadian Metals, July, 1948.
- 4. Roy W. Bennett, "Wet System Reclaims Foundry Sand," AMERICAN FOUNDRYMAN, May, 1949, pp. 58-61.

#### DISCUSSION

- Chairman: J. B. CAINE, Consultant, Wyoming, Ohio. Co-Chairman: H. W. DIETERT, Harry W. Dietert Co., Detroit. Recorder: R. G. THORPE, Cornell University, Ithaca, N. Y.
- C. E. Wenninger: 1 High air velocities tend to break down the sand grains.
- V. M. ROWELL: 2 What is the appearance of the recovered
- MR. FAIRFIELD: The recovered sand grain surfaces were polished but were not white and clean. Apparently some of the binder was retained on the surface of the grains. However not too much oil was required to rebond the sand.
- MR. ROWELL: Is the reclaimed sand high in gas evolution? MR. FAIRFIELD: The reclaimed sand would be gassier than
- E. C. TROY: 8 Evidence accumulated by a method of mechanical agitation several years ago indicated that:
- a. Clay or oil bonds destroyed by the heat of casting represented but a portion of the total bonded grains.
- b. During mechanical agitation, the burned bonding material chaffed free of the sand grains more quickly and easily than the unburned bonding material.
- c. Removal of material other than the burned bonding substance seemed unnecessary and wasteful.
- d. An optimum and more economical removal of bond material seems more desirable as a goal than does complete removal of the old bonding substance.
- e. A minimum amount of new bond is necessary for re-use if either some clay or oil in the unburned condition remains as a residue on the grains.
- f. The final conclusion drawn from this work was that selective removal of burned bonding substance was not only desirable but possible.

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 Archer-Daniels-Midland Co., Cleveland.
 Foundry Engineer, Palmyra, N. J.

# REFRACTORY PRACTICE AS APPLIED TO MALLEABLE DUPLEXING

By C. W. Meyer*

#### ABSTRACT

In order to render a clear understanding of refractory practice, the author discusses the various materials used in a charge and their effects upon the refractories used in the cupola.

The improvement that has been made in the refractory practice by experimenting with various types of tuyeres and melting zone linings is discussed. The forehearth, transfer ladle and electric furnace are described as to their design and length of life as determined by the different refractories used.

In a study of the refractory practice in a foundry, it is first necessary to describe the method of charging the cupola and the materials used in a typical charge. This charge consists of approximately 55 per cent remelt which is placed in the cupola with the use of an overhead jib crane and a cylindrical bucket with a conical-shaped drop bottom, shown in Fig. 1. The design of the bottom distributes the sprue to the inside circumference of the cupola. This bucket also handles the coke which is used at approximately a 1 to 10 ratio (1 lb of coke to 10 lb of metal). Silvery pig (11 per cent), and limestone are loaded on a specially designed gasoline powered dump truck and charged, as illustrated in Figs. 2 and 3. About 33 per cent hydraulic pressed steel bundles, along with manganese briquettes, make up a 10,000 lb charge. Due to the variation in size and weight of the bundles, the charging and placing of the charge is very important because of flues which may develop and cause unnecessary burn-out of the upper lining in the cupola and loss of carbon and silicon. This not only makes extra repair necessary, but also results in off analysis iron.

The zone of the cupola in which charging occurs must be built with a refractory block that will withstand the constant hammering and shock as each charge is put into the cupola. This is illustrated in Figs. 4, 5 and 6. These blocks must have a heat resisting characteristic and also a resistance to abrasion. This area of the cupola must have close inspection and careful maintenance to prevent development of

holes in the lining, which could cause bridging and also result in the falling in of blocks, causing a heavy slag. A high quality, hard, burned block was tried in this zone and found to chip off and break out, making it unsatisfactory. When all the charging was done manually, cast iron blocks were used in the charging zone of the cupola, but since the mechanical charging was installed, the sprue hammering against the cast iron blocks caused them to loosen and drop into the cupola. Therefore, that practice was also discarded as unsatisfactory.

At the present time, a soft sand block that gradually wears away is being used. This method has proved to be the best and lasts four to five weeks.

#### Cupola Melting Zone

For years the lining of the bosh section of the cupola has presented a problem due to the burn back and resultant hot spots on the shell. This condition not only increased refractory cost but hindered the melting operation. As the blocks burned back, some would slide off and go down with the charge. When these blocks were in the well, there was the possibility of plugging the tap hole and causing considerable trouble for the operator and restricting the melting operation.

In an effort to improve this condition, almost every refractory known to the industry was tried, including both commercial and plant mixes. Two years ago, air-placed refractory was tried and produced satisfactory results in relation to the life and cost of this essential repair. This method is currently being used.

In order to obtain a more vivid picture, it will be necessary to compare the old and new methods of construction and discuss the features of each. Formerly the melting zone in the cupola was built completely of a preshaped, pressed, Missouri and Kentucky fireclay block (Fig. 7). These blocks were laid in a double course against the water glands and the cupola shell. A fireclay, mud slurry was used on the vertical face of the block at the joint, and the horizontal face was laid dry. The blocks were supported across the continuous tuyeres by gray iron plates. These plates were cast to the circumference of the

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Fig. 1-Charging cupola with drop bottom bucket.

Fig. 2-Loading dump truck with pig iron and limestone.



Fig. 3—Charging pig iron and limestone into cupola.

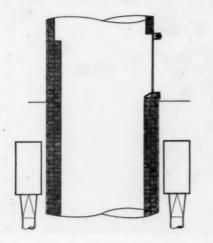


Fig. 5—Cross-section through cupola charge door.

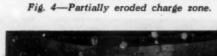




Fig. 6-Completely rebuilt charge zone.

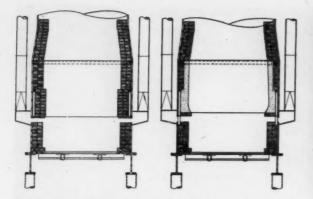


Fig. 7 (Left)—Cross-section through cupola, old method; Fig. 9 (Right)—New method.



Fig. 8—New type tuyere.

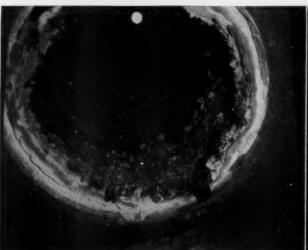


Fig. 10-Cupola after being used.



Fig. 11-Cupola ready for air placement of lining.



Fig. 12-Cupola ready for bottom.



Fig. 13-Completely relined cupola.

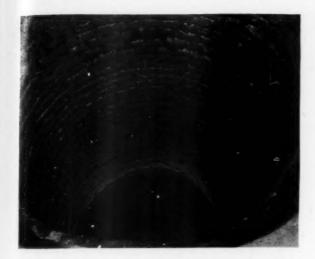


Fig. 14-Cupola ready for lighting of bed coke.

cupola in 36-in. lengths. Burned out portions were removed each day and replaced with new sections. Some of the refractory was removed through this process, although it may have had further use had it been left in. When the tuyere plate failed by burning through, the refractory immediately above it would begin to slide down and cause heavy slag, a sticky tap hole and possibly hot spots on the shell due to the thinner lining.

The first change was made in the tuyere section (Fig. 8). A system of intermittent tuyeres was developed without the use of the gray iron tuyere plate. Blocks were laid so that there was a solid section of brick between each tuyere which supported the upper courses of block. This eliminated sliding of the upper refractory with a resultant improvement in the melting operation and an improvement in the cost of refractory as it was no longer necessary to remove blocks that had further use.

The above-mentioned change in tuyere design did not affect the analysis of the iron, and it gave a more penetrating tuyere blast which, with the charge materials used, was very effective in melting more iron in the cupola. The type of tuyere described has proven to be very successful. Later, air-placed refractory became available and experiments were conducted in the melting zone with this new type of lining (Fig. 9).

Based on past experience with the block cupola, it was decided that a bosh would be necessary if a monolithic lining was to be used. The 3-in. water glands behind two courses of brick caused an offset with the 12-in. lining; as a result, the cupola was 6 in. smaller in diameter before the change to monolithic in the area that is now formed as a bosh.

After making the change from block lining to airplaced refractory, a bosh was constructed, and poor melting practice resulted. The cupolas again had heavy slag, sticky tap holes and some hot spots on the shell. A progressive daily study was then made of the burn-out. This was done through measurements of location and depth of burn. It was found after composite graphs of each cupola were made, using figures obtained over a two-week period, that each cupola burned out in the same area. A dual reading thermometer was then installed to check water temperature as it entered and left the water gland. It was noted that after approximately 3 to 4 hours, the temperature of the water leaving the gland showed an increase. One morning, due to an interruption of the gas supply, it was necessary to drop the cupola after only 3 hours of operation. It was then proven by direct observation that more than half of the 12-in. lining had already been burned out. It was also determined after all other conditions were as equal as possible in the cupola, that the lining appeared to burn back about the same amount, regardless of the original thickness. With these facts established, it was decided to cut down the thickness of the lining over the glands.

As the thickness of the lining was progressively reduced, the recurrence of the trouble due to heavy slag decreased until now it occurs very rarely. At the present time we have no bosh at all, and the thick-

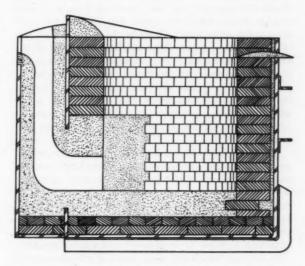


Fig. 15-Cross-section through spout and forehearth.

ness of the lining 4 in. above the tuyere is 7 in., tapering upward and inward to the double coursed shaft of 12-in. thickness.

The reduction of lining thickness at the melting zone reduced the amount of refractory needed for daily repair, thus reducing the cost of refractory, as well as eliminating the resultant troubles.

This modified contour lining, which was made possible by air placing the monolithic lining, was not practical when using block because of the difficulty of installation. Three distinct advantages of using modified contour lining are:

- Cheaper cost per ton of iron melted due to smaller quantities of refractory material used, and the re-use of that portion of the lining which is not burned out.
- Elimination of the major troubles caused by lining failure.
- 3. Increased size of melting zone at no extra cost.

The current practice of cupola repair is illustrated in the following figures. Figure 10 shows a cupola as it appears after being used. Figure 11 shows a cupola ready for the air placing of the lining. Figure 12 shows the cupola ready for the closing of the bottom. Figure 13 shows a completely relined cupola. Figure 14 shows a cupola ready for the burn-in of the bed coke.

#### Forehearth

In the practice of duplexing malleable iron it is desirable to use a forehearth or holding ladle. It is here that the iron is collected, desulphurized and mixed. The desulphurization is accomplished with the addition of 3 lb per ton of sodium carbonate. This addition of sodium carbonate produces a highly corrosive slag which, in turn, erodes the refractory block used for building a forehearth.

As the foundry grew and melting capacity increased, it was necessary to enlarge the forehearth from 9 tons to the present capacity of 15 tons. Before the change from a 9-ton forehearth to a 15-ton forehearth, the entire holding portion was made of 9-in. keys and straight brick, while the spout was constructed with monolithic refractory. These bricks were laid on the sidewalls, using a very high quality refractory slurry on the joints to insure compactness. The joints of the brick, particularly in the breast area, had to be repaired after the first few days' operation. Unless there was a close inspection daily and repairs made where necessary, runouts were sure to occur, or slag would come up the spout through the burnt out breast.

To minimize this constant repair, the areas which necessitated the most repair were replaced with a rammed material in the following manner (Fig. 15). The ends and back of the forehearth were laid as before, and the breast was rammed with monolithic refractory by the use of an air hammer. To do this, the forehearth must be tilted on its side so that all ramming is done in a pressure-on-hammer position. The bottom, which has two courses of refractory brick



Fig. 16-Tapping forehearth into transfer ladle.

laid in a cross joint pattern, is built up to an 11-in. total depth with air-hammer rammed monolithic refractory. This refractory is made of commercial bond (neutral to basic) and western ganister (1/8-in. to 1/4-in. screen), thoroughly mulled with enough water to give a uniform 6 per cent moisture content.

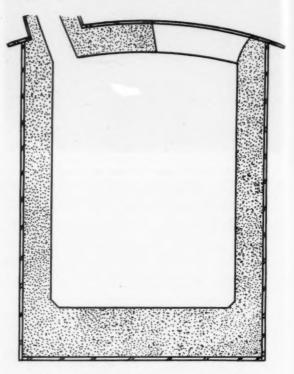


Fig. 17-Cross-section through transfer ladle.

The spout is of the teapot variety and is air-hammer rammed. The formula of monolithic refractory for the spout differs from the above mentioned in that an acid bond is used in place of the neutral to basic bond. Today, after processing approximately 5,500 tons of iron through the forehearth, it is completely rebuilt. This, of course, is made possible by daily checking which has greatly decreased the number of minor repairs necessary, as compared with the previous practice.

#### Transfer Ladle

The transfer ladle is used to carry iron from the forehearth to the furnace for further refining, blending and heating, before delivery to the foundry. This ladle is a monorail type ladle and carries slightly under 4 tons to the furnace each trip. The ladle travels at a height of 5 ft above the floor and must be as safe and dependable as possible (Fig. 16). Here, again, air-rammed refractory has superseded the old method of lining the shell with brick and tile. This ladle is built to hold a given amount of iron when it is put into service. No allowance is made for burnout, because if the proper practice is followed, it will give the desired service necessary (Fig. 17). The transfer ladle has a covered top with two spouts, one on



Fig. 18-Transfer ladle and form.

each side, and after seven to nine days' use, it is removed to be rebuilt.

The ladle is rebuilt with a combination of commercial ramming bond and ganister, which is mixed sufficiently with enough water to give a uniform 6 per cent moisture content. This material is air-hammer rammed, around a form in the ladle shell, until it rings. The material is rammed in, layer upon layer, until the shell top is reached. The form is then removed, and a sectional top form is installed. The top is rammed, and the form pieces removed through the larger of the two spouts. A steel cover is placed on top and bolted down on the shell.

The ladle is placed in an oven for 24 to 48 hours at a temperature of 300 F to 400 F. All ladles are dated and none are removed unless a minimum of 24 hours, and preferably 48 hours, has elapsed since the time of entry (Fig. 18). After baking, a slow gas fire preheats this ladle 3 to 4 hours before it is put into service.

As yet, none of these linings have burned through, although when the length of service has been extended, hot spots were noticed. Upon examination of these trouble areas, it was found that the remaining 3/4 in. of lining has been the cause of these hot spots. This remaining thickness will still allow adequate time to remove the ladle from service.

#### **Electric Furnace**

The furnace used in the plant is of the three-phase, direct-arc type with acid lining. It is 11 ft in diameter and has a 15-ton load or bath capacity. The operation of the electrode arms and the furnace tilting are done hydraulically.

In a cross-section of the furnace bottom, illustrated in Fig. 19, there are two courses of fireclay wedge brick. The first layer is put in and the second is laid on top of it in a cross-jointed pattern. These courses are laid following the contour of the shell, to a point level with the rivet line of the furnace. Both courses

are laid with a fireclay slurry. The sidewalls are built next. They are composed of No. 1 and No. 2 silica key brick of the standard 13-in. series. These bricks are laid up dry using a sharp silica sand between the joints and courses. A ganister mixture is used to level off at the top of the shell and at the spout locations.

The next step is the installation of the bottom refractory which consists of a mixture of 2 parts sand, 2 parts clay and 7 parts stone, using equal parts of sodium-silicate and water for bonding. This mixture is packed to a level depth of 7 in. and the top is then but on.

A gas burner is placed inside the shell and also one underneath the bottom on the outside. The furnace is heated until the interior attains a cherry red glow. After a dry-out period of approximately 8 hours, short pieces of carbon are placed under the electrodes so as to make a uniform arc around the inside of the furnace when the power is turned on.

The burn begins and continues until the stone on the bottom commences to melt and fuse together. During this process the top and sidewalls also are fused, and the siliceous material that drips and runs off is of benefit to the bottom. After there has been sufficient fusing, the electric current is turned off. The furnace is allowed to cool. A layer of unbonded siliceous stone (1/4 in. to 1/2 in.) is shoveled in and the process is repeated. This continues until the desired thickness is obtained (18 in. to 22 in. total). During the burn-in, the electrodes used for arcing must be moved from time to time to get a uniformly developed bottom. Extreme care must be used to prevent electrodes from freezing into the bottom.

To complete the burn-in after the desired bottom depth is obtained, the furnace must be allowed to cool so that the bottom can harden and excess gas escape before addition of the iron. This takes from 3 to 4 hours. The bottom, if properly maintained, will last indefinitely. The proper maintenance consists of completely draining the furnace at the end of each day and sanding the shallow holes. Holes that are beyond sanding repair are repaired with silica stone without the use of any bonding agent.

The sidewalls must be renewed periodically, and the method of replacing this is as described above. The furnace is usually operated 6 weeks before sidewall repairs are necessary. About 15,000 to 18,000 tons of iron pass through the furnace before repair-

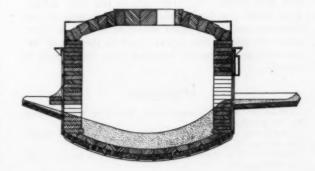


Fig. 19—Cross-section through electric furnace.



Fig. 20-Furnace before rebuild.

Fig. 21-Furnace after rebuild.

ing. Figures 20 and 21 show before and after repair. A third shift is engaged on the furnace operation so that proper maintenance is given to the furnace, and all minor repairs are made when the pouring operation is over for the day.

#### Furnace Roof

The furnace roof, Fig. 22, is constructed of silica brick which is designed and manufactured to form concentric rings. They are assembled on a roof form designed to meet the brick manufacturer's specifications for spacing. If the bricks are not correctly spaced, there is a possibility of spalling, gaps or brick cave-ins. These bricks are encircled with a cooling pipe at the area inside and next to the furnace top ring.

The life of the roof is dependent upon the method of heating employed. A gas burner is placed in the spout and a natural gas flame is burned slowly to preheat both sidewall and roof brick. It is impossible to reach the operating temperature by this method, but through preheating the thermal shock is reduced to a minimum.

The only time the furnace roof is removed is when sidewall repairs are necessary or the life of the furnace is depleted. After completion of such repairs and ascertaining the future life of the roof, if any, it is either put back into service, preheated in the same manner as described above, or discarded. Very often the life of this top is equal to the life of the sidewalls and is usually replaced when the furnace is rebuilt.

#### DISCUSSION

Chairman: W. G. FERRELL, Auto Specialties Manufacturing Co., St. Joseph, Mich.

Co-Chairman: ERIC WELANDER, John Deere Malleable Works,

East Moline, Ill.
Recorder: Wm. Zeunik, National Malleable & Steel Castings Co., Indianapolis.

R. N. SCHAPER: 1 Is the stone here referred to used in the bottom sand or ganister and what is the grain sizing?

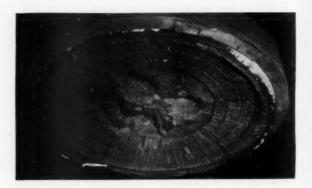


Fig. 22-Furnace top.

MR. MEYER: It is 1/8 to 1/4-in. ganister, added dry and burned

H. E. LEICKLY: 3 Have you ever experienced any silicon pickup after repairing the burned out spots with sand?

MR. MEYER: Not since the sand is burned in.

R. W. Heine: 8 How do you measure lining thickness and contour when gun placing a lining? Do you depend upon the operator?

MR. MEYER: It is measured every day by measuring the thickness through the tuyeres.

MEMBER: Have you tried air placement of linings in ladles? MR. MEYER: No, we did not.

MR. SCHAPER: How do you repair spots or holes in the furnace bottom at night in the electric furnace?

MR. MEYER: The furnace is drained, the washed and dried sharp sand is thrown on the spots needed and it fuses in place. CHAIRMAN FERRELL: In burning-in a furnace bottom how deeply is the ganister fused in?

MR. MEYER: It is fused approximately 3 in. The rest of the bottom is semi-fused.

R. S. HAMMOND: 4 How much burnout do you get in a 7-in. section of melting zone lining above the tuyeres?

Mr. Meyer: From 3 to 31/2 in. W. R. Jaeschke: 4 In one of your illustrations one section of this lining appears to be burned down to the water jacket.

MR. MEYER: The thickness of lining remaining at this spot was about 1 in.

Wisconsin Appleton Co., South Milwaukee, Wis.
 Fanner Manufacturing Co., Cleveland.
 University of Wisconsin, Madison, Wis.
 Whiting Corp., Harvey, Ill.

R. H. GREENLEE: 5 Why was the use of cast iron blocks in the charging zone area discontinued?

MR. MEYER: Because they became loose and fell out.

R. N. ZOLLER: 6 Have you tried rammed or hand-placed as well as gun-placed linings?

MR. MEYER: Yes, but the tests were not exhaustive.

MR. HAMMOND: Have you tried long cast iron slabs in the charging area instead of cast iron blocks?

MR. MEYER: Yes, but they burned off at the bottom.

MR. LEICKLY: Do your temperatures get that high so near the charging door?

MR. MEYER: Yes, red hot. We have a 120-in. cupola and the brick get red clear to the charging door particularly at the end of a heat.

MR. ZOLLER: Have you tried basic refractories?

MR. MEYER: No, we did not; cost was one factor, and basic linings require continuous melting to keep the lining hot to prevent spalling.

Mr. ZOLLER: We use basic linings and costs have been cut 30 per cent over a 2-yr period.

CHAIRMAN FERRELL: How do you account for the saving?

MR. ZOLLER: In labor.

Mr. JAESCHKE: Mr. Zoller is speaking of gray iron practice

and Mr. Meyer is speaking of malleable practice.

This is a very important point. In basic practice one of the claimed advantages is higher carbon pick-up whereas in malleable duplex melting we work for a minimum carbon pick-up. We wonder what experience anyone else has had.

J. T. Mackenzie: 7 We produced iron in basic-lined cupolas

with as low as 2.50 per cent carbon.

Mr. JAESCHKE: The records I have seen were from very high steel mixtures and in malleable duplex melting we generally use 55 per cent sprue scrap and must allow for some silvery pig or equivalent and this leaves us a maximum possibility of some 40 per cent steel in the mixture. It is a question whether we can get low enough carbon from mixtures with a maximum of 40 per cent steel.

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DR. MACKENZIE: We melted mixtures with as much as 40 per cent returns. We must remember that chemistry of slag can control the carbon pick-up. We have obtained carbons in gray iron from 4.30 to 3.20 by varying the steel content and adjusting slag analyses. With slags of low viscosity, the sulphur runs higher but if this does not matter we could use basic-lined cupolas with slags of low basicity. With basic-lined water-cooled cupolas we can run the slag basicity down almost to acid slag and get any carbon range.

MILTON TILLEY: 8 In basic cupola melting, reference is made to the basic slag cleaning the coke more efficiently thereby giving higher carbons. Would Dr. MacKenzie elaborate on their experience?

Dr. MacKenzie: That is true.

MR. ZOLLER: Do you get greater carbon pick-up from the coke or from the carbon atmosphere?

DR. MACKENZIE: From the coke.

MR. HAMMOND: What is the distance from the bed plate to the charging door on your cupola? Have you done anything to reduce the temperature at the charging zone at the end of a

MR. MEYER: The distance is 21 ft. No, we did not try to reduce the temperature. We stop charging about 1 hr before dropping bottom.

J. H. RICKEY, JR.: What type of refractory do you use for tap-out block in a 16-hr operation?

MR. MEYER: We use a fire clay tap-out block. A gun placed block was not satisfactory.

Auto Specialties Míg. Co., St. Joseph, Mich.
 Zoller Castings Co., Bettesville, Ohio.
 American Cast Iron Pipe Co., Birmingham, Ala.

⁶ National Malleable & Steel Castings Co., Cleveland.
⁹ Ironton Fire Brick Co., Ironton, Ohio.

## ACTUAL APPLICATION OF MOTION PICTURES IN FOUNDRY MOTION AND TIMESTUDY

By

M. T. Sell*

To make use of the camera for motion study, we must, of course, have the proper equipment and follow a few simple rules if we are to secure useable motion studies. It is assumed that the reader is interested in this subject and acquainted with timestudy techniques; how a detailed elemental study is made; how the study is converted to normal times; how the normal times are transferred to a comparison sheet, then to the charts; the curves developed, relaxation added, and finally developed into useable Standard Data Tables. Direct standards may be determined from each individual motion study the same as from a stop-watch study, but with such valuable and detailed information available, it would indeed be wasteful not to use the information for future standards. This, of course, means developing the necessary tables for all the elements of work relative to each task, thereby permitting the setting of standards by what we term "Standard Data" before the job is made.

The film shown in connection with this paper was on the making of small cores on work benches suitable for that purpose. This film is one of the motion studies used to develop the writer's Standard Data Tables for bench coremaking in 1946. Those same tables, with a few minor corrections for methods changes, are still in use today.

At the time this motion study was taken, the operators were working on many jobs which had standards determined by direct stop-watch timestudy. In order to have more accurate and consistent standards, and to make possible the setting of standards at a pace which would permit 100 per cent coverage, it was decided to develop Standard Data Tables. Upon examination of the timestudies it was found that because some of the elements were of such short duration, certain constant values and other very short elements of work had been, in many instances, combined and recorded as a single element of work when the stop-watch studies were made. It was then that the decision was made to use the camera, which would

permit as fine a break-down of the work elements as might ever be required. Bench coremaking is not a fast operation when compared to many assembly operations, etc., but there are certain elements which are practically impossible to segregate when using a stop watch. For example, the element "strike-off," and the element "jab vent." I believe you can readily see the trouble one would experience while attempting to get an accurate stop-watch reading on that portion of the element which we term the "constant value," such as grasping the trowel or the vent wire and bringing them into working position. We also have the constant values for the return of these tools to their place on the work bench.

#### **Advantages of Motion Pictures**

By using the camera, we were able to determine these constants and separate them from the variable portion of the overall element with ease and accuracy. It was our experience that when this was attempted with the stop-watch, we were inconsistent and invariably on the loose side, simply because we could not record the times with any real degree of accuracy.

Film is the perfect record. You cannot quarrel with the film. What the operator did—what the person did—what the crew did—is there on the film. If there are any questions as to what was done while you were making the study, it is there on the film. You can bring them up again at any future time for discussion or for analysis. Each film loop is a permanent, unquestionable record of operations readily available for reference at any time. It replaces interpretations and memories by permanent facts. The film tells an impersonal positive story, undimmed by months or years. They are a record and a piece of history.

The analysis of a camera study is not complicated. In fact it is very simple, and when completed there is absolutely no doubt that the element times represent the exact time taken by the operator.

To analyze the study, the projector, table screen and loop stand are set up in a darkened room. The counter on the projector is set at zero, and the pro-

^{*} Standards and Payroll Supervisor, The Sterling Foundry Co., Wellington, Ohio.

TABLE !

	Elements	Indicated Reading	Elapsed Time	Indicated Reading	Elapsed Time	%	100%	Allow
1	Clean Sand From Work Space			13	.013			
2	Waste to Box			28	.015			
3	Wipe Box with Waste			127	.098			
4	Waste Away			134	.007			
5	Dusting Bag to Box	13	.013	163	.029			
6	Dust Box	81	.068	198	.035			
7	Dusting Bag Away	86	.005	202	.004			
8	Assemble Box	116	.030	227	.025			
9	Pick up Loose Piece	133	.017	239	.012			
0	Clean Loose Piece	143	.010	244	.005			
1	Position Loose Piece	165	.022	291	.047			
2	Assemble Box	227	.062	390	.099			
3	Check Position of Loose Piece	238	.011					
1	Clamp to Box	259	.021	420	.050			
5	Clamp Box	437	.178	513	.093			
	Fill with Sand	470	.033	547	.034			
7	Jolt Box (one jolt)	490	.020	565	.018			
8	Mallet to Peening Position	510	.020	583	.018			
9	Peen	588	.078	667	.084			
0	Mallet Away	593	.005	672	.005			
1	Level Sand	606	.013	685	.013			
2	Get Shaped Rod (one)	625	.019	730	.045			
3	Position Shaped Rod	752	.127	842	.112			
4	Fill with Sand	778	.026	857	.015			
5	Rammer to Ramming Position	795	.017	871	.014			
6	Ram (Butt-off)	854	.059	934	.063			
7	Rammer Away	863	.009	942	.008			
3	Trowell to Strike-Off Position	878	.015	956	.014			
9	Strike-off	950	.072	1016	.060			
0	Reposition Box	986	.036	1076	.060			
1	Vent Wire to Position	1027	.041	1095	.019			
2	Vent (4 jabs)	1054	.027	1117	.022			
3	Vent Wire Away	1064	.010	1123	.006			
1	,	1004	.010	1139	.016			
	Move Trowell	1079	.015	1156	.017			
ò	Mallet to Rapping Position	1116	.037	1240	.084			
	Rap Box	1119	.003	1248	.008		4	
3	Mallet Away Trowell to Strike-off Position	1119	.003	1264	.016			
)	Strike-off			1274	.010			
)			-	1281	.007			
,	Trowell Away			1375	.094			
	Reposition Clamp	1154	.035	1409	.034			
2	Position Plate			1452	.043			
	Roll Box	1198	.044	1483	.031			
	Remove Clamp	1241 1248	.043	1491	.008			
,	Mallet to Rapping Position	1248	.007	1527	.036			
	Rap Box	1305	.013	1536	.009			
	Mallet Away			1649	.113			
	Draw One-half of Box	1458	.153		.082			
	Draw Other Half of Box	1560	.102	1731	.040			
	Remove Loose Piece from Core	1590	.030	1771				
	Touch up Core	1044		1826	.055			
	Lift Core and Plate from Bench	1644	.054	1850	.024			
	Core and Plate to Turntable	1734	.090					
	Return to Bench	1774	.040					

jector is turned by hand. Starting with the first element, the projector is turned until we arrive at the desired "breaking point" between it and the second element. We may turn the projector both forwards and backwards past the breaking point to arrive at the desired movement. Observing the frame counter, we read the number of frames registered and record them opposite the element on the analysis sheet. The projector is then turned until the next element of work is completed, and the accumulated total of frames recorded opposite the second element. This is continued throughout the entire study, breaking the

study into as many work elements, constants, and variables as desired. After analyzing a motion study, the ease and accuracy with which the breaking points between the elements can be determined, as compared to a stop-watch study, can be easily appreciated. On short cycle work this is a very important factor in determining accurate standards.

Recording the accumulated readings from the frame counter, gives the analysis sheet the appearance of a continuous reading stop-watch study, except that the values are in increments of thousandths of a minute. The element times are determined by simple sub-

TABLE 2

Patt. No. 1356-1 Customer	Box No. 2 W&S Co.	Std	Bench 2 Cores/Box		Floor Cores/Catg.	1	C. O. Type Box		Box Reg. Split	
Part Name			H/T 57/	Corr	. Factor	Class _	Factored	by AF-	VB Date	3-14-46
	Meas.	Std.			Meas.	Std.			Meas.	Std
Sand In & Ram	27/8"	.33	Bedding				Nails-Lay	in Reg.		
Box Value	27/8	.75								
Projected Area	53/4x33/8-194	.23					Lay in Ir			
Strike Off Area	53/4x21/2-144	.08	Cinder Relie				On End			
Slick Area			Rods-Wires				On End	Reg.		
Lug or Recess			1-9 1/2 Ton	end		.32				
1-1B		.20					On End	Irreg.		
							Clamps Tie Wires			
Dowell Pins										
Vents-Rod										
Jab	4	.10								
Groove								2		
Scratch							Spray Oil			
Wax							Rod Time			
Oil Spots							Total Per E	Box		
Oil Edges							Adjustment			
Hooks							STANDARI	)		2.01

traction, and the number of frames, which represents the actual time taken by the operation, is recorded for each element description. We know that each frame represents one thousandth of a minute because the camera used for the study is electrically driven and takes exactly 1000 frames per minute. With the camera running constantly throughout the study, all of the operator's time is accounted for and there can be no errors on the element times caused from misreadings or any loss of time.

The effort rating or speed rating of the operator does not have to be made at the time of the study. This is because the projector can be set to project the picture at the same rate of speed that it was taken. This permits the study to be rated at any future time and by anyone desired. When we use the stopwatch method, the only persons qualified to rate the operator are those who may be actually watching the operator at the work station. When we use the camera, the film can be shown to individuals or groups who may be interested in the rating or the methods, and there is no need to rely on someone's memory for the information.

Motion studies are invaluable in the training of timestudy observers, as well as for reviewing at intervals to keep their thinking in line with the original conception of standard. Many difficult problems which in all probability would have become major grievances have been settled amicably by management and union, after sitting down and viewing the film which showed the operator working under the actual conditions, using certain methods, and working at a given pace. Management could actually duplicate all of this on the screen before them.

Observing the analysis sheet (Table 1), we find a list of all the element descriptions. In the first column opposite each element description is recorded the accumulated reading from the frame counter on the

projector. This reading is taken at the completion of each element. After the study has been completely analyzed, the readings are subtracted and the elapsed time recorded opposite the element applicable. As each frame on the film equals 0.001 minutes, the elapsed time is expressed accordingly—0.013, 0.068, etc.

You will observe several elements listed (Table 1) the elapsed time for which is so small that they can not be accurately separated and recorded when using a stop-watch. Most of these elements are constants, or that portion of the overall element which is present regardless of the amount of variable. Using the camera, the ease and accuracy by which these values can be determined can be readily observed. We are thus enabled to determine the true constants, and then plot only the values on the chart which are truly relative to the variable being used.

For example, let us observe the overall element of "jab vent." A common understanding of timestudy technique tells us that elements No. 31 and No. 33 are constant values regardless of the number of vents required. You will also observe that on this particular study, the two constant values for this overall element are greater than the time required to perform the variable, or the four jab vents. This is indicated by the small amount of elapsed time for element No. 32. If the job required 40 or 50 jab vents, then the time required to perform the variable would be greater than the constants. It is for this reason that the constants must be separated from the variables if we are to allow enough time on the four jabs, and not an excessive amount for the 40 or 50 jabs.

We may be inclined to think that the individual constants for these various elements are only trivial and that we are wasting time when we separate them as is demonstrated on the analysis sheet (Table 1), but when we add the values for all the constants pres-

ent in this particular study, they represent 0.268 minutes of the entire cycle. This is approximately 15 per cent of the total overall time taken to make the core. As the job becomes more simple, the overall time decreases, while many of the constants remain, thus making it still more important and necessary to be as nearly accurate as is humanly possible on these small allowances if we are to have consistent standards.

Table 2 is a copy of the core factor card. For this particular core it is self-explanatory as to the necessary measurements, information, etc. The values ap-

pearing on the card are the standard times allowed for the various elements or combination of elements as taken from the Standard Data Tables. The total of these standard allowances represents the Total Standard for the box.

The uses to which the camera may be applied by the Standards and Methods Engineer are many such as methods improvement, left and right hand movement, body movement, training programs, etc., making the camera one of the most useful tools available to the Timestudy Engineer.

### EFFECTS OF CERIUM ON GRAPHITE FORMATION IN ALLOY CAST IRON

Edward A. Rowe* and Howard A. Johnson**

#### ABSTRACT

This paper is concerned with the effect of cerium on an alloy cast iron containing nickel, manganese, and molybdenum. The investigation was conducted to observe the changes in graphite structure, tensile strength, and hardness caused by the cerium additions. The double treatment, addition of a secondary graphitizing inoculant with the cerium, is also discussed.

The structural engineering uses of gray cast iron have been restricted by its limited tensile strength and ductility. Recent research has shown that the mechanical properties of this metal are greatly improved if the graphite is in the nodular or spherulitic form. Processes which result in the production of nodular graphite structures in the as-cast condition involve the use of cerium or magnesium.

The problem studied and reported in this paper is the nodulizing effect of cerium on a specific cast iron.

#### Review of Literature

Nodular graphite structure, first observed in malleable cast iron, contrasts with normal flake structure in that the graphite is in spherulites or nodules rather than in the normal elongated-flake form. Some of the disadvantages of malleable cast iron are the section limitations and the necessity for very close composition control to permit satisfactory annealing. Carbon structures similar to those in malleable iron can be obtained in the as-cast condition merely through the use of small ladle additions of cerium.

The general action of cerium in cast iron as discussed by Morrogh and Williams4 is first to desulphurize the melt and then to inhibit the carbide decomposition. Retardation of graphite formation until after the melt has built up interior pressure forces the graphite into the stable, spherical configuration. If the internal pressure of the melt becomes greater than that of the forming graphite, the carbon is retained as a carbide, and the cast iron is susceptible to chilling. The harsh carbide stabilizing action of cerium can be tempered by secondary graphitizing inoculants such as alloys of the ferrosilicon type.

There are three general theories of inoculant behavior. These are discussed in a paper by Lownie:3 (1) The theory that gases cause structural changes in metals and that the inoculant acts as a degasifier; (2) the graphite nuclei theory that the inoculant facilitates the formation of graphite nuclei; (3) the silicate slime theory that foreign material is provided as a nucleus. Which of these theories is correct is not known.

Morrogh and Williams* state these rules for consistent results:

- 1. The iron must solidify gray.
- 2. The iron must be hypereutectic; that is, the







2% Nital Etch

2% Nital Etch 1500X

Plates I and II-Melt E; 0.04 per cent Ce added, 0.027 per cent Ce retained. Tensile strength, 61,800 psi; Brinell hardness, 300.



Note: This manuscript is taken from a thesis submitted in partial fulfillment for a Master of Science Degree, University of Washington, Seattle, Wash.

Professor of Metallurgy, University of Washington, Seattle.

^{**} Engineer, Hanford Engineer Works, Richland, Washington.

TABLE 1-ANALYSIS OF CAST IRON USED

Element							1	Percent	age
Carbon								2.25-3	3.20
Sulphur							*	, (	0.03
Nickel					*		×		2.00
Mangane	ese	*			*			0.75-	2.50
Phospho	rus							. 1	0.02
Silicon								. :	2.00
Molybde	num							. 1	1.00
Iron							F	Remain	der

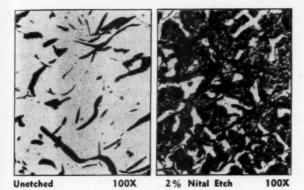


Plate III—Untreated cast iron. Tensile strength, 67,400; Brinell hardness, 310.

percentage of carbon must be greater than 4.3 minus  $\frac{1}{3}$  of the combined percentages of silicon and phosphorus.

- The silicon should be between 2.3 and 7 per cent.
- 4. The sulphur should be below 0.02 per cent.
- 5. The phosphorus should be below 0.5 per cent.
- The effects of manganese, copper, nickel, chromium, molybdenum, and vanadium are negligible so long as the previous conditions are met.

The alloy cast iron used in this investigation was hypoeutectic of composition shown in Table 1.

Three deviations from the foregoing rules are con-



2% Nital Etch 100X

Plate IV — Melt H; 0.01

per cent Ce added, 0.007

per cent Ce retained. Tensile strength, 67,200 psi;

Brinell hardness, 402.



Plate V — Meit V; 0.04 per cent Ce added, 0.035 per cent Ce retained; 0.3 % Si added as 85% terrosilicon. Tensile strength, 73,000 psi; Brinell hardness, 363.

Unetched

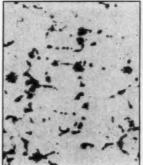


Plate VI—Melt 1W; 0.04 per cent Ce added, 0.035 per cent Ce retained; 1.0 % Si added as 85% ferrosilicon. Tensile strength, 76,400 psi; Brinell hardness, 352.

100X

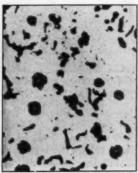


Plate VII—Melt 1V; 0.04 per cent Ce added, 0.026 per cent Ce retained; 3% Si added as 85% ferrosilicon. Tensile strength, 31,-500 psi; Brinell hardness, 321.

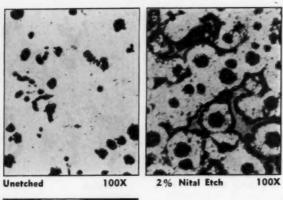
TABLE 2-EXPERIMENTAL DATA AND TEST RESULTS

							Tensile Streng	th	
	Pouring	Ceri	um, %	Silie	con, %	Treated Sample,	Control Sample,	Strength Ratio,**	Hardness,
Melt	Temp., F	Addition	Retained	inoculant*	total	psi	psi	%	BHN
A	2670	0.10	0.075	_	2.02	unmac	hinable		475
E	2720	0.04	0.027	_	1.96	61,800	67,300	91.8	300
H	2730	0.01	0.007	-	1.93	67,200	77.250	87.0	402
I	2750	0.02	0.013	-	1.93	74,700	77,400	96.5	407
M	2780	0.04	0.035	0.03	1.75	unmac	hinable		475
Q	2730	0.04	0.032	0.04	1.82	70,600	70,900	99.6	321
V	2780	0.04	0.035	0.3	2.53	73,000	79.000	92.4	363
X	2780	0.04	0.023	0.5	2.72	80,400	78,800	102.0	310
1V	2900	0.04	0.026	3.0	5.87	31,500	55,250	57.0	321
1W	2790	0.04	0.035	1.0	3.20	76,400	55,800	137.0	352
AA	2920	0.05	0.017	3.0	6.17	20,000	72,700	27.5	363
BB	2890	0.04	0.018	3.0	5.92	25,000	73,500	34.0	341
CC	2750	0.04	0.024	1.5	4.38	73,600	72,200	102.0	375
EE	2820	0.05	0.026	2.0	5.10	33,200	73,000	45.5	302
FF	2620	0.04	0.018	2.0	4.85	42,600	72,800	58.5	341

• Secondary inoculant; to melts M and Q as CaSi; to remainder as 85% FeSi.

•• 100×T.S. (treated)

T.S. (untreated) = per cent tensile strength.





Plates VIII and IX—Melt
AA; 0.05 per cent Ce
added, 0.017 per cent Ce
retained; 3 per cent Ce
added as 85 per cent ferrosilicon. Tensile strength,
20,000 psi; Brinell hardness, 363.

2% Nital Etch 500X

tained in Table 1. The carbon equivalent of the alloy cast iron used is below 4.3 per cent, the silicon is 2.00 per cent, and the sulphur is slightly high. Carbon is the most serious discrepancy, since the carbide stabilizing effects of cerium cause susceptibility to chilling in hypoeutectic cast irons, as previously stated. The silicon deficiency is partially overcome by the presence of 2 per cent nickel, and additional silicon may be added as an inoculant. The excess sulphur is not serious to the problem although it would increase the cost, since the cerium acts preferentially and desulphurizes before it stabilizes the carbides.

#### **Experimental Procedure**

In order to minimize the possible effects of other metals, especially the rare earths, the purest cerium available was used. This analyzed 92.6 per cent cerium, 1.1 per cent iron, and the balance other rare earth metals. Misch-metal* has been suggested for commercial use provided it is magnesium-free.4

All test melts were taken from regular heats. The alloy cast iron was melted in basic electric furnaces to control composition. The furnace charge was made up of high grade pig iron, foundry scrap, mild steel, coke and alloy additions of manganese, silicon, nickel and molybdenum. Melting practice consisted of melting down, slag removal, oxidizing with iron ore or mill scale, again removing slag, and then superheating to 3000 F. Oxidizing conditions are very important during the superheat. As the melt was tapped, a 1

per cent addition of a nickel alloy inoculant (a 60 per cent nickel, 30 per cent silicon, 10 per cent iron) was added in the preheated ladle.

A portion of the heat was tapped into a shank ladle of 165-lb capacity. The alloy addition, cerium plus any secondary inoculant, was added to the metal as it was tapped. It is necessary to add the cerium before or with a secondary inoculant to obtain the best results.⁴ Stirring action of the pouring metal insured proper mixing. Nine cylindrical test bars

 $(1\frac{1}{2})$  by 15 in.) were poured.

The pouring temperatures shown in Table 2 were measured as the metal was poured into the molds. These temperatures varied, due to exothermic reaction of the secondary inoculants. The molds were made from dry sand cores mounted in one unit and poured from a central downgate. Since the physical properties of the alloy cast iron used vary slightly from heat to heat, an untreated sample was poured for each melt. The effect of the cerium was evaluated by the difference in physical properties of the untreated and the treated samples.

Standard 0.800-in. tensile specimens were machined from the lower portions of the test bars according to ASTM specifications. These were pulled to destruction at a loading factor of 0.4 in. per min. The broken specimens were used for Brinell hardness

tests and metallographic examination.

Drill shavings were taken from the test bars for analysis (Table 3). Because of the small percentages of cerium, colorimetric determinations were run. This determination consists of dissolving a standard sample in hydrochloric acid, oxidizing the solution, adjusting the pH, electrolytically removing the iron, manganese, nickel and molybdenum, reducing the cerium, re-adjusting the pH, and extracting the cerium with a reagent of 8-hydroxyquinoline dissolved in chloroform.

The absorption of the cerium-8-hydroxyquinoline solution when using a green filter (540 millimicrons) is measured in a Klett photometer and the percentage of cerium read from a calibration graph. Full details of the method are given by Westwood and Mayer.6

Standard determinations were run for the other elements.

The first series of tests were made to investigate

TABLE 3-CHEMICAL ANALYSIS OF MELTS

Melt	Components, %										
	C	Mn	P	S	Si	Мо	Ni	Ce			
A	2.65	1.40	0.024	0.031	2.02	1.01	1.90	0.075			
E	2.41	1.57	0.034	0.033	1.96	1.00	1.95	0.027			
H	2.69	1.47	0.026	0.026	1.93	1.01	2.64	0.007			
1	2.69	1.47	0.026	0.025	1.93	1.01	2.64	0.013			
M	2.57	1.74	0.025	0.020	1.75	1.10	1.87	0.035			
Q	2.64	1.44	0.036	0.017	1.82	0.96	1.86	0.032			
Q	2.52	1.35	0.020	0.015	2.53	1.00	1.71	0.035			
X	2.52	1.35	0.020	0.017	2.72	1.00	1.71	0.023			
IV	2.77	1.33	0.019	0.013	5.87	1.00	2.01	0.026			
1W	2.77	1.33	0.019	0.018	3.20	1.00	2.01	0.035			
AA	2.47	1.34	0.033	0.017	6.17	1.02	2.27	0.017			
BB	2.47	1.34	0.033	0.015	5.92	1.02	2.27	0.018			
CC	2.47	1.34	0.033	0.013	4.38	1.02	2.27	0.024			
EE	2.47	1.34	0.033	0.018	5.10	1.02	2.27	0.026			
FF	2.47	1.34	0.033	0.009	4.85	1.02	2.27	0.018			

^{*}Misch-metal, an alloy containing 45 to 52 per cent cerium, 45 to 48 per cent other rare earths, 0.5 to 2.5 per cent iron, and traces of silicon, aluminum, and calcium.

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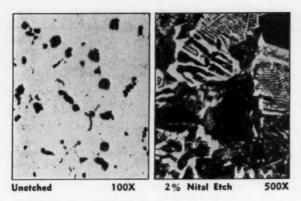
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TABLE 4—EXPERIMENTAL RESULTS FOR CERIUM ADDITIONS

Melt /	Additions, %			Properties					
	Cerium Re-		Inoculant		Tensile Strength			Plate	
	Added	tained	%Si	Source		BHN		No.	Remarks
A	0.1	0.075			_	_	. White Iron	-	Literature gives 0.1% as upper limit of Ce addition However, white iron was obtained. There was a pro- nounced inverse chill.
E	0.04	0.027	-		61,800	300	Mottled iron in out- side bars, gray iron in center bar. Distinct cast structure with acicular pearlite and carbides in matrix.		Strength lower than the untreated specimen due definitely oriented Type E* graphite. Rosette patternoted, indicating solidification prior to nodule form tion. Martensitic regions in Plate II verify rapid chil ing. Melt indicates an upper limit of 0.04% Ce und the test conditions.
Untreated	_	-	_		67,400	310	Type A Graphite	III	Cast structure is similar to that of Plate I.
Н	0.01	0.007	-		67,200	402	Coarse, well-oriented Graphite structure.	IV	Decrease in tensile strength in structure of this kin caused by planes of weakness. Etching** reveale matrix of acicular pearlite and carbides.
1	0.02	0.013	-	_	74,700	407	Imperfect Rosettes	-	Grouping of graphite flakes produced imperfect rosette similar to type B graphite. Fine matrix shown outching.
M	0.04	0.035	0.03	Ca Si		-	White Iron	-	Amount of silicon added as secondary inoculant wattoo small.
Q	0.04	0.032	0.04	Ca Si	70,600	321	Gray Iron	-	Graphite structure did not fall into any standar classification. Etching revealed fine, acicular matrix.
v	0.04	0.035	0.3	85% FeSi	73,000	363	Quasi-flakes of graphite – some small nodules.	V	Difficulty in control of the powerful action of calcius silicon led to use of 85% FeSi. Chill areas were visible on etching, indicating lower limit of FeSi addition. Tensile strength was lowered by chill areas.
X	0.04	0.023	0.5	85% FeSi	80,400	310	Similar to V.	-	Chill areas were eliminated, and the size of the carbid areas was reduced.
1W	0.04	0.035	1.0	85% FeSi	76,400	352	Very fine quasi-flakes and nodules.	VI	On etching, this melt showed an extremely fine matri with small areas of carbides.
1V	0.04	0.026	3.0	85% FeSi	20,000		Considerable change due to 3% Si addition. True nodules appeared for the first time, sur- rounded by ferrite areas.	VII	Metal was very friable and brittle, making it highly machineable, but of little value because of low tensil strength. Hairline cracks were noted in the ferrit areas. The cause of the cracks is not known, but the were undoubtedly the cause of the brittleness.
AA	0.05	0.017	3.0	85% FeSi	20,000	363	Almost entirely nodular.	IX	Etching revealed structure similar to IV but a net phase, thought to be steadite was noted. The steadit is not explainable, since the phosphorus was very low Plate IX contains two graphite nodules surrounded ferrite, in turn bounded by pearlite. The light mottle area surrounded by pearlite is steadite. Crachs again caused low strength.
ВВ	0.04	0.018		85% FeSi	25,000	341	Similar to AA.		This heat was made to determine reproducibility or results of AA. A similar structure resulted and again apparently, a new phase (Plate X). Plate XI shows the skeleton-like structure resembling material described as steadite. Solidifying out of the carbide areas. Why this ingredient presents such a pattern on solidification is not known.
CC	0.04	0.024		85% FeSi	73,600	8	Fine, unoriented flake graphite mixed with some malformed no- dules.	1	Tensile strength very high compared to AA and BI containing nodular graphite. The differences in matrices explains this increase, Plate XII showing acicular pearlite surrounding decarburized areas of ferrite.
EE	0.05	0.026		85% FeSi	33,200		Similar to preceding Melts.	CIII I	Low tensile strength.
FF	0.04	0.018		85% FeSi	42,600	341	Similar to EE.	8	There is an interesting comparison between this melt and CC. A variation of 0.5% silicon added in the econdary inoculant resulted in a difference of nearly 100% in tensile strength.



Plates X and XI—Melt BB; 0.04 per cent Ce added, 0.018 per cent Ce retained; 3% Si added as 85% ferrosilicon. Tensile strength, 25,000 psi; Brinell hardness, 341.

the effect of various cerium additions on the physical properties of the alloy cast iron. The second series investigated the effects of cerium plus a secondary inoculant. The results of these experiments are presented in Table 4.

#### Summary

Cerium affected the graphite structure in all melts. Cerium alone tended to produce oriented graphite of Type E, which lowered the tensile strength. A pronounced chilling effect was noted in melts treated with cerium alone. Desulphurization was not completed to the extent reported by Morrogh and Williams.⁴

The addition of a secondary innoculant with the cerium improved the tensile strength and graphite structure. Both the type and amount of the secondary inoculant influenced the amount of acicular pearlite produced. In the alloy cast iron the maximum tensile strength is associated with the amount of acicular pearlite and the size of Type A graphite.

Nodular graphite structures were obtained in the alloy cast iron, but at the expense of modifying the

acicular pearlite matrix.

The most desirable physical properties were obtained in the range of 0.04 to 0.05 per cent cerium with 1.0 to 1.2 per cent silicon added by ferrosilicon as a secondary inoculant.

#### Acknowledgments

The authors are indebted to the Pacific Car and Foundry Co. and to Dr. Alexander Finlayson, technical director, and Herbert A. Conyne, metallurgist, for use of foundry facilities, and assistance in the con-

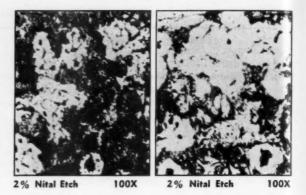


Plate XII—Melt CC; 0.04 per cent Ce added, 0.024 per cent Ce retained; 1.5 per cent Si added as 85 per cent ferrosilicon. Tensile strength, 73,600 psi; Brinell hardness, 375.

Plate XIII — Melt EE; 0.05 per cent Ce added, 0.026 per cent Ce retained; 2 per cent Si added as ferrosilicon. Tensile strength, 33,200 psi; Brinell hardness, 302.

duction of the investigation. Grateful acknowledgment is made to the Engineering Experiment Station, University of Washington, sponsor of the project.

#### References

1. Metals Handbook, 1948 edition, American Society for Metals, Cleveland.

2. R. H. Heyer, Engineering Physical Metallurgy, D. Van Nostrand & Co., New York, 1939.

3. H. W. Lownie, Jr., "Theories of Gray Cast Iron Inoculation," TRANSACTIONS, American Foundrymen's Society, vol. 54, p. 837 (1946).

4. H. Morrogh and W. J. Williams, "The Production of Nodular Graphite Structures in Cast Iron" Journal of the Iron and Steel Institute, March, 1948, pp. 306-322.

W. Rosenhain, Introduction to Physical Metallurgy, p. 310,
 D. Van Nestrand & Co., 1914.

6. W. Westwood and A. Mayer, "The Determination of Cerium in Cast Iron," Analyst, May, 1948, vol. 73, no. 866, pp. 275-282.

#### DISCUSSION

Chairman: W. H. WHITE, Jackson Iron & Steel Works, Jackson, Ohio.

Co-Chairman: CARL HARMON, Hanna Furnace Corp., Buffalo. Recorder: C. C. Sigerfoos, Michigan State College, East Lansing.

J. J. Brenza: 1 How do you explain presence of hairlines in the microstructures?

Mr. JOHNSON: As far as we can tell those hairlines are actual cracks in the structure.

CHAIRMAN WHITE: Those hairlines might possibly be due to the metallographic polishing technique.

¹ Koppers Co., Inc., Baltimore, Md.

#### APPRENTICE TRAINING—IT IS NEEDED

By

Maurice C. Sandes*

#### ABSTRACT

A study of the amount and kinds of apprentice training in foundry work has revealed such a lack of interest and planning on the part of industry that a large scale shortage of skilled molders and coremakers will fall upon the industry with great impact.

Regardless of the continued trend for production work with the aid of molding machines, core blowers, etc., and the resultant use of specialist workers who can be quickly trained, there is still the need for journeymen molders and coremakers who can be expected to tackle the unusual, the intricate and the large molding problems with confidence and dispatch.

There are relatively few programs in the country that train such workers. There must be more of them. Most states, especially those in the great industrial regions of the nation, are weefully short of the required number of apprentices needed to keep the proper amount of skilled workers available to the industry. Without the people responsible for the continued flow of sound salable castings, the industry will face a continued loss of prestige, influence and business.

Apprentice training is the answer. The youth of the country must be given the opportunity to earn their livelihood through proper training. This training must be complete; the opportunity to go to the top of their trade must be made to them. The disclosure of the small number of apprentices being trained, the realization that some organizations, large and small, are operating carefully planned apprentice training programs and are satisfied that it pays, should "spark" the rest of the industry and set in motion a nation-wide plan for more foundry apprenticeship.

With the growing shortage of skilled foundrymen becoming more apparent, a greater interest must be shown apprentice training by both industry and the government. The primary objective of the apprenticeship program is to train efficiently, to the degree of competence ordinarily expected of journeymen, the proper number of youths to meet the needs of industry for workers in skilled occupations. This has not been done to the amount nor degree of quality.

The Foundry Industry is an expanding industry and large numbers of men are needed every year to train for this work. Too often the training is inadequate and haphazard. Since World War II, apprentice training has grown but not to the extent that it should. There are still too many organizations who recognize the value of organized and well rounded training but are too willing to "let George do it,"

while they continue with specialist training which, though adequate for them is not a credit to the area or nation-wide industry.

The Foundry Industry still needs mechanics who can make intricate castings; who can secure a hidden flange, make up a charge for a cupola, and figure the weight of a large casting. For such men, a long range training program is a necessity. It should not be a hit or miss proposition. Apprentice training is needed and will pay off if properly planned and carefully administered. Short-term training has been tried in many places and it has paid off for some organizations. However these people who had the short-term training need further training when they attempt to work at their trade at another plant or in another locality. The modern youth wants job security and knows that it is gained through complete knowledge of his work. Dissatisfaction with the industry is very often the result of partial training.

#### Apprentice Training Statistics

Some of the statistics on foundry apprentice training are not generally known. From the records of the Bureau of Apprenticeship of 1950² a total of 1136 registered apprentices is shown in training for the trades of molder and coremaker. The status of registered apprenticeship in molding and coremaking is somewhat different than the true figure of training being done. Numerous large foundries operate programs outside the National Apprenticeship Program, and generally these programs substantially conform to the federal standards except for joint establishment. In many cases they are the finest examples of apprenticeship training and are a credit to the people who started them and to those who are continuing these programs. These programs would account for perhaps another three hundred in training. The government establishments, such as the naval shipyards which have civilian apprenticeship opportunities, could be said to have another fifty. There are other foundries who have apprentices but with no planned or supervised training program. The students are used solely as a source of labor and profit. Any figures on apprenticeship training from such a situation must be excluded as they would not apply.

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[•] Foundry Training Coordinator, Mare Island Naval Shipyard, Vallejo, Calif.

The total figure, then, was 1486 apprentices in training for molding and coremaking. A round figure of 1500 can be used safely. At this time, 5405 establishments operated foundries or foundry departments. This shows an average of less than 0.3 apprentice

per foundry! This is an appalling figure.

To get another view of the situation, we should know the ratio of apprentices to journeymen in the trade. Figures show approximately 70,000 molders and 30,000 coremakers in the continental United States. Not all this number perform work at a level which requires apprenticeship. Perhaps 35,000 molders and 18,000 coremakers are now available for jobs requiring skills and knowledge at the craftsman level. Using the figures of 1500 apprentices and 53,000 journeymen, a ratio of one to thirty-five is derived! Normal loss of apprentices from training before completion of the apprenticeship is close to 25 per cent during these times of great employment opportunities and high wages. This widens the ratio even more. The ratio of one to eight as set by the International Molders Union would make a figure of about 6600 apprentices when using 53,000 as the journeyman figure. The Bureau of Apprenticeship has stated that the total registered apprentices represent but 17 per cent of the required apprentice employment. This is less than a fifth!

#### Graduation of Apprentices

At this time, it can be pointed out that only one-fourth of the total number of apprentices can be graduated and utilized as journeymen every year. This means that out of 1500 apprentices about 400 could be available to the industry in 1951. Four hundred to be spread between 5400 shops!

Contrary to what might be expected there is no section or region in the U.S. which leads the rest of the country in foundry apprenticeships. Indeed the facts are baffling and rather point out that more is done by states than by any federal agency on planning. Consider that Regions VI and VII which include Kentucky, Michigan, Ohio, Illinois, Indiana, and Wisconsin (who have a total of 1947 foundries or nearly two-fifths of the total foundries in the country), have a registered total of only about 8 per cent of the apprentices they need to make the proper ratio. These regions are considered the great industrial section of the country. On the other hand, a section not generally thought of as highly industrial is Region XIII, Idaho, Oregon and Washington (who have a total of 148 foundries or slightly under 3 per cent of the total foundries in the country). This region has registered 63 per cent of the apprentices they need to make the proper ratio.

To pin-point this situation, a comparison can be made of adjoining states in a region. For example, Georgia has 136 per cent of the registered apprentices needed to balance the journeyman-apprentice ratio whereas Alabama has only 1 per cent! Again, for example, Rhode Island has 3 per cent but Massachusetts, her neighbor, has 43 per cent.

Can this perhaps be because some shops were not equipped for training? This is not so. A survey of over 4000 foundries by the field staff of the Bureau

of Apprenticeship found that 55 per cent were adequately equipped to train apprentices. These adequately equipped shops amounted to 2339. However, only 562 of these were training under registered standards. The remaining 1777 shops constitute a great potential for additional development. It is indicated that the gap between required apprentice employment and actual employment can be closed only by the development and registration of systems in foundries which are now failing to train although adequately equipped.

This development is being done in one area of California, largely through the efforts of the A.F.S. Chapter there. Of the states on the West Coast, California has by far the greatest amount of foundries and the most apprentices registered though nowhere near the percentage needed. There are 422 shops in the state and the greatest number of these are in the southern part around Los Angeles. The San Francisco Bay Region, however, with about 60 shops in a relatively small geographical area is carrying the training

load and very efficiently too.

A survey of the Los Angeles Area showed very little organized training of apprentices. Most shops showed very little interest in such a program. Several shops had a post-war program which had failed due to several reasons and were reluctant to start again. Others were interested but were not convinced that it was possible to train apprentices and not lose money by it. (These same reasons apply all over the nation). Even the schools were indifferent towards foundry courses. One high school had a small foundry course which was used in the exploratory program.

The Pasadena City College has a foundry course and a small shop. Besides serving the engineering classes, it is attempting to produce trained men for actual foundry work. Registration for this course is small as very few young men are sufficiently interested in going into the foundry industry. Funds are lacking to properly equip the shop. Not enough enthusiasm is being shown by the industry in the area.

In the San Francisco Bay Region, as was already mentioned, training was well thought of. The local city governments and the schools backed it because it meant better citizens, people with the background which enabled them to find work easily, people who were more contented, more secure, who were better educated. The labor organization backed it because it meant supply of properly trained skilled workers, and a finer potential of supervisory personnel.

#### Joint Apprenticeship Council

The Joint Apprenticeship Council (known as JAC), composed of labor, management and school representatives and a government representative has in this region two committees, one in the San Francisco and peninsula side and one in the East Bay side which includes the cities of Oakland, Alameda, Berkeley, Richmond and several other smaller communities. Both areas are very active and the shops always have the proper ratio of apprentices to journeymen. The JAC meets once a month during school months and takes up the matters of indentures and complaints of the apprentice, management and school.

The East Bay Area boasts three night school classes for trade extension. One in the Berkeley Evening High School and two in the Laney Trade and Technical Institute in Oakland. These are trade extension classes held two nights a week from 7:00 to 9:15 p.m. The classes in both schools include classroom work for related information and actual shop practice.

The instructors of these classes are journeymen molders and coremakers whose work experience and general knowledge of the trade have been approved by the Joint Apprenticeship Committee (JAC) of the area. This committee has been very instrumental in furthering these classes by scheduling speakers to talk on foundry subjects on the level that the apprentice can understand and appreciate. The management members have made available to the classes films on foundry work and related subjects. With the help of the committee, an outline for a book for teaching the related information of the trade to apprentices was formulated. This book, called a workbook, and its companion, a test-book, was written and has been printed by the State of California. Now in use in the East Bay Area classes, it has simplified the instruction and is doing a much better job in that a more evenly progressive rate of learning is offered to the apprentice while in the trade extension classes.

Educating the apprentices to be receptive to such a program is highly necessary. Attendance of these classes is mandatory of the apprentices in order for them to obtain their Certificate or Diploma of Completion of Apprenticeship. In fact, the apprenticeship agreement may be dissolved by the Committee if proper excuse is not given for the non-attendance of classes. However, under the stimulus of good teacher ability, proper classroom facilities and progressive learning, attendance has become voluntary and a much more eager spirit has been noted among the apprentices of the area.

#### Planned System of Training Mandatory

In this state's 422 foundries, only 38 per cent of the amount of apprentices needed are registered in training. However this figure is well over the average of 17 per cent shown for the entire nation. This indicates a planned system of training, at least in part of the state. The preceding system about the Bay Area has been high-lighted in order to show how much can be done with a well integrated system. This has been the result of harmony between labor and management, combined with the willingness and know-how of the school officials. The same thing, perhaps better, can be accomplished anywhere in the country. Some of the reasons why it is not being done will be listed later.

As have been before, there are large companies which operate foundries who have fine apprentice programs especially worked up. Examples of these, to name but a few of the most well-known ones, are Caterpillar Tractor Co., Peoria, Ill., Bethlehem Steel Corp., Sparrows Point, Md., and the Newport News Shipbuilding & Dry Dock Co., Newport News, Va. These organizations have splendid examples of apprenticeship training programs. The foundry apprentice training at the U. S. Naval Shipyards are won-

derful examples too. Here the end product is not just a molder or a coremaker but a skilled foundryman capable of handling any and all of the departments in the foundry.

#### Obstacles to Overcome

It is not expected that all foundries can equal such programs but they can approximate them with the help of the proper agencies. There are obstacles to overcome in order to establish apprenticeship on the scale needed. These obstacles come under five main headings:

- (1) Obstacles created by management.
- (2) Obstacles created by attitudes of unions.
- (3) Obstacles created by conditions in the industry.
- (4) Obstacles created by attitudes of apprentices or applicants.
- (5) Obstacles created by conditions outside the industry.

Under the first heading, Obstacles created by management, can be listed thirteen subheadings. The first three and most prevalent are:

- (a) Management has not yet become sufficiently interested.
- (b) Management objects to participation in government-sponsored program.
- (c) Management believes supply of journeymen is adequate.

The fallacy of subheadings (a) and (c) are shown partly by this paper. "American Industry—is overlooking the importance of building up and maintaining the supply of highly skilled craftsmen. These skilled workers have been the backbone of American superiority in manufacturing and to them must be attributed our sense of know-how and Yankee ingenuity." 3

Under subheading (b) there has never been a valid reason stated for the objection to participation in government-sponsored programs.

The second series of obstacles, that is those created by unions, has four subheadings. The most important of these is:

(a) That unions have not yet become sufficiently interested.

In the third group, those created by conditions in the foundry, are nine subheadings of which four are listed here as the most prevalent:

- (a) Lack, or irregularity, of work.
- (b) Almost all jobs are now semi-skilled.
- (c) Piece rates and short-term training of specialists.
- (d) Applicants avoid trade because of physical working conditions.

In the fourth group, those obstacles created by attitudes of apprentices or applicants, are five headings of which one is dominant and that is:

(a) The wage scale is considered too low to attract the proper kind and number of applicants for apprenticeship.

The fifth and final group has three subheadings of which two are the most prominent and important.

- (a) Shortage of field representatives for promotional work.
- (b) Field staff has not given the industry its share of promotional attention.

The course that must be taken to build up the skilled working force in our nation is by training. It has been agreed that systematic training in apprenticeship is the most satisfactory guarantee of maintaining a skilled force. The first thing that must be done then is to "sell" both labor and management on a nation-wide basis. It must be a cooperative enterprise. Promotional literature must be distributed to local industry. Trade magazines should be encouraged to publish more articles on apprenticeship. The Bureau of Apprenticeship should provide a larger staff and it should be educated to the importance of the foundry industry. Schools should be encouraged to give more information about the industry to students and in foundry centers to operate a small shop or laboratory. Where there are classes that include foundry work or where trade extension classes are already operating in schools, all efforts should be made to keep courses up to date and have competent instructors.

#### Hackley Manual Training School

This last point has been well taken in some places although, unfortunately, not duplicated everywhere. As a model, the Hackley Manual Training School, Muskegon, Michigan is beyond a doubt one of the finest.4 There a group of foundries found from a survey that 22 per cent of the employed in Muskegon worked in foundries or allied enterprises, yet there was no training available. The school there had a foundry shop but it had not been used in 18 years. This foundry group rehabilitated this shop at a cost of \$20,000. Modern layout, safety and cleanliness were stressed. Mechanical equipment included an electric melting furnace, molding machines, shotblast equipment, complete tools for the students, mechanical sand mixing equipment and a laboratory complete with sand testing equipment. The final step was to see that a competent teacher-instructor was secured.

This fine school foundry is used to the best possible advantage. The students receive training in mechanical drawing, patternmaking, foundry work and machining and finishing. By the time twelfth grade is finished, instructions in nearly all phases of foundry work and laboratory work has been given. Basic physics and chemistry, a special time and motion analysis, elements of metallurgy and cost planning are included in order to cover completely the training.

From such a school lay-out, students are produced who can go into the foundry as molder and coremaker apprentices, as laboratory assistants, as sales trainees or into other trades or fields with a fine background that gives them an understanding of the versatility of the trade and an appreciation of the importance of the industry.

This school program can be duplicated anywhere in the country. Foresightedness is the key. The foundrymen in this one area used this key. It was a wise investment. The results from it will begin to show in ten or fifteen years by the number of young men who will be in responsible jobs and positions because of having had proper training. These are the men who will secure the industrial prominence of this country. For by the act of initiating better

training of their future workers, the industry will reap the benefits derived therefrom.

Various training programs may be selected to fit the problems posed in any given area. One of the finest is the trade exploratory course at the high school level. Another, and the next step up from this, is the preparatory course, either full or part time, night or day school. This may be a pre-apprenticeship program in which the student may receive credit for his studies and work towards completion of apprenticeship. Trade extension courses for the youth already in an apprenticeship is designed to give all the related instruction needed and possibly some actual shop practice when a shop or laboratory is available. The needs of the community will decide which type or combination of training program is needed; however, the procedure for planning is the same for all. The following points are recommended:8

- (1) All local industries should determine the need for vocation-technical training.
- Determine the facilities available for training.
- (3) Plan, develop and conduct training programs. (4) Organize instructional material.
- (5) Secure and keep competent teacher-instructors.
- (6) Publicize the program.

#### References

- 1. "Basic Information Concerning Apprenticeship and the participation of the Public Schools in the Program,' State Dept. of Education, Sacramento, Calif. (1948).
- 2. Registered Apprenticeship in Molding and Coremaking," U. S. Dept. of Labor, Bureau of Apprenticeship, Research & Review Division (1950).
- 3. T. F. Hammer, Editorial, AMERICAN FOUNDRYMAN, Jan. 1951. 4. "Michigan Foundrymen Restore Trade School Foundry Program," AMERICAN FOUNDRYMAN, March 1949. Also W. G. Gude, "Training Young Foundrymen," The Foundry, March
- 5. Vocational Education in the Years Ahead, U. S. Office of Education, Federal Security Agency, Vocational Division, Bulletin No. 234.

#### DISCUSSION

- Chairman: G. J. BARKER, University of Wisconsin, Madison. Co-Chairman: W. J. HEBARD, Continental Foundry & Machine Co., East Chicago, Ind.
- Recorder: W. J. HEBARD. FRANK SHIPLEY (Written Discussion): 1 Mr. Sandes is to be
- commended on his paper which points out one of the salient weaknesses in the industry; namely, the lack of training for key personnel who will share the responsibility for the future of the industry. In the past ten years the "shot gun" courses of specialist training have been in vogue which, of course, causes regular training to be slighted. It is high time that we all think seriously of our past mistakes and concentrate on a greater effort in training in order to supply qualified personnel for future needs.
- As Mr. Sandes points out, most schools are indifferent to foundry programs. Only a few schools have small foundry facilities and even in these the foundry subject is touched on lightly.
- We, in the foundry industry, must correct this by making the students more interested in the foundry. We proclaim to the world that "The foundry is a good place to work." Let's be sure we make this a reality and we will have less difficulty in interesting young men in it. A good starting point, of ccurse, would be to convince the instructor. He could well be included in local A.F.S. chapter membership even if the local chapter must pay his dues. Helping to set up small non-ferrous foundries in the schools would be a cheap but profitable method of selling foundry work to students. Classes could be escorted through

¹ Foundry Supt., Caterpillar Tractor Co., Peoria.

the foundry with competent guides who would actually take enough time to thoroughly explain the foundry methods to the students.

Mr. Sandes mentions the low wage scale for apprentices as one of the obstacles which prevents young men from being attracted to the foundry industry. This could be corrected by making the training course so attractive and so well outlined that it would actually be a privilege to get into the course. The apprentice should be made to feel that his graduation rate is his beginning rate and that all of his training can now be brought into practice in order to increase his earning power.

We at Caterpillar are very fortunate in that we have a management who is thoroughly and completely sold on elaborate

training programs. These programs were begun some 24 years ago and have been improved and expanded until at the present time there are somewhere between 500 and 600 young men in training in our various courses.

If such a program had not been in effect we would have been unable to supply the trained personnel for the manufacturing facilities which have increased over five fold in the past 20 years. With the demand for closer tolerances, faster methods, and better quality products becoming a necessity in foundry work, training more competent personnel is not a luxury but an absolute necessity. The extra effort and time spent in a training program now will result in a more secure future for the foundry industry.

### COMMERCIAL EXPERIENCE WITH HIGHER SILICON NODULAR IRONS

By

Richard Schneidewind* and Howard H. Wilder**

There has frequently been some hesitation on the part of foundrymen to making nodular cast iron with over 2 or 2.5 per cent of silicon because of fear that high silicon harmfully reduces the toughness and elongation of the iron. If true, this limitation is serious in view of the quantity of late silicon added as an inoculant and particularly if the nodulizing alloy is of the ferrosilicon-magnesium type.

A typical nodulizing schedule may call for 0.30 per cent of magnesium for nodulizing and 0.50 per cent of late silicon in the form of ferrosilicon for inoculation. If the nodulizing alloy contains 12.5 per cent Mg, 40.5 per cent Si, 15.0 per cent Cu and the remainder Fe, the amount of alloy required is 2.4 lb per 100 lb of metal. This addition will increase the silicon content 0.93 per cent and introduce 0.36 per cent of copper. A late silicon addition of 0.50 per cent will further increase the silicon to 1.43 per cent above that of the base iron,

Similarly if the nodulizing alloy contains 14 per cent Mg, 64 per cent Si, remainder Fe, the addition of this alloy plus the inoculant will raise the silicon 1.87 per cent above that of the base iron. Unless exceptional care is used in selection of the cupola charge it is obvious that the final silicon can be kept below 2.5 per cent only with great difficulty. This is brought home forcibly if it is assumed that the charge will contain at least 40 per cent of nodular sprue if nodular iron is made in continuous production.

White, Rice, and Elsea1 reported a series of irons ranging in silicon from 2.61 to 5.94 per cent. The 1.25-in. keel blocks in the as-cast condition exhibited a great decrease in ductility between 3.22 and 4.95 per cent Si. The tensile strength decreased sharply between 4.95 and 5.94 per cent Si. Since only five compositions were run and since the nodulizing practice was experimental it is felt that these results although correctly indicating the trends, are not sufficiently quantitative to delimit the maximum safe silicon contents for a ductile nodular iron.

#### Annealed Nodular Irons-Ferrite Matrix

In order to study the effect of silicon on the ductile nodular irons, the annealed irons will be considered first. The microstructure consists of only two phases: ferrite and nodular graphite. The mechanical properties therefore can only be those of the ferrite as modified by the presence of the nodular graphite.

Ferrite in annealed gray irons, permanent mold irons, and malleable cast irons is alloyed with appreciable quantities of silicon and may contain inclusions of manganese sulphide, phosphides, and entrapped non-metallics such as oxides and slag. In nodular iron the sulphides and phosphides are held at a minimum. This ferrite is alloyed with silicon and manganese and may carry copper or nickel in addition. The presence of 2.5 to 4.0 per cent of graphite by weight (up to 10 per cent by volume) should tend to reduce the tensile strength and duc-

Yensen² studies the properties of annealed pure carbon-free iron-silicon alloys. Figure 1 shows the yield strength and tensile strength as influenced by silicon. His results showed an anomalous behavior at 2.65 per cent Si which he was not able to explain fully and which is generally not found in commercial materials. The strength values dropped sharply at 4.5 per cent Si. These data indicate that each 1 per cent of silicon increases the base tensile strength of annealed iron 34 per cent and the yield strength, 89.5

Lacy and Gensamer³ reported an equation which shows the increase in strength of ferrite as influenced by alloys as follows:

$$\triangle S = K X^{0.71}$$

where

 $\Delta S = \text{increase in strength in 1000 psi}$ (yield strength or tensile strength) X = amount of alloy in atomic per cent

K = a factor, for Si it is 11.0

Mn it is 7.0 Ni it is 6.1

They further indicated that the solubility limit of silicon in ferrite may be 9 atomic per cent which equals 4.72 per cent by weight. Although 26 atomic

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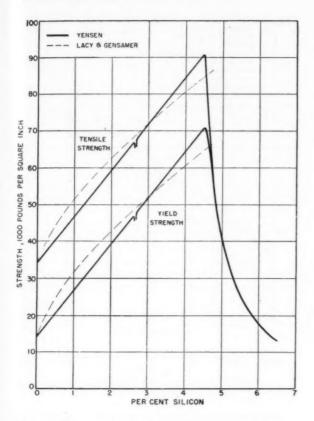


Fig. 1—Effect of silicon on strength of annealed carbonfree iron.

per cent of silicon will form a single phase ferrite, determination of lattice constants indicate that silicon up to 9 atomic per cent decreases the lattice constant at a low rate while silicon above this concentration decreases the lattice at a higher rate. Different properties may be expected in the two ranges. The tensile strength of ferrite with no silicon is taken as 34,000 psi and the yield, 14,000 psi, from Yensen's data. The predicted strengths as computed by Gensamer's equation are as follows:

Computed Strength of Ferrite (Gensamer's Method)

Weight % Si	Atomic % Si	Increase Strength,	in Computed psi Strength, psi	Computed Yield, psi
0	0	0	34,000	14,000
0.5	0.99	11,000	45,006	25,000
1	1.98	17,780	51,780	31,780
1.5	2.95	23,600	57,600	37,600
2	3.91	29,000	63,000	43,000
3	5.80	38,200	72,200	52,200
4	7.67	47,000	81,000	61,000
4.72	9.00	52,376	86,376	66,376

These values are also plotted in Fig. 1 and show reasonable agreement with the actual values obtained by Yensen. The presence of 0.5 per cent Mn in solid solution in ferrite would increase the strength another 3500 psi,

The tensile strength of ferritic nodular iron wheth-

er annealed by a separate heat treatment or selfannealed by a combination of very slow cooling in the mold, low manganese, and high silicon should be the strength of the ferrite matrix, weakened to some extent by the presence of graphite. Grain size would also have some effect.

Figure 2 presents Yensen's data on elongation. As mentioned before, the anomalous behavior at 2.65 per cent Si has not always been experienced with commercial materials.

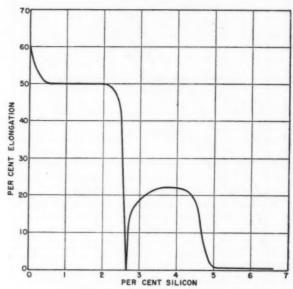


Fig. 2—Effect of silicon on elongation of annealed carbonfree irons (Yensen). Results not found generally in commercial materials.

It may, however, safely be said that the yield and tensile strengths of ferrite increase appreciably as silicon is raised from zero to 4.5 per cent but that the ductility drops in this range. Both properties drop sharply when approximately 4.5 per cent Si is exceeded.

#### Annealed Nodular Irons-Commercial Results

It is to be expected that the strength and ductility, properties of nodular irons will not be so uniform as those of pure iron-silicon alloys which are carbon-free. This is true because the cast irons may contain carbon varying from below 3 per cent to over 4 per cent and may contain traces of Ti, Cr, Mo, and carbides and also appreciable amounts of Mn, Ni, or Cu.

Data were obtained from various sources in this study. Over 200 heats were studied at one foundry where iron was melted in an acid cupola, nodulizing was carried out at 2650 to 2750 F, inoculation was done during reladling. One keel block was tested in the as-cast condition and the other was given a full anneal with the following cycle:

Heated at 1700 to 1750 F for 1.5 hr Furnace cooled to 1500 F Cooled 100 F per hr to 1300 F Held at 1300 F for 1.5 hr Air cooled to room temperature.

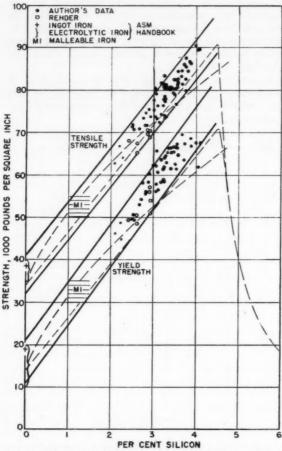


Fig. 3—Effect of silicon on tensile and yield strength of annealed irons.

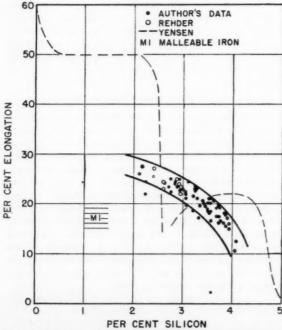


Fig. 4—Effect of silicon on elongation of annealed nodular irons.

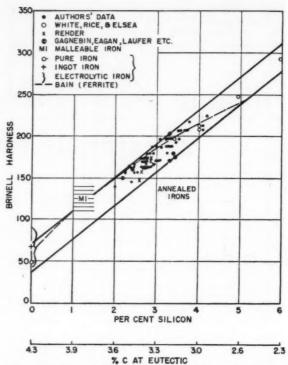


Fig. 5—Effect of silicon on Brinell hardness of annealed irons.

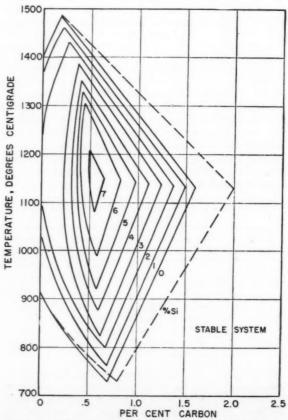


Fig. 6—Influence of silicon on gamma loop in iron-carbonsilicon alloys (Kriz and Poboril).

These data along with a large number of results from other commercial heats and some values from the literature were used in studying the effect of silicon on mechanical properties.

Figure 3 presents the yield strength and tensile strength data obtained. The curves for carbon-free alloys reported by Yensen and as computed from Gensamer are superimposed. The average strengths of malleable iron are also placed in the chart between the silicon limits of 1.0 and 1.5 per cent Si.

The nodular iron values show some scatter since, as pointed out previously, they do contain contaminations of other elements and might show traces of carbides. The freedom from carbides was about as complete as that of commercial malleable irons in most cases.

Nevertheless, the data show satisfactorily that the tensile strength of annealed nodular irons in a range of about 2 per cent to 4 per cent silicon parallels and slightly exceeds the strength of carbon-free ferrite as found by Yensen, despite the fact that the irons contain considerable graphite. The strength of malleable irons falls within the same band as nodular iron, and at zero silicon the strength of ingot iron and electrolytic irons in the annealed state likewise conform.

The yield strength values of ferritic nodular irons also lie in a band which has a slightly steeper slope than Yensen's line for carbon-free irons. The deviation from Gensamer's prediction is still greater at high silicon levels. The values for malleable irons and for electrolytic iron and ingot iron also fall in the band

From these values, since the presence of nodular graphite exerts so small an influence on tensile and yield strengths of commercial irons, it may be inferred that variations in total carbon content of fully annealed, totally nodular irons should have small effect. Of course, segregation of nodular graphite is excluded from this statement.

In Figure 4 the effect of silicon on the elongation is shown. The scatter of points is noted and can be explained by the fact that while small quantities of non-nodular graphite have a minor effect on strength, they do exert a marked influence by their notch effect on the ductility. Also, as is well understood by all malleable foundrymen, small deviations from the ideal heat treatment on the short side will be felt first in the elongation. The figure does show that good nodular irons, properly annealed, can still have elongations in the range of 10 to 16 per cent even if 4 per cent Si is present. Yensen's data are superimposed.

Figure 5 is a plot of Brinell hardness against silicon content. This graph illustrates the hardening effect of silicon on ferrite. Bain's curve showing the effect of silicon on ferrite is superimposed.

#### Heat Treatment of Nodular Irons

It is well known, especially to foundrymen in the malleable industry, that a fully ferritic matrix is most efficiently secured if the second stage anneal involves a slow cooling rate through the critical range (the three-phase region).

Silicon has considerable effect on the iron-carbon

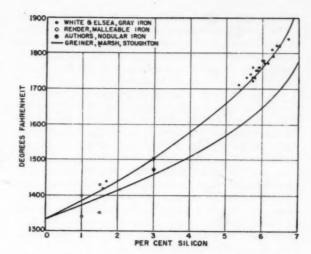


Fig. 7—Effect of silicon on critical range (3-phase region).

phase relationship and therefore on the temperatures for heat treatment.

Figure 6 shows the influence of silicon on the austenite or gamma region. As silicon increases the eutectoid composition decreases from 0.8 per cent C and the eutectoid temperature rises above 1333 F. The curves were plotted from correlated data of Greiner, Marsh, and Stoughton.⁵

Figure 7 presents the influence of silicon on the critical region. In carbon-containing alloys the critical temperature at zero silicon is 1333 F and as the silicon is increased, a three-phase region is formed which at 3 per cent Si lies between about 1500 F and 1470 F. The data of White and Elsea⁶ have been included to cover the range up to over 6 per cent Si.

If a nodular iron in the condition as-cast contains no massive carbides, heat treatment above the critical is not necessary. A treatment between 800 F and 1100 F will provide adequate stress relief without materially decreasing the hardness. At 1200 F or higher the result will be stress relief plus spheroidization of pearlitic carbides and second-stage carbide decomposition. For example, a ½-in. x ½-in. section of nodular iron was 300 Brinell in the as-cast condition. Reheating at 1250 F for 2.5 hr reduced this to 210 Brinell. The same time at 1200 F would produce a hardness of about 240 to 250.

With such heat treatments nodular irons behave like pearlitic malleable irons except that nodular irons, containing higher silicon and closer inter-nodule distances, will anneal faster.

With nodular irons, as may be seen in Fig. 7, higher subcritical heat treatment temperatures may be used with resulting shorter annealing times. Likewise if fully ferritic structures are desired in nodular irons containing massive carbides and first and second-stage annealing are used, the range in temperature where slow cooling is to be employed is appreciably higher than in the case of malleable iron because of the higher silicon.

#### Inoculation

Many foundrymen have felt that when using a nodulizing alloy containing higher silicon, late inoculation with ferrosilicon becomes less necessary than when using a nodulizing alloy with low silicon. This is not always true; although good nodular iron can and is being made without late silicon, inoculation is frequently required in both cases if the best mech-

anical properties are to be produced.

As an example, a ladle of basic-cupola iron was treated at 2850 F with a nodulizing alloy of low silicon content; 0.5 per cent of late silicon was added for inoculation. Fractures of test wedges showed chill only at the tip and a fine, silky surface on the remainder. A second ladle was treated with a nodulizing alloy of equal magnesium content but with a much higher silicon content; no inoculant was used. The total silicon in each case was approximately the same. Yet the test wedge of the second heat showed a coarse dendritic fracture and more chill. The fracture was fine with the second alloy only after about 0.5 per cent late silicon was used.

The effect of inoculation can be further illustrated by the following example: The base iron made in an acid cupola contained 3.56 per cent C, 1.26 per cent Si, 0.47 per cent Mn. After nodulizing, the silicon was 2.47 per cent. Two keel blocks were poured of this iron and successive pairs of blocks were poured after inoculating with 0.2, 0.4, and 0.6 per cent Si

respectively as 90 per cent ferrosilicon.

Typical as-cast properties and annealed properties are given in the following table:

Late Si, %	Y.S., psi	T.S., psi	Elong,	R.A.,		Heat Treatment
0	68900	96800	2.5	2.4	241	
0.2	71400	92800	3.5	1.6	217	As
0.4	63900	93800	3.5	2.8	223	Cast
0.6	64900	95800	5.5	4.7	207	
0	51420	75380	16.5	15.9	163	
0.2	51920	71900	21.5	26.5	163	Annealed
0.4	54420	72400	23.5	24.4	163	
0.6	56900	73400	22.5	23.7	170	

The inoculation has effectively lowered the as-cast hardness and increased the ductility. More surprising is the fact that in the fully annealed condition, the uninoculated iron proved inferior in ductility to the inoculated ones. This behavior was confirmed by many similar tests.

Examination of the table would indicate that for a cooling rate equivalent to that of a keel block, inoculation with increments of silicon over 0.2 per cent Si did not result is so great an improvement as did the first 0.2 per cent. If the iron contains Cr or Mo or other material in appreciable quantity, more inoculation may be desirable than where these are absent.

It has been found that as the section drops, more inoculant is required. A limited amount of data would suggest an inoculating schedule as shown in Fig. 8.

#### Carbon Loss

In nodulizing, particularly with high silicon nodulizing alloys and heavy inoculation, frequently a loss in total carbon is observed. Often this can be explained on the basis of carbon equivalent. For example, á base iron contained 3.56 per cent carbon and

1.26 per cent silicon. The nodulizing alloy and the inoculant together increased the silicon 1.62 per cent to a total of 2.78 per cent silicon. The carbon equi-

valent of this molten iron is  $3.56 + \frac{2.78}{3} = 4.49$ 

per cent.

Upon solidification the analysis showed 3.22 per cent carbon and 2.72 per cent silicon or a carbon equivalent of 4.15 per cent. It is possible that the loss in carbon was due to the supersaturation of the melt so that the carbon either burned out or kished out. The kishing mechanism is observed particularly in heavy sections when the carbon equivalent of the melt is well above 4.5 per cent; in this case nodules migrate to the cope of the casting.

We have observed that reproducible good nodular iron is made when the carbon equivalent of the treated molten iron is 4.5 per cent or under. This is computed by adding the carbon of the base iron to one-third of the silicon of the base iron, of the nodulizing alloy, and of the inoculant. Although good nodular iron has been reported with higher carbon, it is felt that too high a liquid carbon equivalent may contribute to segregation of nodules to the cope and to increase the danger of pinhole porosity.

#### As-Cast Nodular Irons-Properties

Figure 9 presents the tensile-elongation values and Fig. 10 the yield strength-elongation values of a large number of as-cast keel block specimens. It can be seen that they fall in a rather wide band.

A nodular iron casting in the as-cast condition may show inferior properties due to any one or a com-

bination of the following conditions:

- 1. Internal residual stress
- 2. Massive carbides
- 3. Flake graphite
- Defects such as shrinkage cavities, slag inclusions, porosity, etc.

Internal stresses are due to differences in cooling rates and probably exist in all small castings. Massive carbides can be found where the composition of the iron or the quantity of ferrosilicon inoculant was not properly balanced against the cooling rate of the casting. Castings which have very light sections, and which were insufficiently inoculated, and those con-

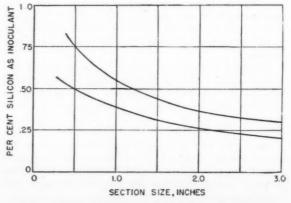


Fig. 8-Recommended range for silicon inoculation.

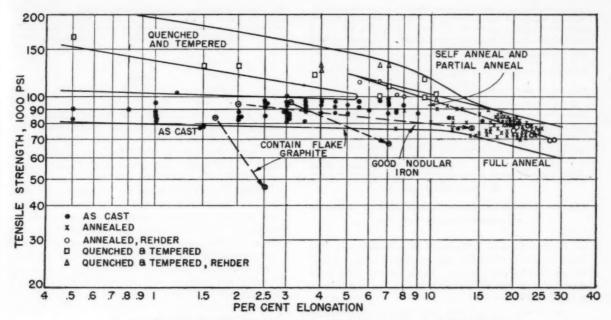


Fig. 9-Tensile strength-elongation relationship of nodular irons, keel block sections.

taining residual alloys such as chromium are liable to have considerable carbide. Flake graphite and pseudo nodules are due to an improper nodulizing operation and shrinkage defects have the same causes as in other castings.

In Fig. 9 three irons are shown. After annealing, one iron developed good properties and the other two lost considerable strength without appreciable gain in ductility. It is obvious that the first iron originally had low properties due to defects 1 and 2 mentioned above. This iron in the as-cast condition was potentially a good nodular iron. The other two irons upon examination showed excessive flake graphite although shrinks, slag, or porosity could have brought about identical behavior.

These three irons were selected for discussion since, as will be shown later, the good iron in the as-cast condition might not have passed specifications whereas one of the poor irons could have passed.

The behavior of as-cast nodular irons free from massive carbides upon annealing seems to follow the same route as similarly treated pearlitic malleable irons. Schneidewind and Reese⁷ showed the gradual increase in elongation and decrease in strength of such irons as annealing progressed. The limiting properties were those of the fully annealed conditions. This is illustrated in Figs. 9 and 10 by values obtained by Rehder⁸ who cast the nodular iron in the white condition and annealed according to a malleable schedule. Also included are values obtained

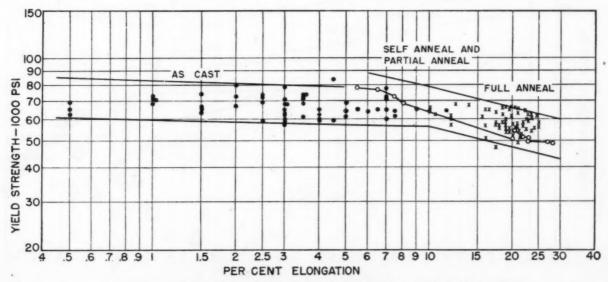


Fig. 10-Yield strength-elongation relationship of nodular irons, keel block sections.

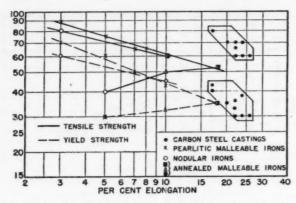


Fig. 11—Strength-elongation relationships of malleable and nodular irons and cast steel.

by Levi⁹ who used self-anneal by covering the keel block molds with sand and permitted slow cooling before shaking the castings. These results along with the results of annealing shown in Fig. 3 and 4 plot in a hand.

Also shown in Fig. 9 are a few points obtained by Rehder¹⁰ and by the authors on quenched and tempered nodular irons. As in the case of quenched and tempered pearlitic malleable irons, the strengths start at higher levels and upon more complete tempering, end in the zone of fully annealed irons.

#### As-Cast Nodular Irons-Specifications

Specifications indicate the minimum properties of the material which the user may expect to purchase in the open market and hence may be a valuable means of comparing one material with another. Although Reese and Gagnebin¹¹ have proposed specifications for nodular irons in the last several years, a new specification has been brought up for official tentative adoption in the ASTM. This reads:

Tensile Strength, psi	Yield Strength, psi	Elongation, %	
80,000	60,000	3	
60,000	45,000	10	

In Fig. 11 are plotted the tensile strength and elongation values of the nodular specification along with those of malleable cast irons, pearlitic malleable irons, and some of the more common carbon cast steels. This figure also presents the same type of comparison using yield strength and elongation.

In Fig. 12 are summarized the properties of nodular irons cast in keel block sections in the conditions ascast, partially annealed, annealed, and quenched and tempered. Also the various specifications¹² for nodular irons, malleable iron, pearlitic malleable irons, and carbon cast-steels are shown. It would seem that if sound nodular irons free from porosity and/or flake graphite are properly heat treated tensile strengths at least 20,000 to 30,000 psi higher than specification can be obtained at any elongation level between 3 per cent and 10 per cent.

The better as-cast, the partially annealed, and the fully annealed irons seem to lie in a band whose equation is of the form

$$E = \left(\frac{a}{Y}\right)^4 \text{ and } \left(\frac{a}{T}\right)^4$$

where

E = % elongation

Y = yield strength in 1000 psi

T = tensile strength in 1000 psi

a = a constant

The constant, a, for good irons lies between 140

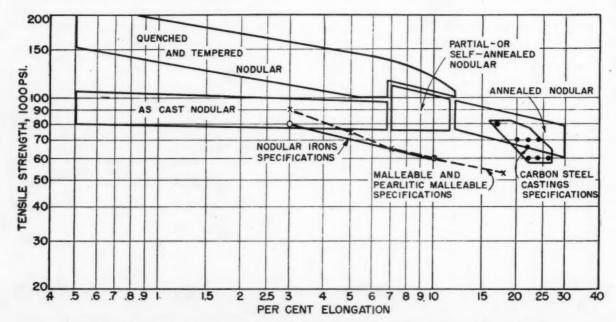


Fig. 12—Comparison of properties of nodular irons (keel block sections) with various specifications.

and 190 for tensile strength and between 100 and 120 for the yield strength. Silicon tends to raise the constant. The tensile constant for quenched and tempered irons may be as high as 210.

The proposed specification lies on lines E =

$$\left(\frac{105}{T}\right)^4$$
 and  $\left(\frac{80}{Y}\right)^4$  respectively

#### Conclusions

1. The yield and tensile strengths and the Brinell hardness of nodular irons in the annealed condition are increased as the silicon content increases.

2. The elongation drops gradually with increasing silicon but at 4 per cent silicon can still be over 10 per cent.

3. In the as-cast condition, the properties of the iron are greatly affected not only by cooling rate but also by the amount of silicon inoculant.

4. Nodular irons can be made with silicon contents up to 4 per cent which will be acceptable with respect to the proposed ASTM specification. It is probable that for special applications nodular irons of over 4 per cent silicon will prove valuable.

#### **Bibliography**

- 1. W. H. White, L. P. Rice, and A. R. Elsea, "Influence of Silicon Content on Mechanical and High-Temperature Properties of Nodular Cast Iron," Transactions, A.F.S., vol. 59, pp. 337-347 (1951).
- 2. T. D. Yensen, University of Illinois Experiment Station Bulletin 83 (1915), 66 pages, also page 85, Alloys of Iron and Silicon by Greiner, Marsh, and Stoughton. McGraw-Hill Book Co. (1933).
- 3. C. E. Lacy and M. Gensamer, "Properties of Alloyed Ferrites," Trans., A.S.M., vol. 32, pp. 85-115 (1944).
- 4. E. C. Bain, Function of the Alloying Elements in Steel, A.S.M., Cleveland, 1939, page 66.
- Greiner, Marsh and Stoughton, Alloys of Iron and Silicon, (See Ref. 2). Correlated data from Kriz and Poboril, Figs. 78, 79, 80, pp. 153 and 154.
  - 6. W. H. White and A. R. Elsea, "Growth and Scaling

- Characteristics of High-Silicon Cast Iron," Transactions, A.F.S., vol. 57, p. 463 (1949).
- 7. R. Schneidewind and D. J. Reese, "Pearlitic Malleable Irons, Plain and Alloyed," Transactions, A.F.S., vol. 56, pp. 410-429 (1948).
- 8. J. E. Rehder, "An Introduction to the Annealing of Nodular Iron," Transactions, A.F.S., vol. 58, pp. 298-309 (1950). 9. W. W. Levi. Private Communication.
- J. E. Rehder, "Effect of Phosphorus Content on Mechanical Properties of a Nodular Cast Iron," Transactions, A.F.S., vol. 59, pp. 501-508 (1951).
- 11. D. J. Reese and A. P. Gagnebin, "The Industrial Status of Ductile Iron" and "Inco Writes Spees. for Ductile Iron," Mechanical Engineering, pp. 101-108, Feb. 1951, Iron Age, p. 89, May 4, 1950.
  - 12. A.S.T.M. Specifications

A- 47-33

A-197-39 for malleable irons

A-220-44T for pearlitic malleable irons

A- 27-39

A- 87-36

A- 95-36

А-215-39 Г

A-216-39R for carbon steels

#### DISCUSSION

Chairman: W. H. WHITE, Jackson Iron & Steel Works, Jackson, Ohio.

Co-Chairman: CARL HARMON, Hanna Furnace Corp., Buffalo. Recorder: C. C. Sigerfoos, Michigan State College, East Landing

H. MORROGH: Are there any results available giving ductility values on the basis of the notched impact method of testing? In our experience many engineers are interested in impact values in addition to the per cent elongation measurement reported in the paper.

Mr. Schneidewind: Impact tests were not run on the metals tested; ductility was measured by per cent elongation.

A. B. EVEREST: ² At 2.5 per cent silicon there is a transition from the ductile to the brittle. Above 2.5 per cent silicon impact values may fall as much as 90 per cent. In Europe we specify silicon at 2.5 per cent maximum (according to conditions) to prevent low impact values.

Mr. Schneidewind: From a practical standpoint it is difficult to keep silicon as low as 2.5 per cent.

British Cast Iron Research Association
The Mond Nickel Co., Ltd., London.

¹ British Cast Iron Research Association, Birmingham, England.

## PNEUMATIC RECLAMATION FOR FOUNDRY SANDS

By

#### Clifford E. Wenninger*

#### ABSTRACT

The last two years has witnessed the laboratory, pilot-plant, and commercial development of a pneumatic method for reclaiming used foundry sands. The key unit in the method is an "air-scrubber" that functions in the dry reclamation system in a manner similar to the functioning of a "wet-scrubber" in a wet reclamation system. This paper presents information on the successive stages in development; the design and function of an "air-scrubber"; the arrangement of component units in a typical installation; the characteristics and economics of a commercial installation which has been in operation for some time.

During the course of an investigation at the University of Kentucky it was demonstrated that successful sand reclamation requires that old clay and carbonaceous bonds be removed from individual grains, and that fragments and particles of such bonds then be removed from a sand mass as a whole. With additional investigation, it was concluded that wet reclamation methods seemed to be superior to existing dry reclamation methods in meeting the foregoing requirements. However, it was noted that most of the dry methods studied involved the use of equipment units that, in reality, had not originally been designed for purposes of reclamation. In the final report for the investigation it was suggested that, possibly, dry reclamation methods could be improved upon if more consideration would be given to designing equipment to specifically meet reclamation requirements.

Subsequently, under the sponsorship of the National Engineering Co., a new project was established to investigate the possible development of improved "dry" reclamation methods and equipment.

#### Laboratory Development

Space will not permit the presentation of a detailed resume of all phases of the laboratory investigation. A great number of electrical and mechanical devices were investigated before consideration was given to the possible use of pneumatic force.

From this preliminary study, it was established

that certain requirements must be met by any pneumatic equipment. One was that such equipment must be capable of subjecting sand grains to repeated passes at a target, or through an area of impact and abrasion. Another was that, if possible, impact and abrasion should be secured with sand grains impinging against other sand grains, not sand grains against a metal surface.

After considerable experimentation a laboratory unit was developed to meet the stated requirements. It operated first with compressed air, then with a blower and was so designed that a small batch of sand could be recirculated continuously with individual grains being repeatedly subjected to impact and abrasion. Tests disclosed the unit to be capable of simultaneously removing old bond from individual sand grains and effecting a separation of such old bond from a sand mass as a whole. This combination of functions in a single unit seemed to be highly desirable.

#### Pilot-Plant Development

The pilot-plant unit was then designed to resemble the laboratory units but on a much larger scale. It was supplied with auxiliary equipment such as a magnetic separator, small crusher, drying unit, and bucket elevators. The pilot-plant unit was designed for batch operation.

A schematic drawing of this initial batch unit is illustrated in Fig. 1. Essentially, it consists of a containing shell, an air nozzle, a bottom-flared center pipe, a conical target, an expansion chamber, and an exhaust duct. The drawing shows the normal height of sand charged into the unit. Its peculiar shape is beter understood when one considers the manner in which it functions.

In operation, the unit is charged with a 125-lb batch of used sand. Air from a positive-pressure type blower is introduced through the high-velocity nozzle.

The elongated portion of the containing shell surrounding the center-pipe (which has been termed the "well") is so designed for a purpose. By concentrating the weight of the charged sand on a small horizontal area, sufficient vertical "sand-pressure" is developed to overcome the static pressure of the air-

^{*} Director, Foundry Research & Development, University of Kentucky and Field Development Engineer, National Engineering Co., Chicago.

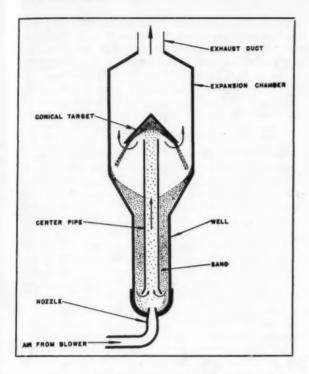


Fig. 1-Drawing of sand scrubbing cell.

stream after it leaves the nozzle and before it enters the center-pipe. This "sand-pressure" confines the air-stream to passing upward through the center-pipe, and prevents "blow-backs" along the exterior of the center-pipe and upward through the "well." The "sand-pressure" also forces sand into the high-velocity air stream, and ensures a consistent loading of the stream at a maximum rate.

As sand is forced into the space between nozzle and center-pipe, it becomes entrained in the air stream, and is hurled upward through the pipe. It emerges at the top of the pipe with considerable velocity and collides with sand trapped in the peak of the conical target.

The conical target possesses the ability to hold an ever-changing, yet more-or-less constant, mass of sand within its peak. The mass is held in position by the upward force of the sand/air mixture emerging from the center-pipe. Sand grains continuously escape from around the circumference of the conical mass while new grains are being added to its center. The net result is that most of the metal target is continuously covered with a layer of sand; and sand grains impinge upon sand grains—not sand grains against metal.

After emerging from the center-pipe and colliding with the sand mass trapped in the target peak, the sand/air mixture is deflected outward and downward by the skirts of the target cone. The sand separates from the air and returns by gravity to the main sand mass lying in the bottom of the containing shell. The air flows outward and upward around the edges of the target cone, and escapes from the shell via the exhaust duct.

The enlarged upper portion of the containing shell serves as an expansion chamber. By being permitted to suddenly expand after emerging from the center-pipe / target-cone area, the high-velocity air stream is converted to slow-moving air currents. Such currents are unable to retain sand grains in suspension. However, small particles of bond and silica that are of a size that can be air-floated do remain in suspension, and are carried out of the containing shell with the escaping exhaust air.

The sand returning to the main mass moves downward with the mass, and soon finds itself being forced to repeat the foregoing cycle. Obviously, the sand mass, and individual sand grains, can be recycled as many times as deemed necessary to obtain a desired degree of cleanliness in the reclaimed product.

The pilot-plant batch unit was used to investigate the time required to reclaim a number of different types of used sands. The effect of continued impact and abrasion upon grain distributions and yields was determined, and various alterations in design were tested to improve efficiency and economy. The sands investigated were obtained from commercial foundries and (in most cases), the reclaimed products were returned to the respective foundries for re-use on production castings. In this manner, practical information and data was obtained to further define the commercial possibilities of pneumatic reclamation.

It was found that most of the used sands could be reclaimed for re-use as molding sands in from 15 to 30 min per batch. However, for re-use as core sands, it sometimes required as high as 45 min per batch to reduce the A.F.S. residuals to a point compatible with rebonding with core oils. It seemed apparent that on a batch basis such extended times would greatly reduce output in tons per hour, and possibly render a pneumatic batch unit uneconomical. Therefore to further perfect the efficiency and economy of pneumatic reclamation it was decided to convert from a batch process to a continuous process.

#### Continuous Development

The construction of a continuous pneumatic reclamation unit can best be understood by referring to Fig. 2, and the photographs in Figs. 4 and 5. The continuous unit consists of two parallel sequences with four batch units combined in each sequence; a total of eight units, or cells, using a common expansion chamber and exhaust system. Through a receiving tank and a manifold system, a single blower supplies air to all cells.

In each 4-cell sequence, continuous flow is achieved by connecting the cells together in step-like fashion. Openings permit sand to flow from feed hopper to cell, cell to cell, and cell to discharge. Sand flowing from feed hopper into No. 1 cell raises its level beyond a set height, and it is forced to overflow into No. 2 cell. No. 2 cell, in turn, is forced to overflow into No. 3; No. 3 into No. 4; No. 4 into the discharge from the unit. Each cell functions exactly like an independent batch cell except that its charge of sand is continuously changing as outgoing grains are replaced by incoming grains.

During operation, in each cell the sand deflected

by the conical target piles up along the exterior walls of the cell, and then flows inward and downward into the "well" surrounding the cell's center-pipe. Flow in this manner creates a steeply sloping upper sand surface in the shape of an inverted cone, or vortex, in each cell. Because of this natural feature, baffles are not needed in the cells to prevent sand grains moving directly from inlet openings to outlet openings. As sand grains enter and fall into the vortex in a cell, they must be cycled one or more times in that cell before target deflection places them in position to escape through the opening to the next cell. While the conical target is deflecting sand from all points on its circumference, and the deflected sand is piling up along all of the exterior walls of the cell, only the sand grains that fall immediately in front of the limited opening escape to the next cell during any one cycle. The sand grains that do not escape must, of necessity, continue to recycle until they are fortunate enough to fall in front of the opening.

Theoretically, it is possible for a sand grain to be cycled only once in each cell (a total of 4 cycles) in its passage from feed hopper to discharge. Actually, and mathematically, this is almost impossible. A

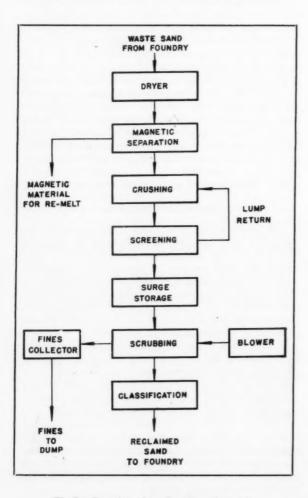


Fig. 2-Drawing of sand reclamation unit.

grain may be cycled only 1 time in one cell; but then be cycled 30 times before it escapes from the next cell. An examination of the individual grains emerging from a continuous unit indicates that they suffer about an equal number of cycles, and, therefore, all grains are cleansed to about the same degree. This uniformity in grain cleanliness is, of course, essential to uniformity in the product as a whole.

#### Control of Quality

In discussing the pilot-plant batch unit tests, it was noted that from 15 to 45 min were required to reclaim different sands to varying quality levels. With the entire charge in a cell being cycled two times per minute, this means that sand grains must be subjected to from 30 to 90 cycles to meet a particular degree of cleanliness.

In a continuous unit, the number of cycles completed by any one sand grain is dependent upon the time it remains in the successive cells. If incoming sand does not tend to push a grain rapidly from cell to cell, it tends to linger in each cell, and complete a greater number of cycles before moving onward. Because of this feature, quality control in a continuous unit is simply a matter of feed control. If sand is fed through the unit at a rapid rate, the number of cycles per grain diminishes. If fed at a slower rate, the number of cycles increases.

Quality can also be controlled by using a constant rate of feed; and changing the number of cells combined in a continuous sequence. With a lesser number of cells functioning in a sequence, individual grains will receive a lesser number of cycles prior to discharge. With a greater number of cells in sequence, they will receive a greater number of cycles in traveling from feed hopper to discharge. Regardless of method used, the average number of cycles per grain (and the quality of the product as a whole) can be controlled and adjusted.

#### A Typical Installation

The operating sequence for a typical commercial installation is diagrammed in Fig. 3. The sand to be reclaimed must be composed of free-flowing, individual grains. Therefore, moisture content must be low (not in excess of 2.0 per cent), and lumps must be crushed or removed by screening.

As indicated by the diagram, some foundries may find it necessary to install a dryer to lower the moisture content of their shake-out sand prior to reclamation. Other foundries with a long sequence of belts and elevators between the shake-out and the reclamation unit may find that agitation and aeration in transmission reduces their sand to a suitable moisture level without benefit of drying. Still others may convey their shake-out sand to a large storage pile—and then feed loose, air-dried sand from the face of such a pile into the reclamation system.

A magnetic separator should be installed to keep tramp metal out of a crusher, and the sand should be screened after crushing to remove and return oversize lumps for additional crushing. If only a small part of the sand in a system is to be reclaimed, a simple screening operation may provide sufficient

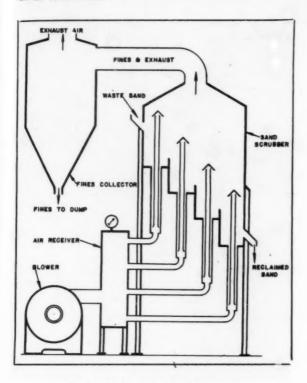


Fig. 3-Sand reclamation flow chart.

sand for reclamation and obliviate the need for the separator and crusher units.

A "surge" bin should be placed immediately ahead of the "air-scrubber" to provide storage space for the sand that has been prepared for reclamation, and to even out the flow from the shake-out. A variable speed belt, or a vibrating feeder plate, can be used to transfer sand from the surge bin to the small inlet hopper of the "air-scrubber."

The products removed from the sand during "scrubbing" are disposed of through dust collectors and an exhaust blower. A cyclone type of collector should be used to trap the heavier fines, followed by either wet or dry dust collectors for the lighter fines.

If it is desired to classify the reclaimed sand to obtain a particular grain distribution, a ladder-type air-classifier can be fitted to the discharge chute. The classifier, of course, can be operated with, and exhausted through, the dust collectors and exhaust blower.

After being reclaimed, the sand can be returned directly for re-use or placed in some storage bin, or silo. The reclaimed product requires neither drying nor cooling, and can be re-used immediately.

#### Cleaning Characteristics of Different Sands

Reclamation by dry impact and abrasion progresses through two stages. In the first, the coatings (or shells) of old bond are broken away from individual grains in a manner somewhat analogous to breaking shells from peanuts. In the second stage, the fragments of the coatings (or shells) are reduced to particle sizes which can be suspended in air, and re-

moved in the exhaust from the reclamation unit.

The physical shape of individual sand grains greatly influences the ease with which they may be separated from their shells of old bond. Round grains and angular grains with their exposed surfaces tend to be freed of their coatings in a comparatively short time. Subangular grains with re-entrant pits and angles tend to shelter the bond adhering in such pits and angles. Consequently, it requires more time, and more impact/abrasion cycles, to clean subangular sands than round or angular sands.

#### Yields with Different Sands

The yield from any particular base sand subjected to pneumatic reclamation reflects (a) the amount of old bond present and removed, (b) the amount of silica fines present and removed, (c) the amount of sand grains fractured and reduced to silica fines and removed. Pneumatic reclamation is most efficient and economical in reclaiming round grain base sands of the Ottawa type. It is least efficient and economical when applied to some of the Lake type sands with a very high percentage of subangular grains in their make-up.

Experience to date indicates the following ranges in yield can be anticipated in pneumatically reclaiming different types of sands.

Predominantly round grains — 85 to 90 per cent Predominantly sub-angular — 80 to 85 per cent Predominantly angular — 80 per cent, or less

However, unforseen benefits sometimes result in the reclamation of poor quality sands. On a pilot-plant test run, one foundry reported their reclaimed sand to be more refractory than their new sand. Eventually, investigation disclosed that the impact and abrasion had removed considerable composite and angular grains from the used sand; and produced a reclaimed sand with a better grain distribution and better flowability for their type of work.

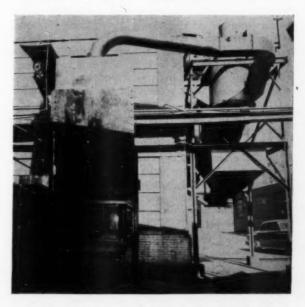


Fig. 4-Side view of commercial installation.

#### Rebonding and Re-use

Pneumatic reclamation does have its limitations. With some of the lower quality base sands, efficiency and economy will not permit residual clay to be removed much below 0.75 to 1.00 per cent. At first glance, such figures seem to bar pneumatically reclaimed sands from being re-used in core mixes. However, in order to get core mixtures that will withstand spalling and erosion, many steel foundries now add from 1 per cent to 3 per cent bentonite to their strong core mixes. Further, since from 15 to 20 per cent new sand must be added daily to replace the loss in reclamation, it is obvious that such new sand should be used in oil sand cores; with the clay-containing reclaimed sand going into stronger core mixes bonded with resins, pitches, etc.

#### A Commercial Installation

As indicated herebefore, the final stages in development required that a full-size installation be operated under the conditions and requirements prevailing within a commercial foundry. For this purpose, a unit was installed in a mid-western steel foundry, and placed in operation in October, 1951. Since the method was considered as still being subject to development, only the minimum essential equipment was installed. As can be observed in Fig. 4 and 5, this consisted of a vibrating conveyor, an operating blower, and air-receiver tank, an air-scrubber unit, a cyclone-type fines collector, a refuse tank for the collector, an exhaust blower, and the piping and wiring necessary for operation. It was understood that auxiliary equipment for better sand handling and dust collection would be added by the foundry after the unit had proven itself in practice.

One of the major reasons for placing the initial installation in this particular foundry was the ease with which the reclamation unit could be attached to the existing sand system. The old used sand from the shake-out passes over belts and a magnetic pulley to an elevator. At the top of the elevator, the sand is dumped into a large rotary screen having ½-in. openings. Sand passing through the openings falls into a large bin over a backing-sand mill. Lumps and core butts retained by the screen are discharged through a chute to a waste sand bin located exterior to the sand-house.

Prior to the installation of the reclamation unit, it was necessary to remove 30 to 50 tons of used sand from the system each day. Another chute in the side of the large bin permitted some of the screened sand to be withdrawn daily to maintain the system at a more-or-less constant tonnage level. The waste sand, plus lumps and core butts, was hauled to a dump located some distance from the plant.

Sand for reclamation was secured by cutting another opening in the large bin, and fitting it with an adjustable slide gate. Sand flows through the gate and onto a plate screen attached to the top of the vibrating conveyor. The screen has %4-in. diameter holes. Rejected material is discharged from the tail of the screen into a small waste pile. The conveyor extends through the side of the building; and all



Fig. 5—Close view of air-scrubber. Note portion of operating blower at rear of air-receiving tank.

sand passing through the screen is carried to the reclamation unit located immediately adjacent to the end-wall of the sand-house.

Under present conditions, the waste sand formerly drained from the bin (and the system) is being fed into the reclamation unit. The former problem of sand disposal has now become one of sand retention. Some sand still escapes from the system in the form of the lumps and core-butts rejected by the large rotary screen. However, the foundry is planning to install a crusher to salvage as much of this sand as possible.

In accordance with foundry operation, the shakeout sand varies from warm to very hot; and may contain up to as much as 3.5 per cent moisture. The agitation and aeration encountered in passing from shake-out to bin reduces the moisture to below 1.7 per cent. Because of this feature, it has not been necessary to add a sand dryer to the system.

The foundry used a washed silica sand for both molding and core mixtures, and a crude sand is used for heavy cores and runner cups. This results in the shake-out sand being a variable blend of the two base sands entering the system. Normally, the shake-out sand consists of about 95 per cent silica sand and about 5 per cent crude sand. Its constitution also changes with the mixing of core and molding sands at the shake-out. Clay residuals may vary from 4.5 to 7.0 per cent, while combustible materials may vary from 0.50 to 0.75 per cent.

In preliminary discussions, the foundry specified

the unit should be capable of reclaiming one ton of sand per hour; with residuals being reduced to about 1.0 per cent clay and about 0.25 per cent combustible material. It was found the unit could meet these specifications; and the reclaimed sand could be used interchangeably with new sand to produce satisfac-

tory castings.

Almost immediately, speculation arose as to the maximum in residual content that would be permissible if reclamation was increased to a rate greater than one ton per hour. On successive days, the feed to the unit was increased by increments of one ton per hour until a reclamation rate of four tons per hour was achieved. With the increases in feed rate, the residual clay contents increased from about 1.0 per cent at one ton per hour to about 2.75 per cent at four tons per hour. Similarly, the combustible material increased from about 0.20 per cent to about 0.45 per cent.

Though the increases in residual contents necessitated some changes in rebonding, it was found that all of the differently reclaimed sands produced castings as satisfactory as those secured from new sand facing mixes. Seemingly, as good casting results could be obtained by reclaiming to a high but consistent residual level as could be obtained by striving for an extremely low residual level. In the light of this development, the reclamation rate was standardized at

three tons per hour.

The base sands entering the system are of the round-grain Ottawa type. Therefore, the yield obtained in reclamation approximates 90 per cent of the sand processed. It is estimated that about eight tons of refuse are removed from the system daily, and this loss is replaced by additions of new sand through the core-room.

Table 1-Grain Distributions and Residual Contents

	Washed New Mold	Crude New Core	Mixed Old Used	Recl. 1-Ton	Recl. 2-Ton	Recl. 3-Ton	Refuse* Fines From
	Sand	Sand	Sand	Rate	Rate	Rate	Cyclone
6	_	_	_	_	_	_	_
12	_	_	0.10	_	0.04	_	_
20	-	_	0.54	0.41	0.12	0.30	-
30	0.07	6.40	0.72	0.60	0.68	0.40	_
40	13.87	39.80	12.12	8.72	9.32	12.65	0.08
50	44.56	25.10	42.30	43.25	44.12	43.24	0.14
70	26.94	11.60	27.44	32.34	30.19	29.12	0.95
100	10.30	7.50	8.10	11.20	11.38	10.02	15.40
140	2.96	5.00	1.34	1.76	1.80	1.44	21.52
200	1.02	0.25	0.42	0.32	0.30	0.20	21.91
270	0.10	0.70	0.30	0.18	0.08	0.06	14.85
PAN	0.03	0.40	0.48	0.10	0.07	0.06	9.71
CLAY	0.10	3.10	5.88	1.09	1.70	2.40	15.11
TOTAL	99.95	99.85	99.74	99.97	99.80	99.89	99.67
COMB.	0.08	0.09	0.69	0.18	0.27	0.37	0.21
* This	analysis	for f	ines is	not ren	resentati	ve of t	he total

* This analysis for fines is not representative of the total material removed from the sand since much of the finer material is not trapped in the cyclone but, instead, escapes in the exhaust from the cyclone.

In Table 1, the grain distributions, clay and combustible contents, have been tabulated for the new base sands, the shake-out sand, the variously reclaimed sands, and the fines from the cyclone refuse tank. It should be noted that the residuals in the sands reclaimed at one, two, and three tons per hour adhere to an obvious mathematical progression. This means, of course, that if residual content would be plotted against tons per hour on a graph, an almost perfect straight line relationship would be made apparent. This seems to confirm the ease with which the quality of pneumatically reclaimed sands can be controlled by simply varying the feed to the reclamation unit. As indicated in previous discussion, the fluctuations in grain analyses are primarily due to the variations in the intermixing of the two base sands at the shake-out, and throughout the system.

TABLE 2-Cost Analysis of Reclamation at 3 Tons Per Hour

Item		Per Hour Operation	
l. Power			
Operating blower,	40 kw		
Exhaust Blower	6 kw		
Vibr. Conveyor,	1 kw		
	_		
	47 kw @ 0.012 per h	r 0.564	0.188
2. LABOR			
*Operator, Full Tin	ne	1.440	0.480
Truck and Driver t	o Remove Fines Refus	e 0.120	0.040
3. MAINTENANCE			
Reserve for Labor	and Parts	0.060	0.020
. GENERAL OVERHEAD			
\$50.00 per wk120	hour week	0.417	0.139
5. DEPRECIATION			
On a 10-year basis		0.390	0.130
# Nation of the Second	TOTALS	*	0.997

* With the installation of a pneumatic elevator, operation will become almost automatic—and a full-time operator will no longer be required. This will reduce cost per hour to below \$2.00; and cost per ton to below \$0.67.

Table 2 presents the foundry's cost analysis of its reclamation operations. At present, it includes a full-time operator to load and transfer tote boxes full of reclaimed sand.

Table 3 presents the foundry's cost analysis for buying, handling, and disposing of each ton of new sand used in production. It shows that it has cost the foundry \$7.50 per ton for each ton of new sand consumed.

TABLE 3-Cost Analysis for Securing and Handling Sand

Item	Cost
Price per ton, f.o.b mine,	\$ 2.55
Freight, mine to plant,	2.86
Unloading and Moving into plant	1.04
Loading and moving to dump	1.05
Total cost per ton	\$ 7.50

#### Summary

A discussion has been presented tracing the laboratory, pilot-plant, and commercial development of a pneumatic method for reclaiming used foundry sands. Information has been included with respect to the operational characteristics and economics of the initial commercial installation in a mid-western steel foundry. It is hoped that this development will be of real value to foundries that have been unable to install reclamation systems because of potential problems involving water supply or sludge disposal.

Acknowledgments

The writer wishes to thank Mr. Walter Horth, of the National Engineering Company, and Mr. S. W. Duncan, of the Duncan Foundry Co., Alton, Illinois, for permission to present the information and data contained herein. Thanks also should be given to Prof. C. S. Crouse, Head, Department of Mining and Metallurgical Engineering, University of Kentucky.

## STEEL DESULPHURIZATION WITH INJECTED CARBIDE

By

S. F. Carter*

#### ABSTRACT

A method is described that has reduced sulphur in small acid electric heats from 0.085 to 0.045 per cent, 0.075 per cent to 0.038 per cent, 0.030 to 0.019 per cent, etc. Sulphur removal in basic heats has been accelerated by the same treatment. The method involves the injection of fine carbide into the furnace bath with inert gas. Results are illustrated by six acid heats and one basic heat with one cost analysis. Experimentation was on production heats.

No completely practical method of lowering the sulphur content in acid-lined steel furnaces has been reported. Sulphur reduction has traditionally required a basic-lined furnace to maintain a slag of sufficient basicity to absorb and retain sulphur from the metal. An optimum ratio of sulphur in the slag to sulphur in the metal requires a highly basic slag very low in iron oxide, which can be accomplished best in a basic electric furnace after a prolonged reducing treatment. The increased heat time due to the slag conditioning plus the higher cost of basic refractories make the removal of sulphur an added expense in steelmaking.

It is not to be expected that acid steel can ever be desulphurized as efficiently as basic steel. However, where acid practice would be preferable from other angles, as in average steel foundry work, and where very low sulphur contents are only occasionally required or when scrap quality only occasionally requires desulphurization, a practical desulphurizing treatment for acid steel should meet a definite need.

Cast iron at higher sulphur levels and at lower temperature has been desulphurized in the ladle with several materials. Steel has been more difficult to desulphurize. Only a few ladle treatments have been successful and then with the recognized disadvantages of ladle treatments. In 1945 the author's plant ran several acid heats on which a second slag of predominantly sodium silicate removed a significant quantity of sulphur in the furnace without severe attack on the acid refractories, but the inclination for the sulphur to revert from slag to metal could

not be consistently prevented by any practical means.

The treatment developed and reported here has desulphurized typical acid electric heats from 0.085 to 0.045 per cent, 0.075 to 0.038 per cent, and from 0.030 to 0.019 per cent sulphur. Desulphurization is accomplished with fine calcium carbide blown under the surface of the metal bath in the furnace with the pressure of inert gas. The steel must be in the proper deoxidized condition and slags high in silica and iron oxide must be removed.

The treatment adds some material costs and some additional operations to acid melting practice, but is not noticeably destructive to refractories and apparently not detrimental to final steel quality. The additional cost appears to be equal to or less than the cost of basic electric double-slag practices accomplishing comparable desulphurization, especially when needed only intermittently.

The treatment was developed in preparation for very low sulphur steels in times when low sulphurscrap might not be available. The increasing emphasis on sub-zero impact properties for armor plate and military equipment suggests a future need for more steels considerably lower in sulphur than generally considered acceptable. Without some desulph-

Fig. 1—Apparatus used to inject carbide into furnace bath.



Note: This paper was presented at the Electric Furnace Steel Conference, AIME, at Pittsburgh, Dec., 1951.

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urizing method acid steel producers will be at a disadvantage in meeting such specifications. The raw material situation and specification requirements have not yet forced us to use this treatment as a regular production method. The process has been used on a relatively few heats on an experimental basis. But it was felt that the results should be reported for the benefit of plants facing raw material and quality problems in their defense efforts.

The ability of calcium carbide to remove sulphur has been academically established for some time, and it has been used with some success in basic furnaces. Some years ago we found that in acid furnaces lump carbide on a bare bath had definite desulphurizing tendencies and did not attack the acid lining, but the reactivity was insufficient for any practical value when applied to the surface of the bath without

agitation.

A device for ladle desulphurization of cast iron with fine carbide was introduced by a carbide manufacturer. Carbide from a pressure-tight hopper was forced by nitrogen gas through a rubber hose and then through a graphite tube immersed into the cast iron in the ladle. This process was found successful on

TABLE 1—HEAT LOGS OF TWO ACID HEATS
DESULPHURIZED AFTER DEOXIDATION

Heat No. 1 (37,425). Argon gas. Carbide injected after deoxidation. [4500-lb heat of 0.40 % carbon steel]

Treatment	Temp.	S, %	C, %
Regular melt and oxidation with 80 lb ore	3100	0.037	0.09
Slag drained. Block 0.60 % Mn and 0.30 % Si			
Injection 47 lb carbide with 21 cu ft argon			
Carbide raked off			
New slag added (limestone, sand,			
manganese ore, iron ore)	2900	0.031	0.29
Addition 0.45 % Mn; tap add			
0.12% Al	3180	0.028	0.38
[20.9 lb CaC ₂ and 9.3 cu ft gas per ton 23.2 lb per ton per 0.0		0.0099	% S =
Time of heat 2 hr 20 min, consuming 1		cold fu	rnace

Heat No. 2 (37421). Nitrogen gas. Carbide injected after deoxidation.

[4500-lb heat of 0.35% carbon steel] Treatment Temp., S,% N,% Regular melt and oxidation with oxygen 3050 0.036 0.11 0.0074 Slag drained. Block 0.60% Mn and 0.30% Si 3120 0.036 0.09 Injection 15 lb carbide with 7 cu ft nitrogen 0.032 0.20 0.0085 Injection 27 lb carbide with 13 cu ft nitrogen 0.028 0.28 0.0136 Carbide raked off New slag added (limestone, sand, iron ore, manganese ore) 2950 0.027 0.28 0.127 Recarb.; temperature raised; add 0.30% Mn; tap 3230 0.026 0.33 0.0139 [18.7 lb CaC2 and 8.9 cu ft gas per ton to remove 0.010% S = 18.7 lb per ton per 0.010% S]

[Time of heat 1 hr 50 min, consuming 1780 kwh (hot furnace)]

cast iron from a desulphurization standpoint. No success was claimed for steel or cast iron of under 0.60 per cent silicon content.

It was apparent that steel would not permit the loss of temperature encountered from such a ladle treatment; that steel would be more difficult to desulphurize than cast iron because of its lower sulphur level and its chemistry; that the oxidizing conditions prevalent throughout most of a steel heat would be unfavorable to the desulphurizing reaction in the furnace; that the graphite tube could not be immersed in steel; and that the nitrogen gas might cause a detrimental increase in the nitrogen content of the steel.

#### **Furnace Practice Alterations**

However, it was felt that the increased agitation and area of contact obtained by inert gas blowing of fine material should provide the increased reactivity needed for carbide to desulphurize steel in the furnace if proper conditions could be obtained without seriously deviating from the established principles of good melting practice.

Using the desulphurization pressure device, an attempt was made to adapt this principle to both basic and acid heats for steel castings. A refractory-protected metal pipe was substituted for the graphite tube. The process was used in the furnace instead of the ladle, making certain alterations of furnace prac-

tice necessary.

Two practices were developed, one in which carbide was injected early in the heat before the oxidizing treatment, and the other toward the end of a heat during the deoxidizing treatment. In both cases slag must be removed before the carbide treatment, and in acid furnaces the carbide slag should be removed after it reaction in order to prevent reversion. On basic heats the carbide can be left on the bath and thinned with fluorspar to serve as the second slag. The carbide size used was 14 N.D.

Argon gas was used first, realizing that its cost was high but that it would be inert to the steel. Less expensive nitrogen gas was used with equal desulphurizing success. However, the nitrogen content of the steel was increased under certain conditions. Carbon dioxide gas was finally substituted, experiencing apparently equal success with a negligible gas cost and no danger of contamination.

#### **Acid Heats**

Heat logs for six acid heats are given in Tables 1 and 2. On heats No. 1 and No. 2 in Table 1 the carbide treatment was applied during the deoxidizing period of the heat.

Heat No. 1 was given a normal oxidation treatment with ore, boiling the carbon down to 0.09 per cent. At 3100 F the slag was drained and a manganese and silicon block added. Carbide was injected through a ½-in. protected pipe, 47 lb with 21 cu ft of argon gas for a 2¼-ton heat. The carbide was then raked off and a final slag added. This synthetic slag consisted of lime, sand, manganese ore, and iron ore so proportioned to give a finishing slag of usual composition and fluidity. This treatment caused a

TABLE 2-HEAT LOGS OF FOUR ACID HEATS DESULPHURIZED BEFORE OXIDATION

[4500-lb heat				SICCII		
Treatment	Temp			er Cent		
1 Italian.	F	S	C	N	Si	Mn
Melt with 10 lb limestor	ne	7				
but no ore Heat to 3100 F. Slag		0.030	0.37	0.0185	0.26	0.52
drained	3100	0.030			0.26	0.48
Injection 88 lb carbide Carbide raked off, new slag added (30 lb iro ore, 15 lb limestone,	2900 n 15			0.0150		0.52
lb Mn ore, 15 lb sand Injection oxygen, 10 min 40 psi, %-in. jet Recarb. Block 0.30% Mn	3100	0.018	0.12	0.0063	0.03	0.13
and 0.20% Si Final add. 0.25% Mn and 0.30% Si; tap 0.12% Al	1	0.019	0.28 Al, %	0.0060 P, %	0.46	0.55
[Heat time 1 hr 26 min,	consu		1160 k	wh (ho	t furr	iace)]
Tensile Test T.S., psi	Y.S.,psi	i E	long.,	% Red.	of Ar	ea, %
0.505 in. 77,100 Heat No. 4 (37815) Nit	41,100 trogen	gas,	31.6		47.5	
0.505 in. 77,100 Heat No. 4 (37815) Nii ox [4500 lb-heat of 0	41,100 trogen	gas, n. carbon	31.6 Carbic	de injec	47.5 ted b	
0.505 in. 77,100 Heat No. 4 (37815) Nii ox [4500 lb-heat of 0	41,100 trogen cidation	gas, n. carbon	31.6 Carbic	de injec	47.5 ted b	
0.505 in. 77,100  Heat No. 4 (37815) Nii ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston no ore	41,100 trogen sidation 0.20% of Temp. F	gas, n. carbon	31.6 Carbic 0.50% Pe	de injec	47.5 tted b el]	Mn
0.505 in. 77,100  Heat No. 4 (37815) Nii ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston	41,100 trogen cidation 0.20% ( Temp. F e, 2850 3130 ith	gas. n. carbon	31.6 Carbid 0.50% Pe C 0.30 0.27	Cr steer Cent N 0.0129	47.5 ted b el] Si 0.13 0.19	Mn 0.31 0.18
0.505 in. 77,100  Heat No. 4 (37815) Nin ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston no ore Heated to 3130 F. Slag drained tinjection 53 lb carbide wi 18 cu ft N (5 min, 3/4-in pipe)  Carbide slag removed, 10 lb ore added. New slag added (10 lb limestom 10 lb Mn ore, 10 lb st (njection oxygen 2½ min,	trogen (idation).20% (Femp. Fe, 2850 ith h)	gas. n. carbon	31.6 Carbid 0.50% Pe C 0.30 0.27	de injec Cr stee er Cent N	47.5 ted b el] Si 0.13 0.19	Mn 0.31 0.18
0.505 in. 77,100  Heat No. 4 (37815) Nit ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston no ore Heated to 3130 F. Slag drained Injection 53 lb carbide wi 18 cu ft N (5 min, 3/4-in pipe) Carbide slag removed, 10 lb ore added. New slag added (10 lb limeston 10 lb Mn ore, 10 lb st Injection oxygen 21/2 min, 40 lb, 3/8-in, jet (265 cu ft)	trogen (idation).20% (Femp. Fe, 2850 ith h)	gas. n. carbon	31.6 Carbic 0.50% Pe C 0.30 0.27	Cr steer Cent N 0.0129	47.5 teed b bel] Si 0.13 0.19 0.21	Mn 0.31 0.18
Heat No. 4 (37815) Nin ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston no ore Heated to 3130 F. Slag drained Injection 53 lb carbide wi 18 cu ft N (5 min, 3/4-in pipe) Carbide slag removed, 10 lb ore added. New slag added (10 lb limeston 10 lb Mn ore, 10 lb scipection oxygen 2½ min, 40 lb, 3/8-in. jet (265 cu ft) Recarb, Block 0.30% Mn and 0.20% Si	41,100 trogen sidation 0.20% of Temp. F e, 2850 3130 ith n. 00 ith n. 3100	gas. n. carbon	31.6 Carbie 0.50% C C 0.30 0.27 0.48	de injec c Cr stee er Cent N 0.0129 0.0117	47.5 teed b bel] Si 0.13 0.19 0.21	Mn 0.31 0.18
0.505 in. 77,100  Heat No. 4 (37815) Nit ox [4500 lb-heat of 0]  Treatment  Melt with 10 lb limeston no ore Heated to 3130 F. Slag drained Injection 53 lb carbide wi 18 cu ft N (5 min, 3/4-in pipe) Carbide slag removed, 10 lb ore added. New slag added (10 lb limeston 10 lb Mn ore, 10 lb s (njection oxygen 2½ min, 40 lb, 3/8-in, jet (265 cu ft) Recarb, Block 0.30% Mn	41,100 trogen sidation 0.20% of Temp. F e, 2850 3130 ith n. 00 ith n. 3100	gas. n. carbon s 0.085	31.6 Carbico 0.50% Pe C 0.30 0.27 0.48	de injec c Cr stee er Cent N 0.0129 0.0117	Si 0.13 0.19 0.21 0.04	Mn 0.31 0.18 0.21

drop in temperature which had to be regained. Final manganese was added and the heat tapped.

The sample caught immediately after the carbide treatment showed a drop in sulphur from 0.037 to 0.031 per cent. However, the more representative final sample ran 0.028 per cent sulphur. (Samples immediately after the treatment were sometimes inconsistent with later samples. The differences may be explained in several ways.) While removing the 0.009 per cent sulphur, carbon content of the bath was in-

Heat No. 5 (57347). Nitrogen gas. Carbide injected before oxidation.

[4500-lb heat of 0.25% carbon steel]

Treatment	Temp	.,	P			
	F	S	C	N	Si	Mn
Melt add 10 lb						
limestone	2800	0.075	0.40	0.0136	0.15	0.40
Heated to 3100. Slag						
drained	3100	0.046	0.36	0.0128	0.17	0.32
Injection 69 lb carbide (5 min 3/1-in. pipe)	2900	0.046	0.54	0.0165	0.26	0.34
New slag mix added () lb limestone, 10 lb san 10 lb Mn ore)						
Injection O2 (6 min, 40	0					
lb, 3/8-in. jet, 640 cu ft)	3030		0.12	0.0095	0.04	0.12
Recarb. All 0.55% Mn a	nd					
0.50% Si added. Tap	3100	0.037	0.24	0.0071	0.43	0.48
[30.7 lb carbide per ton [Time of heat 1 hr	per 0.0	10% S	]			

Heat No. 6 (57437). Carbon dioxide gas. Carbide injected before oxidation.

[4500-lb heat of 0.25% carbon steel]

	Temp.,		Per Cent		
Treatment	F	S	C	Si	Mn
Melt add 10 lb limestone	2850	0.054	0.38	0.27	0.37
Heated to 3100. Slag drained	3100	0.052	0.33	0.39	0.39
Injection 30 lb carbide (3/4-in. p	ipe)	0.043	0.39	0.41	0.30
Carbide raked off, 90 lb ore adde New slag mix added (15 lb limestone, added 15 lb Mn of Injection oxygen (6 min, 40 lb, 3/4-in. jet)	ore)	0.036	0.14	0.05	0.13
Recard. Block 0.30% Mn and					
0.20% Si Final addition 0.25% Mn and		0.038	0.19	0.24	0.26
0.20% Si.					
Tap plus 0.12% Al	3200	0.038	0.24	0.42	0.43
[13.3 lb carbide per ton to remo per 0.01 [Time of heat 2 hr 36 min.	0% S]				

creased 0.20 per cent by reaction with the carbide. This carbon pickup was convenient on this 0.40 per cent carbon steel which required recarburizing, but would limit the amount of desulphurization on lower carbon steels.

Heat No. 2 was almost identical except for the use of nitrogen gas instead of the more expensive argon. Injection of 18.7 lb of carbide per ton with 8.9 cu ft of nitrogen per ton lowered the sulphur in a 21/4-ton heat from 0.036 to 0.026 per cent while raising the carbon from 0.09 to 0.28 per cent. Nitrogen content of the steel was almost doubled, from 0.0074 to 0.0139 per cent nitrogen. This increased nitrogen content would be detrimental in most cases where the decrease in sulphur would be desirable.

#### Melting and Desulphurizing Practice

On the next four heats (Table 2) the entire melting practice and time of desulphurization was changed to overcome the limitations of the practice described. The desulphurization treatment was applied before the working of the heat with no limitations on the

final carbon content and without prolonging the deoxidizing or blocked period of the heat over usual

acid practice.

On Heat No. 3 nitrogen gas was used on a 0.25 per cent carbon steel with a 3/4-in. pipe for injection. The charge was melted with no oxidation treatment and heated to 3100 F. The slag was drained, and 39.1 lb of carbide per ton injected in 5 min, decreasing the sulphur from 0.030 to 0.014 per cent. The carbon was increased from 0.41 to 0.68 per cent, and the nitrogen from 0.0085 to 0.0150 per cent. The temperature was dropped to 2900 F.

#### Use of Nitrogen

The carbide slag was raked off and a synthetic slag mixture of iron ore, limestone, manganese ore and sand was added. Oxygen was injected through a 3/8-in. water-cooled jet for 10 min at 40 psi, producing a violent boil which dropped the carbon from 0.68 to 0.12 per cent, and raised the temperature back to 3100 F. The boil of this violence lowered the nitrogen content from 0.0150 per cent to a normal concentration of 0.0060 per cent. Such elimination of nitrogen seems to make its use more nearly practical when used before the boil. From this point on the heat was finished by the usual practice and behaved in a normal way.

After recarburizing, the heat was blocked with 0.30 per cent manganese and 0.20 per cent silicon additions 8 min before tap. Final additions of manganese and silicon were added 2 min before tapping at 3200 F with 0.12 per cent aluminum added in the tapping ladle. A tensile test from this heat had a good combination of strength and ductility as indicated by the 77,100 psi strength and 47.5 per cent reduction of

area.

On this heat slags were drained and carbide injected very efficiently. Heat time was only 1 hr 26 min on a hot furnace. The desulphurizing treatment prolonged this heat only 10 to 15 min. Power consumption was only 1160 kwh for a  $2\frac{1}{4}$ -ton heat, which is no greater than the average power consumption.

Heat No. 4 followed a similar practice but on a higher sulphur charge caused intentionally by using a large proportion of screw stock that would ordinarily be unfit for use in an acid furnace. Sulphur was lowered from 0.085 to 0.045 per cent with 23.5 lb of carbide per ton and 8 cu ft of nitrogen per ton injected in 5 min through a 3/4-in. pipe. The best desulphurizing efficiency was experienced on this heat, 5.9 lb per ton per 0.010 per cent sulphur removed. Nitrogen was not eliminated as effectively by the boil

as on Heat No. 3. The boil was not quite as violent but the 0.50 per cent chromium seems another likely reason for the decreased nitrogen removal.

On heat No. 5 sulphur was reduced from 0.075 to 0.037 per cent with 30.7 lb of carbide per ton and the same practice. The melt slag was drained and the carbide injected and raked off. A final slag mixture high in iron ore was added and oxygen was injected. This heat was not blocked; all manganese and silicon were added 2 min before tapping at 3100 F. The boil lowered the nitrogen content from 0.0165 to 0.0071 per cent.

The time of this 21/4-ton heat was 1 hr 53 min and the power consumption 1350 kwh. The desulphurization treatment and slag removals apparently prolonged this heat about 30 min and increased the power consumption 50 to 100 kwh. Some of this increase was due to additional samples for research data and unfamiliarity with the practice. The time should

be improved with experience.

On Heat No. 6 carbon dioxide was used as the inert gas. Sulphur was reduced from 0.054 to 0.038 per cent by 13.3 lb of carbide per ton. Desulphurizing efficiency seemed essentially comparable to that of nitrogen gas. Because of a cold furnace the heat time and power were greater than that usual on a hot furnace, but the increase was about the same as on the other heats.

#### Desulphurization Efficiency

All six acid heats are summarized in Table 3. From the last column, indicating the amount of carbide used per ton per 0.010 per cent sulphur removed, it is apparent, as would be expected, that desulphurization efficiency is much better the higher the original sulphur. Less treatment was required to remove 0.040 per cent sulphur, from 0.085 to 0.045 per cent, in Heat No. 4 than was required to remove 0.011 per cent sulphur from 0.030 down to 0.019 per cent in Heat No. 3.

After allowance for differences in respective sulphur levels it appears that efficiency was at least equal and possibly better on the heats desulphurized before the oxidation treatment. This makes it possible for the heats to be worked in their usual manner without the danger of an over-reduced finish or remaining in a deoxidized condition longer than has been found desirable for best ductility² and minimum hydrogen absorption.³

A cost study of Heat No. 5 is shown in Table 4. This heat experienced the second best efficiency. The

TABLE 3-SUMMARY OF ACID HEATS USING CARBIDE DESULPHURIZATION

Heat	Туре	Inert	When	Sulph	Sulphur, % Sulphur A	Sulphur	Sulphur Amount/Ton		CaC ₂ /Ton /0.010%S, lb
No.	Steel	Gas	Desulphurized	Orig.	Final	Removed,	CaC ₃ , lb	Gas, cu ft	
1	1040	Argon	After deox.	0.037	0.028	0.009	20.9	9.3	23.2
2	1035	Nitrogen	After deox.	0.036	0.026	0.010	18.7	8.9	18.7
3	1025	Nitrogen	Before ox.	0.030	0.019	0.011	39.1		35.5
4	1020	Nitrogen	Before ox.	0.085	0.045	0.040	23.5	8.0	5.9
	(0.50 Cr)	0							
5	1025	Nitrogen	Before ox.	0.075	0.037	0.038	30.7	8.5	8.1
6	1025	CO ₂	Before ox.	0.054	0.038	0.016	13.3		8.3

Table 4—Cost Analysis of Desulphurization Treatment Acid Heat No. 5. Sulphur from 0.075 to 0.037 Per Cent

Material	Amt./Ton	Unit Cost	Cost/Ton
Carbide 14 ND	30.7 lb	0.07 per lb	\$2.15
Nitrogen (or CO ₂ )	8.5 cu ft	0.02 per cut ft	0.17
Cost of pipe and	protection (ap	oprox.)	0.50
Additional power i			0.40
Total cost of ti	reatment		\$3.22

total material cost of reducing the sulphur from 0.075 to 0.037 per cent was \$3.22 per ton. Allowance was made for the increased power required, but no attempt was made to evaluate the increased time of the heat. Even after more experience it seems likely that the slag removals and carbide injection will undoubtedly add 10 to 20 min to the length of an acid heat. However, costs of this order are expected when desulphurization of this degree is accomplished.

Since this treatment accomplishes the equivalent of basic practice, costs must be compared with basic practice rather than ordinary acid practice. In the author's plant a conventional reducing basic slag requires a greater power consumption and a greater increase in heat time in addition to the extra cost of basic refractories. Purchase of selected low-sulphur scrap generally requires a premium price in excess of the cost of this carbide treatment.

#### **Basic Heats**

Conventional basic electric practice is certainly adequate for sulphur removal. However, the carbide injection method, by accelerating the reaction rate, offers some possible advantages of shortened heat time. Carbide injection was actually tried first on basic heats to establish the practicality of the reaction before attempting its application on acid heats.

Table 5 shows a heat log of one such basic heat with the most complete sample data. Following the usual oxidation and draining of the oxidizing slag the heat was deoxidized with Mn and Si. Instead of adding a second slag and waiting for it to become reducing and absorb sulphur, the carbide was injected and left on the bath as the second slag after

TABLE 5-HEAT LOG, BASIC HEAT NO. 56846.

Nitrogen gas. Carbide injected after deoxidation. 3400-lb heat of 0.35% C steel

Treatment	Temp.,	S, %	C, %
Melted with 80 lb limestone, 50 lb ore			
Oxidizing slag drained clean			
Deoxidized with 0.40% Mn and 0.20% Si	3070	0.045	
Carbide injection 67 lb with 40 cu ft N:			0.15
After first 1/4 of carbide	3050	0.037	0.20
After second 1/4 of carbide	3000	0.033	0.24
After third 1/4 of carbide		0.029	0.30
After fourth 1/4 of carbide	2950	0.026	0.36
Heated 15 min. Final add. Mn and Si; tap	3200	0.022	0.36
38.3 lb CaC2 and 22.8 cu ft gas per ton to		0.023	% S :
16.6 lb per ton per 0.0109	S		

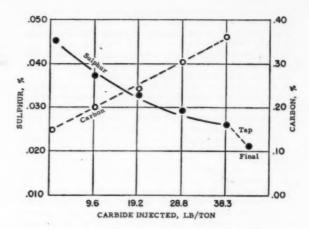


Fig. 2—Graph showing relationship of sulphur elimination and carbon pickup (basic heat No. 56846).

thinning to desired consistency by adding fluorspar. Sulphur was reduced almost immediately from 0.045 to 0.022 per cent with 38.3 lb of carbide and 22.8 cu ft of gas per ton of steel. The efficiency of this heat was 16.6 lb per ton per 0.010 per cent sulphur, which is somewhat better than the acid heats in this sulphur range. Nitrogen gas was used but, on the basis of later experience, carbon dioxide would be preferred.

Samples were taken after each one fourth of the carbide injection. Sulphur and carbon trends are indicated graphically in Fig. 2. Better desulphurization efficiency and lower carbon pickup were experienced on the first half of the treatment which did 63 per cent of the desulphurization. As the sulphur reached lower levels (and the temperature dropped) carbon pickup continued but desulphurization efficiency decreased. Carbon pickup on the first half was 0.06 per cent per 0.010 per cent sulphur removed; and on the last half 0.14 per cent carbon per 0.010 per cent sulphur.

In basic melting this carbide injection would be faster and more direct than conventional reducing practice. No additional steps are required. The second slag is simply injected into the bath in its already reduced and carbidic condition in contrast to usual practice in which the slag components are shoveled in and require a period of time to reach the proper condition to absorb sulphur. Basic heat time can be shortened considerably. Another advantage difficult to evaluate is the shortened period in an inactive, reducing condition. This should reduce the possibility of hydrogen absorption. Actually the inert gas used to inject the carbide should have some tendency to remove or keep out hydrogen. Without production experience these possibilities can only be suggested.

#### References

- 1. C. K. Donoho, "Shop Report on Ladle Treatment With Sodium Metasilicate."
- 2. S. F. Carter, "The Effect of Melting Variables on Tensile Properties of Acid Electric Steel," *Proceedings*, Electric Furnace Steel Conference, AIME (1947).
- 3. S. F. Carter, "The Effect of Melting Practice on Hydrogen Content," Journal of Metals, Dec. 1949.

Aver. lb Per Ton Melt Aver. Sulphur

#### DISCUSSION

Chairman: C. H. Lorig, Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: Снаs. Locke, Atlas Foundry & Machine Co., Tacoma, Wash.

Recorder: E. C. TROY, Consultant, Palmyra, N. J.

A. L. DE SY (Written Discussion): In his interesting study of desulphurization of molten steel with calcium carbide the author used nitrogen gas as a carrier for the carbide powder. At the end of his paper he states that, on the basis of later experience, carbon dioxide would be preferred instead of nitrogen.

This statement appears rather strange to us as carbon dioxide is oxidizing and desulphurization implicates strongly reducing

conditions.

Therefore the author's comments on this point would be

highly appreciated.

MR. CARTER (Reply to Prof. De Sy): We did not find the oxidizing power of carbon dioxide strong enough to reduce the desulphurizing effectiveness to any significant extent. Before trying CO₂ we questioned its performance for the same reason. Apparently the carbide is so strongly reducing that it overcomes the mildly oxidizing tendency of CO₂. Since writing the paper we have used CO₂ on several more heats with consistent success.

H. H. Blosjo (Written Discussion): ² The author has established a valuable procedure for reducing sulphur in the acid process. The acid process of melting has many advantages over the basic process, particularly in production of steel for castings. With the ever increasing sulphur content of the available melting scrap, its removal is becoming a problem. The method used by the author before oxidation is particularly interesting as the carbon pick up from the carbide can be used to boil the heat. This boil also removes any gas that might have been introduced into the melt by the carbide injection.

We thought it would be interesting to report briefly on some experimental work we have done in our plant using calcium carbide. Our work has all been done by adding the carbide to the ladle. This limits the carbide that can be added due to the carbon pickup, particularly in the low carbon steels. No gases were used in adding the carbide. The method used was to finish the heat (acid lining) as is normally done and tap it into an acid-lined teapot ladle. The carbon in the heat as tapped was held 0.01 per cent below that desired for each pound of calcium carbide added per ton of melt. The heat size varied from 4000 to 4700 lb. The heats were then transferred from the first ladle to a second ladle by pouring down through the teapot spout of the second acid-lined ladle. A container having an orifice of such size that it discharged the required carbide in from 50 to 60 sec was mounted over the spout of the second ladle. Shortly after the transfer is started, the carbide is started flowing and is carried down through the spout by the stream of metal into the second ladle. The carbide continues to flow for a period long enough to transfer from one-half to two-thirds of the heat. In the second ladle, 40 lb of pebbled lime is added and preheated with the ladle to form a basic slag cover for the desulphurized metal.

Fifteen heats were transferred this way. Five of these heats had only carbide added to them. Five heats had about 2 lb of fused soda ash added per ton of melt in addition to the carbide. The soda ash was dried and added to the lime in the ladle just prior to the transfer. Five heats had about 2 lb of Mischmetal per ton of melt added in addition to the carbide. The Mischmetal was wired to a rod and plunged to the bottom of the ladle after the transfer was complete.

Table A gives the average data for the five heats produced by each method. It will be noted that it took 12.4 lb of calcium carbide per ton of melt to reduce the sulphur 0.010 per cent starting at a sulphur level of 0.038 per cent when only carbide was used with the lime. The author's data indicates they used about 20 lb per ton starting at the same sulphur level.

Adding soda ash to lime in the ladle reduced the amount of carbide necessary to drop the sulphur 0.010 per cent from the 0.038 per cent level to 9.8 lb. Fused soda ash is a cheap material and adds to the desulphurization at a low cost. We were unable to make use of soda ash safely, however, as we had two explosions in the slag of the second ladle. In the last heat that we ran with soda ash, an explosion took place in the slag fully 5

TABLE A

							Aver. ID CaC
Heats	Lime	CaC ₂		Misch- metal		Drop	per ton melt per 0.010 S drop
A*	20	7.2			.0387	.0058	12.4
B**	20	6.4	2.2		.0381	.0067	9.8
C***	19	7.0		2.1	.0363	.0076	9.2
A*	Aver. of	five he	eats us	ing lime	e and ca	lcium	carbide.
B**	Aver. of	five h	eats u	sing lin	ne, calciu	ım car	bide and purite.
C***	Aver. of	five h	eats u	sing lin	ne. calcin	ım car	bide and misch-

min after the transfer was complete, and the ladle was down on the pouring floor. Fortunately only two men were burned and only slightly, with no lost time, but enough metal and slag was thrown around to have resulted in serious burns. The use of soda ash was discontinued, therefore.

Adding Mischmetal to the transferred heat reduced the amount of carbide necessary to drop the sulphur 0.010 per cent at the 0.036 level to 9.2 lb. The Mischmetal also reacts with the remaining sulphur to form Type I inclusions resulting in ex-

cellent physical properties.

Table II shows the average chemical and physical properties of the five heats produced using soda ash and also the average of the five heats using Mischmetal with the carbide addition. Unfortunately the heats using carbide only were not checked for their physical properties. The physical properties of soda ash-calcium carbide heats were very good, but those of the Mischmetal-calcium carbide are most outstanding. The Mischmetal heats also were more consistent in their properties.

TABLE B—AVERAGE PHYSICAL PROPERTIES OF HEATS NORMALIZED AND TEMPERED AT 1250 F.

Heat	s C	Mn	Si	S	T. S., psi	Y. P., psi	% El.	% R.A.	P Value
В	.25	.80	.44	.032	75,400	48,100	32.7	55.6	81.8
C	.25	.81	.45	.029	78,300	50,000	32.4	58.1	85.4

MR. CARTER (Reply to Mr. Blosjo): The desulphurizing results from carbide added in a ladle spout are very interesting and represent another method of obtaining more contact with carbide and better kinetic conditions than simply adding to the slag. The reladling process would place some limitations on this method in our plant because of the temperature loss. We appreciate the results of this related work.

C. E. SIMS (Written Discussion): ² This is a good example of a basic practice, superimposed briefly on an acid practice. Although it may be done in an acid-lined furnace, there is no doubt but that sulphur reversion would take place if the carbide slag was not raked off promptly after injection.

The classic desulphurization reaction shows S or FeS reacting with CaO and C to form CaS and CO. This is undoubtedly correct and indicates that the ideal conditions for desulphurization exist with a highly basic slag and strong reducing conditions. Recent work^{1,2} has indicated that the reducing conditions are even more important than the basicity.

While this is well established, it is not always recognized that fairly efficient desulphurization can sometimes be attained under conditions less than ideal, whereas, the apparently ideal conditions often fail to give good sulphur elimination. For example, in basic practice, appreciable quantities of sulphur are removed into the oxidizing slag during the boil, while the white reducing or carbide slag usually falls far short of its full desulphurizing potentialities.² The reason for this is in the kinetics of the reaction under the two conditions. The boil produces turbulence that is favorable to rapid reaction, while the static conditions of the reducing slag are definitely unfavorable.

The injection method can attribute much of its success to good kinetic conditions that are inherent to the process. I have used gas injection without calcium carbide with good results to speed up the desulphurization reaction between a basic reducing slag and a steel bath. Here too, CO₂ gas worked just as effectively

¹ Laboratorium Voor Metallurgie, Gent, Belgium.

² Minneapolis Electric Steel Castings Co., Minneapolis.
³ Battelle Memorial Institute, Columbus, Ohio.

as  $N_2$ . It is easy to see that the favorable kinetics of the carbide injection process could be used to advantage even on basic heats. There is some doubt as to how effective it would be on a very large furnace, of, say, 100 tons, however.

Nitrogen is objectionable from the standpoint of dissolving in the steel bath even though most of it can be removed by a subsequent vigorous boil. Carbon dioxide theoretically would be unfavorable because it is an oxidizing gas. As shown, however, its oxidizing power is negligible in the presence of the carbon contents of the steel bath and the calcium carbide.

Calcium carbide is an excellent carburizer for steel and as such has some objectionable features. Considering the kinetics of the process and the fact that highly reducing conditions are not necessary, it is surmized that crushed burned lime alone or a mixture of burned lime and calcium carbide would be just as effective and introduce less carbon. Has the author considered this?

#### REFERENCES

 N. J. Grant and J. Chipman, "Sulphur Equilibria Between Liquid Iron and Slags." AIME (April 1946—T.P. 1988, Trans. AIME (1946) 167, 134.

2. S. F. Urban and G. Derge, "Oxygen in Basic Electric-Furnace Baths," Electric Furnace Steel Proc. (1946), Vol. 4, pp.

31-45.

MR. CARTER (Reply to Mr. Sims): The explanations by Mr. Sims are appreciated. A mixture of lime with carbide has not been tried, but has been considered. For our occasional applications we have needed the maximum desulphurizing efficiency with no serious limitations from carbon pickup. For that reason we have not tried less active materials.

It appears quite likely that other materials and combinations of basic fluxes injected in this manner might have certain advantages. If this injection process should become a routine procedure, I feel sure improvements in cost and effectiveness could be effected by experimenting with some of these combinations.

E. A. Loria (Written Discussion): ⁴ The author is to be complimented for providing a real service to the acid electric steel producers. He has devised a practical method which should certainly aid them in their efforts to reduce sulphur. Blowing fine calcium carbide under the bare bath by means of a ¾-in. diam pipe and moving it through the bath produces little shutes of flame on the surface and a powdery slag that is easily raked off. This "falling" slag which disintegrates to a fine powder on cooling is the normal type of reducing slag which contains large amounts of di-calcium silicate. Even gas flow during the operation is necessary in order to prevent clogging of the carbide.

There are several factors in the process which should be noted. It is done at a time when the metal has reached a high temperature so that the solubility of the carbide is enhanced. It is done on a bare bath, before or after deoxidation, because if the usual slag were left on it would lower the efficiency of the carbide and might even cause a sulphur reversion under certain circumstances. Finally, the ability of the process to lower sulphur is dependent on the initial sulphur content of the charge and the amount of reduction desired. For example, a reduction of from 0.085 pct to 0.045 pct is more easily accomplished than a

reduction of from 0.030 pct to 0.019 pct.

Apparently the author has not made enough heats yet to say at what carbon level the injection should be made for best efficiency and what carbon pick-up is obtained from the carbide when it is used in varied amounts to remove different amounts of sulphur, though the carbon pick-up can be readily consumed by the later injection of oxygen? No doubt the desulphurization ratio is not constant for changes in the analysis of the metal. In other words, the activity of the iron sulphide, its tendency to leave the metal as iron sulphide or calcium sulphide, varies with metal analysis. For example, when carbon and silicon are high, as in cast iron, iron sulphide has a greater tendency to leave than in low carbon, low silicon steel. Perhaps the way a particu-

lar heat of steel is made, i.e., the oxidation of the carbon and silicon and conversely the amount of iron oxide dissolved therein, affects the results obtained with carbide or burnt lime injection. This would mean that better sulphur removal would be achieved by the use of carbide injection before oxidation when the carbon and silicon contents of the heat are much higher than they are after melt and oxidation with oxygen.

Three gases have been employed as pressure agents. Argon has been deemed too expensive, nitrogen as undesirable because of metal absorption, and carbon dioxide apparently favored because it is the cheapest of the three. However, Tables 1 and 2 give data for four heats employing nitrogen and only one heat employing carbon dioxide. Have anymore heats been made in a hot furnace utilizing carbon dioxide? Maybe the carbon dioxide is not detrimental in raising the oxygen level of the bath during the process, providing but a very small increase which does not throw the heat out of line. Nevertheless, perhaps the efficiency of the carbide may be damaged by the presence of an oxidizing agent at the point of contact with the steel in much the same manner as, for example, would occur if raw lime were used with its oxygen carriers, calcium carbonate or calcium hydroxide. Has the author tried freshly burnt lime in combination with calcium carbide in view of the fact that the desulphurization ratio increases with increasing amounts of excess (or free) lime? Experiments are underway employing just pulverized burnt lime which is discharged into the bath under nitrogen gas pressure by means of an ordinary furnace patching gun. Does the author have any figures on phosphorus removal for the four heats where the carbide was injected before oxidation?

The statement is made that carbide injection after deoxidation prolonged that stage of the heat. The times required for the treatment of the two heats by this practice are not given in Table 1, but should be of the same order as the time interval of 5 to 10 min given in Table 2 for the treatment of four heats before deoxidation. Does the author believe this time interval is too excessive even though the injection gas pressure should have some tendency to remove or keep out hydrogen? Also, is it not possible that, in cases where a marked decrease in sulphur is desired, it would be worthwhile to do it in the deoxidation stage when the oxygen content of the bath is lower and hence does not hinder sulphur removal? Perhaps a quick fluxing coverslag (burnt lime, cryolite, alumina mixtures) would protect the bath from oxidation. These oxides are very stable so they are not easily reduced by calcium carbide. Of course, with this practice, an inert gas would have to be used so that oxygen would not be introduced by the carrier. No doubt, silica has an important role in desulphurizing. Silica in the slag can add silicon and oxygen to the metal and the latter slows up sulphur removal even though present in a very small quantity.

MR. CARTER (Reply to Mr. Loria): We appreciate the theoretical elaboration by Mr. Loria but cannot answer many of his

questions.

We agree that carbon content effects sulphur activity and that higher carbon steels should be desulphurized more efficiently but we have no data to support any quantitative comparison.

As previously indicated several more heats have been successfully treated with CO₂ as the carrier gas than are indicated in

the paper.

We do not have sufficiently accurate data to prove any significant phosphorus elimination. Some of the samples might have shown a slight trend but the difference was almost within analytical precision and normal variations over a bath.

MR. CARTER (Author's Closure): In closing I would like to repeat that this treatment as described here is not intended as a completely perfected process but rather as an opening up of a field of desulphurization. Gas injection offers the advantages of good kinetic conditions and accelerated reactivity. The equipment used proved relatively simple and effective. Carbide desulphurizes effectively without interference from acid linings.

Numerous modifications and combinations may be applied profitably to meet various conditions.

⁴ The Carborundum Co., Niagara Falls, N. Y.

## PRODUCTION EXPERIENCES WITH A BASIC CUPOLA

## By Sam F. Carter*

The basic cupola has been the object of increasing consideration during the past few years. A previous paper¹ described the first three years of experimentation and production trials in the author's plant. On a small cupola the magnesite brick lining was actually more resistant than any clay refractories, but refractory costs were very high. Experimental heats melting a wide range of metal charges with two different cokes and a variety of flux combinations accomplished attractive desulphurization within the basic cupola and produced carbon pickups beyond the limitations of acid cupola melting.

These chemical accomplishments seemed to elevate the cupola from a melting furnace to a refining furnace, and to make it a more suitable unit for melting the newly developed nodular or ductile iron. On the basis of the chemical results obtained, several different applications were suggested in which basic melting might be justified by improved iron quality or the use of larger proportions of scrap and poor

grade raw materials.

The plant has continued to operate one production cupola with a basic lining. Continual progress is being made in understanding basic operation and reducing the cost of refractories. Basic cupolas installed by a number of American plants for several different objectives, together with experiences in Europe, have collectively established more clearly the application of basic melting in the foundry field.

#### Refractory Lining and Patching

The cupola operated at the plant has a 36-in. inside diameter with a 9-in. lining, and melts at a rate of 4 tons/hr for 5- to 8-hr heats. A description of the cupola is given in Table 1. Properties of the coke are given in Table 2. The iron melted is predominantly a medium phosphorus gray iron for air test pipe fittings and machine tool castings. Such a 0.20 per cent phosphorus iron is special in this plant where higher phosphorus contents are predominant. In addition to this iron, most of the occasional special irons are melted in this cupola.

The original basic lining consisted of dead-burned

TABLE 1—DESCRIPTION OF CUPOLA

Inside Diameter, in	36
Lining Thickness, in.	9
Number of Tuyeres	
Total Tuyere Area, sq. in	336
Distance Tuyeres to Bottom, in	22
Distance Tuyeres to Charge Door, ft	13
Weight of Metal Charge, lb	1000
Volume of Air, cfm	2500
Average Pressure, oz	10

magnesite brick. In the basin  $4\frac{1}{2}$ -in. arch brick were backed by  $2\frac{1}{2}$ -in. straight brick against the shell. In the melting zone 9-in. keys were used beyond the area of burn-back. For a completely basic cupola the stack above the melting zone was lined with two layers of  $4\frac{1}{2}$ -in. arch brick, the back course clay brick and the inside course magnesite brick. Currently clay brick are used altogether in this portion of the stack from 60 in. above the tuyeres up to the cast iron blocks used for the final 2 ft below the charging door.

Daily patching of the burn-back has been the object of continual experimentation in order to reduce the cost of basic refractory maintenance. A number of materials have been tried, from which five essentially different patching procedures have been satisfactorily used and their relative performance and cost compared.

The acid cupola, with the original practice of patch-

TABLE 2—PROPERTIES OF COKE

(Range of properties of foundry coke used on all heats in paper unless exception indicated.)

	_	_	_	_	_	_	_	_	_	_	 	
Sulphur, % .											 	0.65-0.85
Volatile, %											 	0.60 - 1.00
Ash, %											 	9.0 - 10.5
True Specific	Gra	vity									 	1.85 - 1.95
Apparent Spec	ific	Gra	vit	y							 	0.95 - 1.05
Porosity, %											 ******	48 _51
Shatter Test, 2	in.	plu	S.								 	81 -86

	Typical Ash	Composition	
SiO ₂ , %			51.0
Fe,O, %			9.3
Al ₂ O ₃ , %			32.0
CaO, %			3.5
MgO, %			

Assistant Melting Superintendent, American Cast Iron Pipe Co., Birmingham, Ala.

ing with clay brick, experienced a maximum burnback averaging 6 in. per side. Daily patching cost was \$1.51 per ton, material and labor. During this study the pneumatic guns were introduced and the acid patching procedure was changed to the blowing of a ganister-clay mixture. Acid burn-back was reduced to 4.8 in. and, excluding the cost of the gun, the patching cost was reduced to \$1.10 per ton.

Comparison of refractory costs obviously must be considered only relative because of variations among cupola diameters, charges, and melting conditions. Within this plant refractory maintenance costs have ranged from \$0.25 per ton to \$2.50 per ton among the six cupolas operated daily. The lowest cost is experienced on a 35-ton/hr furnace operated on 18-hr heats, and the highest cost was on a 1½-ton/hr furnace operated on 8-hr heats. The 36-in. cupola in this study is midway between these extremes.

TABLE 3—COMPARATIVE BURN-BACK AND PATCHING
COSTS
(25-ton heats—36-in, cupola)

	No Dans	Mary Down backs	Detak
Patch Material	No. Days Averaged	Max. Burn-back* per Side, in.	Patch cost per ton**
Acid			
Clay brick	18	6.0	\$1.51
Ganister-clay, blown	4	4.8	1.10
Basic			
Magnesite brick	18	2.4	3.80
Magnesite, rammed	3	3.5	+
Magnesite, blown	4	3.8	2.68
Dolomite, rammed	4	4.0	+
Dolomite, blown	30	4.2	1.85

Difference in lining before and after heat at point of maximum attack.

In patching the basic cupola, magnesite brick properly laid gave the best performance, but the highest cost. A maximum burn-back averaging only 2.4 in. made it necessary to patch only once every 2 or 3 days, but costs were \$2.29 to \$2.70 per ton higher than acid costs.

Next, an attempt was made to improve basic costs by ramming a patch from granular magnesite. With the addition of 1 per cent bentonite, a magnesite ramming mixture, used for electric furnace hearths, was made to adhere successfully to the vertical walls of the cupola. The burn-back of this material was 3.5 in.

The next step was to try to blow such a magnesite mixture in one of the newly introduced guns. After some experimentation, increasing the bond and slightly altering the grain distribution, magnesite mixtures from two different manufacturers were blown with reasonable success. Average burn-back was 3.8 in. and the cost \$2.68 per ton.

Dolomite is the most available and least expensive of basic refractories, and the most attractive for cupola application. A stabilized dolomite mixture was first rammed for four days to determine relative performance of the material without the application variables from gunning. The application was good and an attack of 4.0 in. was less than that experienced from acid refractories.

With its practicability established, the dolomite mixture was blown with a gun. Increasing the bond slightly and increasing the proportion of course aggregate seemed to improve the performance of the particular gun. Records kept for 30 days showed an average burn-back of 4.2 in. Although not as stable nor quite as resistant as magnesite, dolomite gave us the lowest refractory cost because of its much lower price. Refractory maintenance cost was \$1.85 per ton, which is \$0.34 per ton higher than the original acid cost and \$0.75 per ton higher than current acid cost using blown ganister.

Dolomite refractories, by adding both CaO and MgO to the slag, produce slags higher in CaO than slags of equivalent basicity from MgO refractories. The higher CaO slags from dolomite refractories seem to have better fluidity and better desulphurizing ability than higher MgO slags. This is an advantage of dolomite refractories worthy of consideration. These refractory data are summarized in Table 3.

This cupola is equipped with water cooling by external sprays. As expected, the water cooling has not appeared to affect lining consumption until the lining has been worn very thin. However, this basic cupola has been operated a few days, and another acid cupola is operated regularly, with essentially no refractory lining in the melting zone. Without some alteration in design, erosion is severe around the tuyeres and in the basin. However, the author has found, like several others, that operation can be satisfactory with virtually no refractory in the melting zone.

It has not been determined to what extent reduced refractory consumption can offset the cost of the water. Water cooled acid and basic cupolas with either internal jackets or external sprays have been operated successfully by a number of plants both in Europe and the United States.

Experience with these five refractory combinations gave assurance that several basic materials can give reliable cupola performance and can be adapted to all fundamental types of patching. However, at the present time it appears that basic refractories can be expected to cost roughly twice as much as acid refractories. Economic justification can be expected only in applications where this refractory cost can be overcome by greater savings in the metal charge or improved iron quality.

Further research by the refractory manufacturers should bring further improvements in the adaptation of basic materials to cupola use and some improvements in cost. If refractory costs are not reduced further, extended application of water cooling should prove profitable.

#### Slag-hole and Tap-hole Refractories

Since a basic slag is more corrosive, slag-hole refractories with better than average slag resistance are required. In this plant the cupola has been operated both as a back-slagging, intermittent-tap cupola, and currently as a continuous-tap, front-slagging cupola.

When operated as a tap cupola, the back slag hole required much attention. Clay refractories were con-

^{**} Both material and labor costs of patching back to original dimensions.

[†]Insufficient experience with ramming to properly appraise labor cost.

sumed rapidly. Clay-graphite refractories held up sufficiently for only short heats of 2 or 3 hr. Slag holes rammed from both magnesite and dolomite have lasted satisfactorily for one shift when properly bonded and densely rammed.

Silicon-carbide slag-hole blocks generally lasted through a one-shift day and frequently for two days. Graphite and carbon slag-hole blocks (cut from scrap electrodes in the plant) proved completely resistant to the slag but were gradually burned larger by the escaping air blast when left open.

In continuous front slagging the refractory problem is moved to the tap hole, slag skimmer knife, and runner box. Tap holes rammed from both dolomite and magnesite have performed with consistent satisfaction for a one-shift operation on a slag of medium basicity. Frequently these have been run the second and third day with essentially no patching. Some early troubles were experienced when these materials were not properly bonded, tempered, or rammed.

A tap hole made up from high-alumina brick (98 per cent +  $\mathrm{Al_2O_3}$ ) has performed satisfactorily for two or three shifts but at a higher cost than the other materials. The runner box behind the slag knife is lined with magnesite brick with a 1-in, inner lining of rammed dolomite. The daily attack is confined to this dolomite layer which is patched after each day's operation. The slag knife or skimmer is made of magnesite brick, with one high-alumina brick as the base brick which lasts two weeks. The slag overflow and run-off trough are lined with thin graphite or carbon blocks which last two or three weeks.

Graphite and carbon materials are completely resistant to slag attack or adherence, but are attacked slightly by hypoeutectic iron if extended below the slag level. Carbon pastes seem to offer better metal resistance and equal slag resistance, plus the flexibility of a plastic.

The plant is currently experimenting with a continuous-tap system with a rear-slagging dam. Separating the metal and slag outlets should permit the use of materials with optimum slag-resistant properties for the back slag outlet and materials with metal resistance for the front tap hole, while retaining the advantages of continuous flow. Another advantage might be the independent adjustment of metal and slag levels within the cupola hearth. Some advantages might be found for highly basic slags or large slag volumes on long heats.

#### Effect of Varying Degrees of Slag Basicity

No particular chemical pattern is typical of basic operation. A basic lining permits varying degrees of slag basicity and varying chemical accomplishments. As slag basicity increases, carbon pickup increases and sulphur content decreases. The lower metal sulphur content is accounted for in the greater sulphur absorbing capacity of the more basic slag. Higher carbon pickup seems to be caused by the basic slag washing away the acid ash from the coke, leaving it cleaner and more reactive. An additional effect on carbon pickup seems to be caused by the increased carbon solubility of a lower sulphur iron.

Table 4 shows charge combinations for the acid

mix, and five different basic mixes of varying basicity that have been used to produce iron of the same chemistry, except for sulphur. The iron produced was an ordinary gray iron of 2.00 per cent Si, 3.40 per cent C, and 0.20 per cent P.

The basic cupola has been operated with no flux at all. In this size cupola the quantity of basic refractory consumed from the lining is more than enough to neutralize the acid from the coke ash and from silicon oxidation so that the resulting slag is slightly more basic than an acid cupola slag. The charge is not greatly different from an acid mix, only 5 per cent higher steel, and the final sulphur only slightly lower than acid experience.

As the flux is progressively increased, slag basicity increases, and final sulphur content decreases. To

TABLE 4—BASIC MIXES WITH VARYING SLAG BASICITY (Producing same final chemistry—Si, 2.00%; C, 3.40%; P, 0.20%)

Flux Charge	Proportion of Steel,%	Silicon Loss, %		Sulphur Final,%
Acid			-	
4% limestone	30	15	0.45	0.090
Basic				
None	35	20	0.65	0.075
1% limestone	40	24	0.85	0.070
2% limestone	50	27	1.25	0.060
4% limestone-				
1/2% fluorspar	60	30	1.60	0.050
1% CaC2, 4% lime	estone,			
1% fluorspar	80	30	2.40	0.030
D		Jane	annation and	based on

Rounded averages of several days operation and based on average raw material conditions.

Silicon loss is percentage of over-all silicon addition with 0.25% silicon inoculation added to ladle.

Carbon pickup is percentage carbon content above calculated carbon content of charge.

compensate for the increased carbon pickup, increasing proportions of steel scrap have been used. As shown in Table 4, charges of 80 per cent steel have been melted with a flux of 1 per cent calcium carbide, 4 per cent limestone, and 1 per cent fluorspar.

The flux quantities in Table 4 are based on charges of comparable cleanliness and physical condition. More flux has been used with dirty charge material and for thin, rusty scrap to obtain a slag of the same basicity.

Silicon loss is much higher in basic melting, and adjustments must be made for the same factors. No manganese is used on the basic cupola. The better manganese recovery leaves a manganese content of 0.45 to 0.70 per cent, which is sufficient to neutralize the reduced sulphur content of 0.040 to 0.070 per cent obtained under most conditions.

#### Nodular and Special Irons

The basic cupola has produced irons beyond the chemical limitations of the acid cupola. Many of these chemical extremes have desirable metallurgical properties. Table 5 shows several special irons made advantageously in the basic cupola.

The first is a nodular iron heat for casting research. A 0.010 per cent sulphur and a 3.90 per cent carbon were obtained with a flux charge of 4 per cent carbide, 4 per cent limestone, and 2 per cent fluorspar.

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TABLE 5—SPECIAL IRONS MADE IN BASIC CUPOLA (Typical Short Heats—36-in, Production Cupola)

				Metal C	chemistry, %	b
Type Iron	Metal Charge	Flux Charge	S	Si	С	Alloys
1. Nodular	50% Steel	4% CaC ₂				
(Test Castings)	50% Low P Pig	4% Limestone				
, ,	,,	2% Fluorspar	0.010	2.07	3.90	Mg
2. Permanent Mold	30% Steel	6% Limestone	0.040	1.79	3.80	Cr
	70% Pig	2% Fluorspar				
3. Forging "Die Pots" (A)	100% Alloy return	4% Limestone	0.061	1.10	3.97	Ni, Mo, Cr
(-)	70	1/2% Fluorspar				
(B)	100% Alloy return	2% Limestone	0.073	1.22	3.70	Ni, Cu, Cr
4. Nickel Alloy	100% Alloy return	6% Limestone	0.082	2.18	2.84	Ni, Cu. Cr
	70	1% Fluorspar				
5. High Strength	90% Steel	1% Limestone	0.082	1.67	2.87	

In the second iron for permanent molds a very high carbon content is desired for heat-shock resistance. A carbon content of 3.80 per cent and higher has been regularly obtained in the basic cupola. Iron No. 3 (Table 5) for forging die pots is another heat-shock application in which a high carbon is desired. This iron is alloyed with nickel, molybdenum, and chromium.

With 2 per cent limestone a 3.70 per cent carbon and 0.073 per cent sulphur were obtained from 100 per cent return scrap which consisted of used dies from the customer plus shop return. The scrap composition averaged 3.25 per cent carbon and 0.095 per cent sulphur. With 4 per cent limestone and ½ per cent spar the more basic slag produced a 3.97 per cent carbon and a sulphur of 0.061 per cent. In acid practice no more than 60 per cent return could ever be used, and then carbon was not as high as desired and a sulphur of about 0.120 per cent required ladle desulphurization.

Iron No. 4 is a nickel alloy iron on which basic melting made it possible to melt 100 per cent returns of the same iron. Sulphur was reduced from 0.120 per cent in the charge to a final of 0.082 per cent sulphur, and the carbon of 2.84 per cent was advantageously higher than could be obtained in acid melting.

Iron No. 5 is a low carbon-high strength iron with over 47,000 psi tensile in a 2-in. test bar. The basic cupola seems to offer no particular advantage on this type of iron. Low carbons are naturally more difficult to obtain from a basic cupola because of its tendency toward higher carbon pickup. As shown by Renshaw² the interrelated reactions make it difficult to obtain a carbon below 2.80 per cent with a sulphur below 0.050 per cent. However, higher sulphurs do not appear as detrimental in the larger section sizes into which iron of this type is cast.

#### **Economic Considerations**

By reducing the flux and the slag basicity and allowing the sulphur to run higher, low-carbon irons may be produced with fair success in the basic cupola. This flexibility is important where a single furnace must be used for various types of irons, some of which would be benefited by basic melting.

Table 6 shows comparative metal costs of the regular acid mix and two of the basic mixes shown in Table 3. This is a typical application in a particular

raw material situation. By replacing \$70 per ton low phosphorus pig with \$48 per ton steel scrap the 60 per cent steel basic mix saved \$3.50 per net ton over the acid mix on charge material. After subtracting the cost of basic refractories the net saving is still over \$2.50 per ton.

By the use of 1 per cent calcium carbide and with the steel proportion increased to 80 per cent, another \$1.80 per ton has been saved over the 60 per cent steel mix. Prices are local and current. This is an example of one of several raw material savings under various local conditions, made possible by the flexibility of basic operation.

#### Smaller Magnesium Additions Required

In making the nodular iron shown in Table 5 the basic low sulphur iron required only one-third the magnesium alloy addition needed for the usual acid iron. This reduction in magnesium alloy amounts to approximately \$10.00 per ton. On the same heat a saving of \$7.00 per ton was realized on the base metal charge by substituting steel for low phosphorus pig iron.

After liberal allowance for basic refractories and special fluxes, the net saving on this iron can be \$10.00 to \$12.00 per ton. From the cost and control standpoint the basic cupola seems to offer the most advantages on nodular iron. Flinn and Kraft³ reported such advantages in the regular production of this type of iron in a basic cupola.

On the permanent mold iron shown in Table 5 the net saving was \$4.00 over an acid mix producing less desirable chemistry.

On the forging die iron in Table 5 the basic cupola mix of 100 per cent alloy return (used dies from customer plus shop return) effected a net saving of \$18.00 per ton from the use of cheaper scrap plus the value of the alloys reclaimed.

Successful use of poorer grade coke is another potential advantage of basic operation that is difficult to evaluate. Low sulphur iron sufficiently high in carbon has been produced from coke of 15 per cent ash and 2.00 per cent sulphur content—too poor for acid use.

These examples illustrate some of the various applications on which the cost of basic refractories can be more than offset by greater savings in the metal charge. In abnormal times relative prices may be distorted but the supply of select materials is generally insufficient. In such cases basic operation can prove

valuable in extending the melt tonnage even if economies are not realized.

On the other hand basic operation cannot be justified in many applications. In this plant where high phosphorus content is satisfactory and a sufficient supply of high phosphorus pig iron is locally available, there is no advantage in using more steel.

In a high silicon-low carbon iron basic operation was not found to offer any net cost advantage. Forehearth desulphurization could be eliminated but the increased silicon loss added to the refractory cost made basic melting more expensive.

### Slag Composition

In basic cupola melting, slag composition is important since the reactions obtained depend upon slag chemistry. Methods of sampling and analysis need further standardization and development and many relationships remain to be determined with better accuracy. However, the general factors suggested in the previous paper have been found to apply with reasonable dependability. Desulphurization efficiency is generally indicated by the sulphur content of the slag compared to the sulphur content of the iron.

Desulphurization efficiency of the slag improves with increased basicity and decreased iron oxide content. Metal desulphurization may be increased by melting larger volumes of slag. However, equal and better desulphurization efficiency can be accomplished by melting smaller volumes of slag of more effective composition.

For a simple indication of basicity the author has

used the ratio — Flinn³ included Al₂O₃

SiO₂
in the denominator. In the author's cupola experience, Al₂O₃ has appeared to have relatively little effect on desulphurization as compared to SiO₂. For that reason Al₂O₃ has been left out of the denomin-

TABLE 6—COST COMPARISON OF CHARGE METAL (Average analysis—2.00% Si; 3.40% C; 0.20% P)

		Acid	Basic	Basic+ Carbide
Steel Scrap	\$48.00/GT*	30%	60%	80%
Low P Pig	\$70.00/GT	50%	20%	None
High P Pig	\$48.50/GT	20%	20%	20%
Silicon Briquette	0.07/lb	3%	5%	5.5%
Spiegel	0.04/lb	0.5%	None	None
Flourspar	0.02/lb		0.5%	1.0%
Carbide	0.06/lb			1.0%
Total metal cos	t/net ton	\$57.70	\$54.20	\$52.40
Saving in me	tal charge	\$3	50 \$1.	80

 All prices necessarily local and current at time and only of relative value under similar circumstances.

ator until its relative effect can be more clearly established. CaO and MgO would not be expected to have equal desulphurizing effectiveness in all ranges.

However, in the ranges encountered, variations in excess basicity are generally found in the CaO content with the MgO essentially within the amount necessary to neutralize the acids. From both the fluidity and desulphurizing standpoint it appears advantageous to keep the CaO much higher than the MgO. Further slag research should develop a more accurate relationship of these constituents.

Table 7 shows some typical slag compositions representative of several degrees of basicity. These have been arbitrarily grouped into mild basicity with basicity ratios 1.0 to 2.0, moderate basicity with ratios 2.0 to 3.0, and high basicity above 3.0.

Slag pancacke samples taken at intervals have given reasonably good indications of the basicity range. Some degree of slag control has been accomplished by examination of color and surface luster of slag samples. The mildly basic slags are generally dull black tending toward brown as basicity goes up and FeO comes down. As the basicity drops near or be-

TABLE 7—TYPICAL BASIC CUPOLA SLAGS WITH VARYING DEGREES OF BASICITY

	Metal	Metal				Slag (	Compo	sition,	%			Basicity	
Flux Charge	Charge	Sulphur, %	SiO ₂	Al ₂ O ₃	FeO ^b	MnO	CaO	MgO	F	CaC ₂ ^d	S*	Ratio ^a	Slag Appearance
Mild Basicity		-											
1. First Ladle, 5% Lst, 1% Sp.	50% Steel	0.079	36.5	11.5	13.4	2.6	26.1	5.7	0.5		0.04	0.87	Glassy, black
2. No Flux	35% Steel	0.074	37.5	12.8	3.1	1.8	23.2	20.1		_	0.28	1.15	Vitreous, dark black
3 1% Limestone	40% Steel	0.061	34.3	9.4	1.7	1.7	30.5	22.0			0.39	1.53	Dull, black
4. 2% Limestone	50% Steel	0.064	35.1	7.8	2.2	1.7	35.0	16.8			0.46	1.47	Dull, brownish black
Medium Basicity													
5. 4% Lst, 1/2% Spar	65% Steel	0.047	28.6	6.5	1.8	1.2	41.2	19.6			0.56	2.12	Dull, brownish gray
6. 4% Lst Hot Blast!	80% Steel	0.030	26.0	9.2	2.9	0.6	32.5	28.6	1.0	(0.2)	1.26	2.35	Dull, dark gray
7. 1% CaC2, 4% Lst, 1% Spar	80% Steel	0.032	25.8	6.5	2.0	0.6	40.3	23.2	0.9	(0.4)	3.08	2.46	Dull, brownish gray
High Basicity													
8. 4% CaC, 4% Lst, 2% Sp.	50% Steel	0.010	18.8	12.5	0.4	0.6	62.4	9.0	2.5	(2.5)	2.50	3.80	Powder, gravish white
9. 5% CaC, 5% Lst, 3% Sp.	All pig	0.005	15.8	15.2	0.2	0.1	59.3	8.6	1.6	(4.9)		4.29	Powder, gravish white
100% Scrap Charges with		asicity											
10. 3% Limestone	100% Return	0.069	35.4	10.1	1.9	1.9	34.1	16.2			0.46	1.42	Dull, brownish black
11. 4% Lst, 1/2% Spar	100% Return		32.4	11.7	1.2	0.6	39.0	15.1			0.62	1.67	Dull, medium brown
12. 6% Lst, 2% Spar	100% Return		30.6	7.8	0.7	0.8	42.3	15.2			0.70	1.88	Dull, light brown
Total CaO													

* Basicity ratio: SiO₂

^b Total Fe₂O₃ and FeO determined as FeO. ^e Total calcium determined as CaO.

⁴ Calcium in CaC₂ content not subtracted from total calcium as CaO because of less reliable analysis.

Total sulphur determined by gravimetric precipitation as BaSO₄.

450 F preheated air.

low 1.0 and the FeO goes high, the slag approaches a glassy luster characteristic of acid slags.

In the medium basicity range the slags become dark gray or brownish gray. In the high basicity, very low FeO ranges, the slags become lighter gray. In the highest CaO/FeO ratios the slags disintegrate upon cooling into a white powder, and small quantities of calcium carbide are found in the slag analysis.

#### Operation and Control

On comparable types of charges and coke ratios the basic cupola without any exothermic fluxes seems to melt slightly faster and at a temperature that averages possibly 20 to 30 degrees hotter at the 2750 F level. This seems to be accomplished by the cleaner more active coke washed by the basic slag. The advantage of basic melting in this respect seems to increase with high-ash coke and dirty charges. Because of better combustion, coke reductions can be made with resultant increases in output if chemical objectives are not defeated.

With carbide additions to the charge basicity can be greatly increased, and much greater increases in temperature, desulphurization and carbon pickup have been produced. Further reductions in coke and increases in output have been accomplished. However, when the basic cupola is justified it is primarily for improved chemistry or raw materials adaptation, and any improvements in output generally are only incidental.

The effect of preheated blast is much greater on the basic cupola than on the acid furnace. In addition to the usual thermal boost, the basic cupola becomes chemically more effective. Much higher temperatures and slags of higher basicity are obtained, resulting in lower sulphur contents and higher carbon pickup. Hot blast makes the basic cupola potentially more basic. Operating the cupola for a few days on a 450 F blast required a coke reduction of 25 per cent and a steel increase from 60 to 80 per cent for a comparable carbon content.

A mildly basic slag is very fluid and maintains a very clean cupola. As the basicity increases the more typical basic slag chills more abruptly than an acid slag, especially in small quantities. The slag runoff requires more attention.

#### Careful Control Required

From the standpoint of chemical control the basic cupola is generally more complex and requires more careful supervision. Its complexity increases with the amount of work done by the basic slag. It has advantages in some respects and disadvantages in others. From the standpoint of sulphur control and a sufficient manganese to sulphur ratio, control of the basic cupola is unquestionably superior.

Carbon control appears to be better on a high carbon iron, almost equal on a medium carbon iron, and more difficult on a low carbon iron. Silicon consistency can be made equal to acid performance but is inclined to be a little more difficult to control because of the greater loss. However, in basic operation the charge frequently consists of more scrap of less predictable chemistry, or more of the silicon is added as concentrated alloy because of more steel. These less consistent charge materials increase the risk of chemical variations within themselves irrespective of basicity. A receiving ladle of sufficient capacity levels out many of these irregularities.

Since most reactions in the basic cupola depend upon slag chemistry, factors effecting slag chemistry require better control. The flux must be calculated, weighed, and charged with as much accuracy as any ingredient. Consistent refractory consumption, uniform charging, and constant quantities of dirt on the scrap are necessary for consistent slag control, because of their indirect contributions to slag chemistry. Flux adjustments have been found necessary to compensate for changes in the physical condition and cleanliness of steel and cast scrap.

From the standpoint of physical quality basic iron appears to have slightly better mechanical properties. The difference seems to be due chiefly to the lower sulphur content. With comparable chemistry the lower-sulphur iron seems less inclined toward chill, slightly lower in hardness, higher in deflection, and higher in strength to hardness ratio. These differences are more advantageous on thin-section castings. The difference was more apparent on small diameter test bars. In larger diameter test bars the difference decreases to little practical significance.

#### Summary

The present work indicates that after some experience a basic cupola can be operated without serious trouble. Several basic refractories and methods of application have been used with fair success for cupola lining and patching. Refractory costs may be expected to be twice as high as acid refractories. Due to refractory cost and increased silicon loss, basic operation cannot be justified where sufficient materials of good quality are available to produce acid iron with no quality problems. However, the added costs have been easily offset in a number of different applications either by improved metal quality or use of cheaper or more available raw materials.

With increasing experience chemical control is becoming almost as good as in acid practice. More technical supervision and attention is required on refractory practice, flux additions, and charge condition as they affect slag composition.

By control of flux additions and slag condition a basic-lined cupola may be operated at several levels of basicity. A less complex mildly basic slag may be used where the only objective is freedom from high sulphur troubles. A moderately basic slag may be used to obtain a lower sulphur and/or higher carbon iron with quality advantages for some applications, or the same chemistry may be obtained with larger proportions of cast scrap, steel and/or poor coke with economic or availability advantages.

The basic cupola enables better utilization of poor grade materials along with the diminishing supply of select materials, reclamation of scarce alloys from scrap, and conservation of manganese which is not needed in basic melting. The chemical extremes made possible by the more complex slags with a basicity ratio over 3.0 have proved advantageous in melting nodular and other special irons.

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#### Acknowledgments

The author wishes to acknowledge the assistance of the melting department, and especially cupola foreman Glenn Jones, in operating the cupola. Analytical data were obtained through the cooperative efforts of the laboratory staff. The author appreciates the permission to publish this work granted by the management of the American Cast Iron Pipe Co.

#### Bibliography

1. Sam F. Carter, "The Basic Lined Cupola," A.F.S. Trans-ACTIONS, vol. 58, pp. 376-392 (1950).

2. E. S. Renshaw, "Basic Cupola Melting and its Possibili-

ties," A.F.S. Transactions, vol. 59, pp. 20-27 (1951).

3. R. A. Flinn and R. W. Kraft: "The Importance of Slag Control in Basic Cupola Operation," A.F.S. Transactions, vol. 59, pp. 323-331 (1951).

#### DISCUSSION

Chairman: J. T. MACKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: T. J. Wood, American Brake Shoe Co., Mahwah, N. I.

Recorder: H. W. LOWNIE, Jr., Battelle Memorial Institute, Columbus, Ohio.

W. R. JAESCHKE (Written Discussion): 1 As usual Mr. Carter has prepared a thorough paper and presented his data in such a precise manner that there is little occasion for questions.

However, most of the data on basic cupola operation has dealt with gray and alloy irons and indicates a relatively high degree of carbon pick-up in the melt. For duplex melted malleable irons requiring 2.70 per cent carbon or less, a low carbon pick-up in the cupola is necessary. A charge for a malleable duplex melting cupola usually contains 50 to 55 per cent sprue (foundry returns), 10 per cent of silicon-bearing and other alloys and a maximum of 35 to 40 per cent steel scrap. With the maximum of 40 per cent steel, the carbon content of the charge would be from 1.40 to 1.55 per cent carbon depending on whether the sprue were of 2.30 or 2.55 per cent carbon.

With such a charge, would Mr. Carter estimate the carbon pick-up to be as low as 0.85 per cent as shown in his Table 4, for 40 per cent steel mixture and a mildly basic slag?

Under acid slag operation, the sulphur in duplex melted mal-

leable irons runs as high as 0.18 per cent instead of 0.09 per cent as shown in Table 4 for acid practice. With a mildly basic slag, would Mr. Carter estimate the sulphur to be as low as 0.10 or 0.12 per cent?

If basic operated cupolas could not produce iron under 2.70 per cent carbon and 0.10 per cent to 0.12 per cent sulphur from a mixture as mentioned it would not be worthwhile for malleable foundry duplex melting except possibly for those producing the high carbon cupola grade of malleable iron.

MR. CARTER (Written Reply to Mr. Jaeschke): Because of the tendency toward higher carbon pick-up basic melting is not so obviously advantageous on low carbon irons like malleable.

Low carbons in the malleable range can and have been obtained with only a mildly basic slag but with reduced desulphurizing efficiency.

In my paper "Basic-Lined Cupola for Iron Melting," A.F.S. Transactions, vol. 58 (1950), pp. 376-392, examples of some lower carbon irons were given. A high silicon malleable iron was produced with a mildly basic slag. A charge of 60 per cent steel and 30 per cent return, calculating 1.60 per cent C, produced final carbons averaging 2.59 per cent and a sulphur of 0.09 per cent compared to 0.12 per cent S from an acid cupola. I believe now lower sulphurs and carbons could have been obtained with

On some experimental heats using a high sulphur, high ash coke a 70 per cent steel charge calculating 1.20 per cent C produced a 2.40 per cent carbon and a 0.14 per cent S compared to 0.24 per cent S from acid melting.

Renshaw reported that from a charge calculating 1.40 per cent carbon, final carbons of 2.60 per cent to 2.90 per cent are obtained with sulphurs of 0.040 to 0.070 per cent. Lower slag basicity should lower the carbon and raise the sulphur.

The first basic melting of acid sprue might hardly reduce the sulphur from 0.180 per cent to the 0.10 to 0.12 per cent range if carbon is held low. But recycling the basic sprue might drop the sulphur to a 0.07 per cent to 0.09 per cent S equilibrium before experiencing excessive carbon absorption. Whether this will reduce annealing time sufficient to offset basic refractory costs I could not say. Neither could I guarantee such low car-bons with successful control. On the other hand cleaner cupola performance and reduced coke consumption might prove sufficiently advantageous to favor basic operation. With no experience I cannot predict a net advantage but would like to see basic operation given a thorough trial on malleable iron.

¹ The Whiting Corp., Harvey, Ill.

# SOME PRACTICAL APPLICATIONS OF PERMEABLE METAL CASTING PLASTER

By

R. F. Dalton*

#### ABSTRACT

Permeability of plaster molds for non-ferrous metal casting is discussed. The process describes the entraining of air into a formulated plaster mix wherein permeability in the mold is created by the connections between the finely divided air cells. The cellular mold structure is made by mixing with a rapidly rotating rubber disc which operation incorporates air cells of the proper size. Making molds for typical aluminum castings by this process is described.

Use of plaster as a mold material for non-ferrous metal casting began thousands of years ago. The use of plaster investment molds for the production of bronze statuary was a well practiced art even in the middle ages.

The past 25 years has seen an increase in the use of plaster as a molding material for non-ferrous metal casting. The successful incorporation of 20 to 25 per cent long fiber talc with plaster, has resulted in a suitable mold material for plaster mold castings. This mixture has been especially adapted to the so-called "Capaco" process and the pressure-cast process. The latter method is used extensively for production of pressure-cast matchplates,

Use of plaster mold formulations for these casting methods has one common limitation—low permeability. In the "Capaco" process, the mold is dried at 1500 F for a few hours to remove all traces of water. The molds may then be poured by gravity. In matchplate practice, the molds are dried for longer periods of time (12 to 24 hr) at 300 to 500 F. To aid in casting the metal (because the mold has a very low permeability) air pressure is used. Thus the objection of low permeability in plaster molds has been met by high-temperature "burnout," and by pressure casting.

If molds could be quickly produced with the smoothness of plaster and the permeability of sand, then many of the problems of these plaster molds processes would be eliminated.

#### Theory of Permeable Plaster

A new method of producing permeable plaster molds was described by K. Miericke and E. Johnson¹ in 1948. Use of this new plaster for producing permeable plaster riser sleeves was described by K. Miericke² in 1949. Mention of this plaster was made by the author³ in 1950.

The method of producing a permeable plaster mold is essentially the formation of a myriad of perforated air cells (Fig. 1) throughout the *interior* of the mold. These interconnecting air cells are the channels through which steam and even the air in the mold cavity itself can escape during the metal pouring operation. This ability is known as permeability, and it should be noted that permeable plaster differs from sand in the way the gas escapes through the mold. In sand molds, the gases pass through voids around the sand grains, whereas in permeable plaster, the gases pass through interconnecting air cells contained in the paster mold.

Properly made and dried molds may be produced with a much higher permeability than is otherwise obtained with conventional metal casting plasters. The process has the further advantage of using less



Figure 1-Mag. 25x

^{*}Development Engineer, United States Gypsum Co., Chicago.

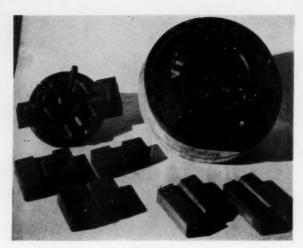


Figure 2

water, and shorter drying periods. Molds with the permeability of sand molds are readily made, dried at low (250 to 500 F) temperatures and readily poured by gravity.

By using this type of plaster formulation and a new mixing technique, molds are made with plaster mold smoothness on the surface, "backed up" by a rigid sponge-like mass of cellular plaster. Thus molds may be produced with the surface smoothness of plaster and the permeability of sand. This increase in permeability has several advantages:

 Molds may be poured by gravity. Heads and gates may be scientifically planned according to good foundry practice.

2. Calcination of the mold is not required. Steam formed during the pouring operation may escape readily through the permeable plaster mold. It has been suggested that the formation of steam has a "chilling" action of the metal, resulting in a smaller grain size and higher physical properties.⁴

3. Molds are more "true to size." Since calcination



Figure 6

is unnecessary, high temperature "burn out" is not needed and mold shrinkage is minimized.

#### Shop Application

The following is a description of how permeable plaster was used to make cored, aluminum water pump castings for a newly designed washing machine. The customer of the matchplate shop wanted 50 complete sets of castings. This would allow him to check the design, investigate machine shop procedure and to a limited extent determine customer approval on the finished article. The matchplate shop attempted to make the castings by the pressure cast method in "regular" matchplate plaster. Because of the small cored holes and the use of pressure in the casting operation, they were unable to produce satisfactory pressure tight castings. Using the permeable plaster

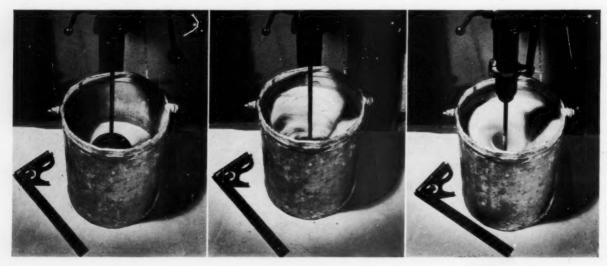


Figure 3

Figure 4

Figure 5



Figure 7

technique, they successfully filled the order.

The pattern equipment required to make one of the castings is shown in Fig. 2. The pattern and core boxes were made of mahogany to close dimensions. The patterns and core boxes were water-proofed with lacquer and a thin coat of separator applied (regular foundry liquid parting and water ground mica powder).

The mixing operation for permeable plaster varies considerably from "regular" matchplate practice. A rubber disc of the proper size is used instead of the conventional propellor type mixer. Machine mixing with high shaft speeds is essential. (Note the use of the drill press in Figs. 3, 4 and 5). Attempting to produce the proper permeable structure by hand mixing is practically impossible. The mixing operation may be divided into three phases:

1. Add a weighed amount of permeable plaster to a measured quantity of water (Fig. 3). Allow this to soak until the plaster is wet and then mix for a few



Figure 9

seconds until a uniform slurry is produced.

2. Raise the mixing disc in the center of the bucket until half of it is visible (Fig. 4). The rapidly rotating disc will then entrain air into the mix. Continue to raise the disc as the volume increases until the original volume has been "expanded" by 50-75 per cent. (The amount of air added depends on the permeability desired.⁵)

3. Lower the disc in the center of the bucket, below the surface of the mix, so that the disc is completely covered (Fig. 5). Large air bubbles will then be drawn to the center of the bucket and down over the rapidly rotating disc. This action "cuts up" or "shears" the large air bubbles to a smaller size. Continue to mix until they are barely visible.

In this manner, a uniform mix is made of water, permeable plaster and air, all in definite proportions.

The prepared mix is then poured (Fig. 6) over the properly prepared patterns and into the core boxes. The thin film of separator, previously applied, also



Figure 8



Figure 10



Figure 11

acts as a "foam breaker" to produce a thin film of "air-free" plaster on the surface of the pattern.

After a period of 15-25 min, the plaster is "set" sufficiently hard to extract the pattern. One method of extraction (Fig. 7) is to form a small hole to the pattern surface (through the permeable plaster) and then blow air gently down the hole. The air slowly forces the mold away from the pattern surfaces. Other methods used in the matchplate shop may be employed.

The completed mold and core assembly is shown in Fig. 8; plaster mold smoothness on the surface, "backed up" with a rigid sponge-like mass of permeable plaster. The molds are then dried in conventional foundry core oven, Fig. 9. These particular molds were dried for 4 hr at 400 F. Note the moisture meter in the foreground in Fig. 9. This instrument will detect "free" water and when the reading is below 1 per cent, the mold should be dry enough to cast the metal. The amount of water that must be removed is determined by the amount of metal poured, its temperature, and the ability of the mold to vent the steam formed during the pouring operation. This "venting ability" is known as the permea-

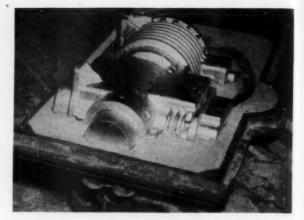


Figure 13

bility of the mold.

The molds were then poured by gravity using No. 43 aluminum alloy (Fig. 10) at 1250 F. Thus castings may be gated and risered according to good foundry practice. No sudden "rush" of metal as in the pressure casting method, and any steam and even the air in the mold cavity itself is readily vented through the permeable plaster mold.

The completed water pump assembly is shown in Fig. 11 together with a single casting and one that is sectioned to show the cored passages. Other examples of aluminum castings produced by the permeable plaster technique are shown in Figs. 12, 13, and 14. All of these castings were poured by gravity, thus eliminating the need for, and the disadvantages present in the pressure casting method.

#### Conclusions

The difficulty encountered due to the low mold permeability of conventional plaster molds is met by high temperature "burnout" and pressure casting methods. In mixing permeable plaster, high speed disc mixers are advisable in order to produce a uniform cellular structure in the mold. The use of permeable

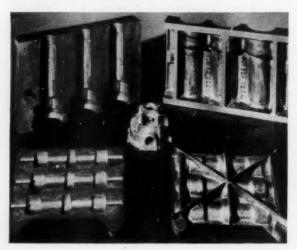


Figure 12

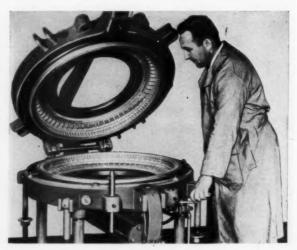


Figure 14

plaster allows gravity pouring of non-ferrous metals with low temperature mold drying. The use of permeable plaster molds should permit the production of "sounder," smoother and more accurate non-ferrous castings.

#### Acknowledgments

The writer wishes to express his appreciation to Mr. Steven Denkinger of the Atlas Plastic and Pressure Matchplate Company of Butler, Wis., and to Mr. James C. Heintz, Sr., of the J. C. Heintz Co. of Cleveland, for the use of many of the photographs shown in this paper.

## Bibliography

- 1. K. Miericke and E. Johnson, "A New Permeable Metal Casting Plaster," Transactions, A.F.S., vol. 56, pp. 479-485 (1948).
- 2. K. Miericke, "Recent Developments in Theory and Practice of Insulating Sleeves, Pads and Risers for Non-Ferrous Transactions, A.F.S., vol. 57, pp. 204-209 (1949). Casting,"
- 3. R. F. Dalton, "Production Patterns and the Matchplate," Transactions, A.F.S., vol. 58, pp. 699-704 (1950).
- 4. H. Rosenthal and S. Lipson, "Precision Casting Aluminum in Moist Investment Molds," Transactions, A.F.S., vol. 58, pp. 460-464 (1950).
- 5. I.G.L. Bulletin No. 35, "Hydroperm, Permeable Metal Casting Plaster," United States Gypsum Co., Chicago.

#### DISCUSSION

- Chairman: HIRAM BROWN, Solar Aircraft Co., Des Moines,
- Co-Chairman: W. E. Sicha, Aluminum Co. of America, Cleveland.
- Recorder: HIRAM BROWN.
- MEMBER: Is it possible to produce magnesium alloy castings in permeable plaster molds?
- MR. DALTON: It is necessary to add inhibiting agents to permeable plaster molds for magnesium alloy castings. Sulphur and boric acid, or the fluosilicate of sodium, calcium or potassium, may be introduced into the plaster mix in order to provide the necessary protection. The fluosilicates retard the setting time and boric acid extended the setting time ex-

- K. D. Love:1 What are the collapsibility characteristics of permeable plaster cores or, in other words, what is the breakdown temperature?
- MR. DALTON: Core hardness and collapsibility can be influenced by the amount of water added to the plaster and the amount of air stirred into the slurry. However, it may be necessary to wash the cores out with water.
- C. M. LAMBE: In drying plaster molds or cores, the free water is removed first, then the combined water is driven off and the gypsum dissociates at 1700 F.
- W. P. O'BRIEN: Is it possible to substitute some liquid other than water in the slurry? It appears that it would be preferable to employ a material like sulphur which would go through the solid, liquid, solid cycle.
- MR. DALTON: At present, water is essential in slurries for permeable plaster molds although a substitute material of the type described may be developed in the future.
- MR. O'BRIEN: My interest in a substitute for water is an economic one because in producing a permeable plaster mold for brass casting the cost of materials was 10 to 11 cents whereas the drying cost was 26 cents.
- M. A. BEAN:4 The mechanical properties of castings produced in plaster molds also are of great importance. Attainment of satisfactory properties generally requires that the plaster molds be reasonably heat conductive. As an example, if the heat absorptive power of the sand mold is rated as 100, the comparable figure for a permeable plaster mold would be
- MR, DALTON: Permeable plaster molds obviously should not be used where the properties produced in the casting would not be adequate for service requirements. Chills may be used in the plaster molds to assist in accelerating the solidification of castings, and plaster molds provide an advantage in producing castings with thin sections.
- MR. BEAN: We produced crankcases during the last war that were 4 ft long and weighed 190 lb in plaster molds which had to have the permeability of sand molds.
- O. A. CARLSON:5 How can the heat of reaction in setting of the plaster be reduced? Will this heat of reaction damage the parting materials applied to patterns?
- MR. DALTON: It is generally preferable to employ waxes with relatively high melting temperatures as the parting material. Carnauva wax is especially suitable and commercial floor waxes may be used to waterproof wood patterns and also provide a parting.

The Colonial Foundry Co., North Canton, Ohio.

*United States Gypsum Co., Chicago.

*W. B. Allen Mfg. Co., Bellwood, Ill.

*Morris Bean Co., Yellow Springs, Ohio.

*Carlson Pattern Shop, Springfield, Mass.

## THE NATURE OF BONDING IN CLAYS AND SAND-CLAY MIXTURES

By

L. M. Diran* and H. F. Taylor**

#### Foreword

The chief purpose of this paper is to re-emphasize the fact that sand-clay mixtures are relatively complex aggregates; and that a really basic interpretation of bonding action will be needed before true improvement can be made in molding mixtures. Current understanding of clays is sketchy at best, but our knowledge is being increased steadily by ceramists, chemists, and physicists—the technology of siliceous aggregates is so intriguing and complex that Dr. Hauser at M.I.T. is suggesting a new science to be known as Silicic Chemistry.

Not many years ago the authors considered clay a simple chewing-gum-like, material with a convenient propensity for sticking sand grains together. There must be many foundrymen who still think this to be the case, in spite of the fact Norton, Hauser, Ries, Endell, Grim, and others have written that the situation is much more complex.

At any rate it seems desirable that foundrymen take a look at sands and clays periodically through the eyes of their more scientific brethren; this look might inspire a fundamental idea for improving our ability to make sand and clay a more obedient and dependable servant. This look will still leave many things unanswered (particularly from a foundry standpoint); it will surely be incomplete and perhaps even incorrect since the science of the subject is in its early stages. But it is comforting to know we can at least view the field as a partially developed science.

The bibliography at the end of the paper is perhaps the most valuable feature; this will surely be so if the readers will take time and trouble to read and study them. The authors found the scientific interpretation of bonding action described herein a valuable aid to research work which will be published later. It is hoped others find the same inspiration, and that foundry science will be furthered as a result of it.

In making molds for metal casting, two general types of molding sand are used. One is naturally bonded and the other is bonded synthetically.

Naturally bonded molding sands consist of quartz grains, clay, silt, and sometimes organic matter combined as nature left them. The quartz grains range from about 100 to 800 microns in their largest dimension; they may be rounded, subangular or angular in shape depending upon the geological history of the deposit. The clay particles—usually kaolin—are thin hexagonal plates that range in size from 0.10 to 44 microns. The silt consists of inert ingredients—for the most part fine silica. Organic matter is probably decomposed vegetation.

Synthetic molding sands are produced by mulling a plastic clay with dry silica sand, preferably of known particle size and distribution. The sand used may contain as much as 1.50 per cent residual clay, or may be essentially clay free.

Pure silica sand has no bonding ability; it cannot be worked into a self-supporting shape. For this reason, clays capable of developing bond are required. The clays most commonly used in the foundry are sodium and calcium bentonites, and fireclays containing kaolinite and illite.

The bentonites are highly plastic clays of the montmorillonite group, derived from the chemical and mechanical weathering of volcanic lava; their ideal crystal structure is shown schematically in Fig. 1. The essential difference between sodium and calcium montmorillonite is exchangeable sodium and calcium ions attached between silicon-oxygen layers of the montmorillonite crystal. The character of the adsorbed ions determines the bonding power of the clay. Sands prepared with calcium bentonite have higher green compressive strength than those containing sodium-bentonite. Sodium bentonite bonded sands, on the other hand, have much higher dry compressive strength. Bentonite bonded sands generally contain 3 to 6 per cent clay and 21/2 to 4 per cent water.

Illite and kaolinite clays also (Figs. 2 and 3) devel-

Introduction

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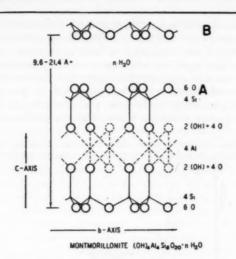


Fig. 1—Schematic presentation of montmorillonite after Hofmann, Endell and Wilm.¹

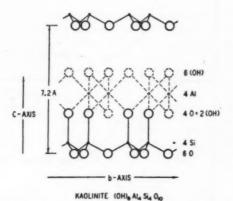


Fig. 2—Schematic structure of kaolinite crystal after Gruner.²

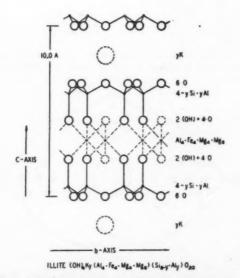


Fig. 3—Schematic structure of illite crystal after Grim, Bray and Bradley.³

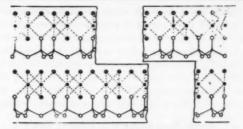


Fig. 4—Fractured kaolinite crystal showing unsaturated areas due to broken valence bonds after Johnson and Norton.⁵

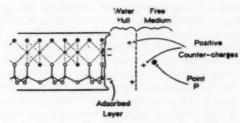


Fig. 5—Micelle of a kaolinite fragment showing position of adsorbed cations after Johnson and Norton.⁵

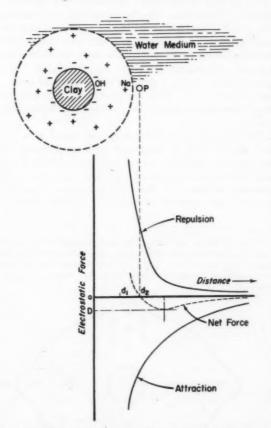


Fig. 6—Schematic sketch of a sodium-kaolinite micelle in a water medium; the hydroxyl ions are represented by negative signs and the positively charged counterions by positive signs according to Johnson and Norton.⁵ The electrostatic forces associated with the micelle are plotted as functions of the distance from the site of the negative surface potential.

op useful bond strength when properly hydrated. Illite bonded sand contains from 5 to 7 per cent clay, and 4 to 6 per cent water is generally required to develop maximum bonding strength. Kaolinite bonded molding sand contains approximately 15 per cent kaolin and the moisture required to develop its maximum strength ranges from 5 to 8 per cent. Both illite and kaolinite clays develop useful dry strengths.

#### Green Bond Strength in Molding Sands

Electron photomicrographs⁴ show that kaolinite particles are thin hexagonal plates. They are formed by fracturing parent monoclinic crystals along crystallographic planes parallel and normal to the c-axis. Broken valence bonds are located at the free faces of the kaolinite particles (shown schematically in Fig. 4). The net electrical charge upon a *dry* kaolinite plate is zero since they exist as discrete particles.

During hydration water dissociates and hydroxyl ions are preferentially adsorbed upon free surfaces of clay plates. This clay-water hull is shown schematically in Fig. 5. Upon adsorbing hydroxyl ions (OH-), clay particles become negatively charged.* As such, the clay-water hull attracts positively charged ions (H+) in the surrounding medium. The positively charged counterions together with the clay particle and adsorbed hydroxyl ions make up the double diffuse layer of a hydrated clay plate. The clay particle surrounded by its double diffuse layer is called a clay micelle (illustrated schematically in Fig. 6). The dotted circle of Fig. 6 denotes the theoretical limit of the water hull in which are contained adsorbed hydroxyl ions and most of the positively charged counterions.

Electrostatic attraction forces (in regard to a positive charge source) associated with an isolated clay micelle are attributed to the adsorbed hydroxyl ions, and electrostatic forces of repulsion are due to the presence of positively charged counterions. The magnitudes and spheres of influence of the repulsion

forces depend upon the type of counterions* (Na+, H+, K+, Ca++ and Ba++) in the double diffuse layer. In Fig. 6 the intensities of both electrostatic attraction and repulsion forces are plotted as functions of the distance from the site of the free surface charge (i.e., the adsorbed hydroxyl ion —OH-). The point  $\mathbf{d}_1$  is assumed to be the mean position from which the electrostatic repulsion forces of all counterions in the

^{*}Ions attached to free surfaces of clay particles. During hydration these ions behave much like hydrogen counterions.

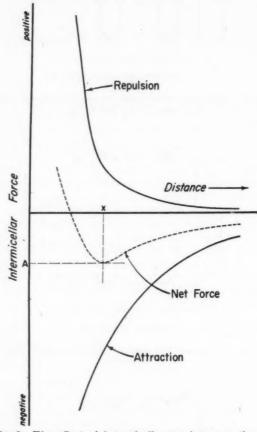


Fig. 8—The effect of intermicellar spacing upon the attraction and repulsion forces between two clay micelles (dipole).

^{*}This is due to an excess of negative charges in the micelle (Fig. 5). Electrodialysis experiments show hydrated clay plates move from the vicinity of the cathode through the free medium, water, to the anode; consequently, they must be negatively charged.

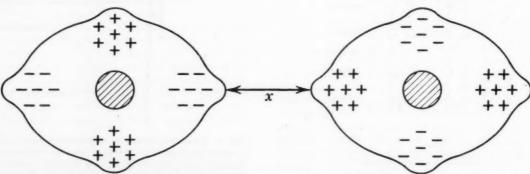


Fig. 7—Sketch schematically showing the localized concentrations of hydroxyl and positively charged counterions

I conin clay micelles obtained in dipole formation. The diserions tance x is the theoretical equilibrium distance between
dipole elements.

double diffuse layer are considered to act. From the curves for sodium-kaolinite in a pure water medium (Fig. 6) the intensities of both attraction and repulsion forces decrease progressively in magnitude with distance from the site of the surface charge. The intensity of the repulsion forces decreases more rapidly with distance than does the intensity of the attraction forces.

Repulsion forces are dominating to the left of point  $d_2$ . The net force is zero at point P, which has been computed to lie a short distance outside the sodium-kaolinite micelle. If hydrated sodium-kaolinite particles are permitted to align themselves in a water medium, they will approach each other until a certain minimum intermicellar distance is attained. The minimum intermicellar spacing is defined as the distance between the point P and the edge of a neighboring micelle. This is true if intermicellar attraction (bonding) depends solely upon the electrostatic forces associated with the adsorbed hydroxyl ions, and positively charged counterions in the outer layers of the micelles.

The electrostatic forces associated with hydrated clay particles are relatively small and their spheres of influence extend only short distances outside the micelles.⁶ These forces do not account for the total bond realized between clay micelles. Hydrated clay particles are drawn together over distances that extend well beyond the spheres of influence of the elec-

trostatic forces associated with the micelles. Obviously, other forces must exist that exert a greater influence upon neighboring particles than do electrostatic forces.

As shown schematically in Fig. 6, the double diffuse outer layer of the clay micelle contains both hydroxyl ions and positively charged counterions. When a number of hydrated clay particles exist in an aqueous medium, as they do in molding sand, negative hydroxyl ions in the double diffuse layer of one micelle exhibit an attraction for clay centers of neighboring micelles, and for their positively charged counterions. Hydrated clay particles also tend to repel one another; this is due to the mutual repulsion of like ions in double diffuse layer and clay centers of neighboring micelles.

Attraction and repulsion forces between neighboring micelles effect localized concentrations of adsorbed hydroxyl and positively charged counterions (Fig. 7), and the clay micelles (dipoles) tend to be drawn together. The magnitudes of attraction and repulsion forces, and the average intermicellar force of hydrated clay dipoles are presented as functions of intermicellar distance in Fig. 8; the curves show the net intermicellar force is one of attraction, which increases in magnitude as the intermicellar distance decreases. Maximum average attraction force (A), i.e., the bond between hydrated clay particles, is realized when micellar dipoles are a critical distance

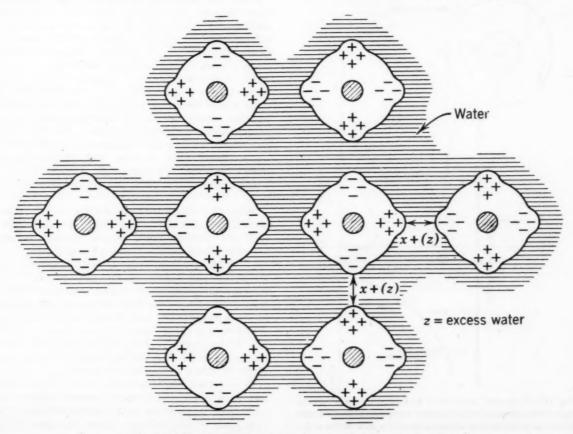


Fig. 9-Schematic illustration of the clay dipoles as in green sand.

(x) apart.⁶ The structure of the hydrated clay clusters about sand grains may be described as networks of micellar dipoles which are retained at finite distances from one another by strong intermicellar forces. The resultant disposition of clay dipoles (a state of minimum energy) is illustrated schematically in Fig. 9. In the moist (green) condition, intermicellar space

ings are filled with water.7

Viscosity measurements of several clay-water systems of equal concentration indicate the cohesion (bond) between hydrogen-kaolinite micelles is strongest of all, the bond between hydrated sodium-kaolinite particles, weakest, and calcium-kaolinite micelles develop bonds of intermediate strengths.⁵ Sand test data show molding sands bonded with calcium bentonite develop greater green strength than sodium bentonite sands. Apparently, the intensity of bond between clay micelles is materially affected by the kind of metallic ion adsorbed in the free surfaces of clay particles. A comparison of sodium- and hydrogen-kaolinite micelles (Figs. 6 and 10) shows small ionic size and low degree of ionic hydration favor thin double diffuse layers, and hence low repulsion forces.6 Assuming that attraction forces for a particular clay micelle are unaffected by the kind of ions adsorbed on surfaces of clay particles, those ions effecting low repulsion forces favor a greater bond be-

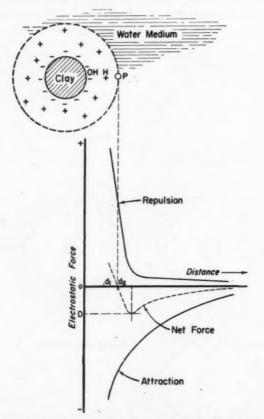


Fig. 10—Schematic sketch of a hydrogen-kaolinite micelle in a water medium indicating the variation in electrostatic forces as a function of distance from the site of the negative surface potential (Johnson and Norton⁵).

tween clay micelles. The average intermicellar forces are presented in Fig. 11 as functions of intermicellar spacing for hydrogen, calcium and sodium clays; maxima of average intermicellar force curves represent maximum bonding strengths of the three clays.

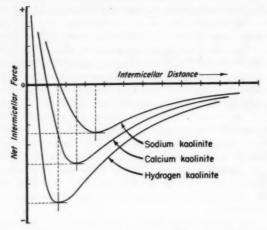


Fig. 11—Schematic presentation of the net intermicellar force as a function of intermicellar spacing for hydrogen-, calcium- and sodium-kaolinite micelles. The minima of these curves indicates the magnitude of intermicellar bond strength.

Illite- and montmorillonite-bearing clays form claywater micelles in much the same way as kaolinite. Sodium-montmorillonite, the active clay mineral in sodium-bentonite, forms complex micelles; owing to its loosely-bonded crystal structure, sodium-montmorillonite crystals can absorb from one to four layers of water molecules between adjacent silicon-oxygen sheets (layers A and B, Fig. 1). The absorption of water causes the lattice to expand. Just how water is absorbed within the sodium montmorillonite crystal has not been determined conclusively. Bradley, Grim and Clark⁸ consider the sodium montmorillonite crystal to absorb water in steps. The initial expansion occurs when the first continuous sheet of water dipoles, one molecule thick, penetrates the crystal lattice.

Additional expansions are realized as the second, third and fourth continuous sheets of water dipoles, each one molecule thick, assume position between adjacent silicon-oxygen layers (A and B, Fig. 1) in the sodium montmorillonite crystal. Magdefrau and Hofmann⁹ and Nagleschmidt,¹⁰ on the other hand, consider the swelling of sodium montmorillonite a continuous rather than step-like process; the swelling increases progressively and continuously as more water penetrates the crystal lattice between layers A and B, Fig. 1. Maximum swelling results when the silicon-oxygen sheets have become separated a distance equal to the thickness of four water molecules.

As a sodium-montmorillonite crystal absorbs more and more water, the distance between adjacent silicon-oxygen layers in the crystal lattice increases. The magnitude of the attraction forces decreases according to Coulomb's equation,

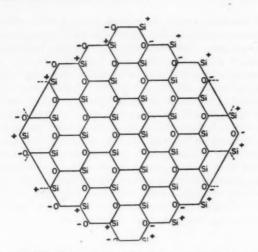


Fig. 12—Two-dimensional network of a quartz particle. Silicon atoms located on the free surfaces of a quartz particle are positively charged and oxygen atoms, negatively charged.

$$F = k \frac{q_1 \cdot q_2}{r^2},$$

where F is the magnitude of the attraction force; k, a proportionality constant; q1 and q2, the intensities of the net electrostatic potentials on each siliconoxygen layer, and r, the distance between adjacent silicon-oxygen sheets in the lattice. It is apparent that the shearing forces required to distort a sodiummontmorillonite micelle decrease with increasing amounts of water absorbed in the micelle. The decrease in stress required to deform a sodium-montmorillonite micelle is of practical importance. When sodium bentonite, silica sand and water are mixed together in a sand mill, the shear forces developed by the action of the mulling wheels cause "swollen" sodium-montmorillonite micelles to become elongated and so effect a more nearly complete coverage of sand grains.4

Shop operations have proved hydrated bentonites are three to five times stronger than either illite or kaolinite clays. The bonding efficiency of clay is largely a function of its surface area, since micelle formation is a surface reaction between clay particles, and hydroxyl ions and positively charged counterions. On a unit weight basis, greater micelle formation is realized with clays of finer particle size (greater surface area). As shown in Table 1, bentonites are appreciably smaller in particle size and provide greater surface area per unit weight than other clays commonly used in the foundry.

Table 14—Average Particle Size in Millimicrons of Several Common Clay Minerals

Clay	Thickness	Width
Sodium-montmorillonite	1	100-300
Calcium-montmorillonite	1	100-300
Kaolinite	20	100-250
Illite	20	100-250

Heretofore, the discussion has been limited to development of bond between hydrated clay particles. Bond is also realized between hydrated clay particles and quartz grains. A quartz grain consists of a continuous network of silicon and oxygen atoms as indicated schematically in Fig. 12. There are unsatisfied positive and negative valence bonds at free surfaces of quartz crystals. In aqueous media, quartz, like clay particles, preferentially adsorb hydroxyl ions, and the subsequent attraction of positive counterions forms double diffuse layers about quartz centers.

The net electrostatic charge associated with a quartz particle is negative;* just outside the limits of an isolated micelle, the net electrostatic force, although very weak, is one of attraction. The net intermicellar force between hydrated quartz particles is also very small as shown in Fig. 13; consequently, the overall bond (net electrostatic and net intermicellar forces) is very small. When a mixture of clay and sand is hydrated, micelles of clay and quartz are formed. The bond is due in part to satisfaction of the net electrostatic attractive forces between clay and quartz micelles and in part to the net intermicellar forces between clay and quartz dipoles. The bond between clay and quartz micellar dipoles is equivalent to the net intermicellar force (B)* in Fig. 13 and the critical intermicellar spacing is equal to distance (y).

*Hydrated quartz particles are found to gather at the anode of an electrodialysis cell.

 The net intermicellar force is equivalent to the graphical difference between the attraction and repulsion forces of clay and quartz micelles.

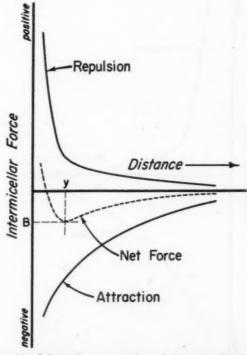


Fig. 13—Schematic presentation of intermicellar forces as a function of the space between hydrated quartz particles.

## Dry Strength in Molding Sand

Green molding sand consists of clay and quartz micelles firmly held a finite distance apart; the intermicellar voids are filled with extraneous water. In this condition, clay and quartz micelles (dipoles) orient themselves to form an aggregate of minimum free energy (Fig. 15). The magnitudes of bonds between hydrated clay particles, as well as between clay and quartz micelles as a function of intermicellar spacing, are obtained from the average intermicellar force curves in Figs. 8 and 13.

The intermicellar spacing in green molding sand is considerably greater than the critical intermicellar spacing (x) and (y) in Figs. 8 and 13. As a molding sand dries in air, it shrinks in volume, and clay and quartz dipoles move closer together. The decrease in volume is due solely to evaporation of extraneous intermicellar water. Decreases in intermicellar spacing, as indicated in Figs. 8 and 13, effect increases in bond (average intermicellar force).

Maximum bond is approached as the molding sand attains its minimum volume; at this stage of drying, micelles have attained their critical intermicellar spacing and bonds between hydrated clay particles and between clay and quartz micelles are equal to the maxima (A and B) of the average intermicellar force curves in Figs. 8 and 13. Thereafter, additional drying of molding sand evolves intermicellar water; however, no additional increase in bond is realized since hydrated clay and quartz particles have previously attained their critical intermicellar spacing.

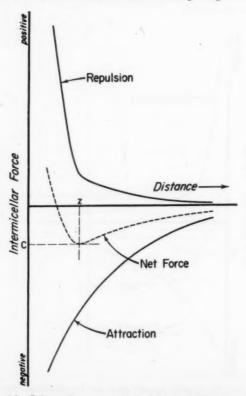


Fig. 14—Schematic presentation of intermicellar forces as a function of the space between clay and quartz micelles.

Laboratory results show sands bonded with sodium bentonite develop dry bond strengths considerably greater than those bonded with calcium bentonite, illite or kaolinite clays. The marked difference in bonding ability is due to the structure of the hydrated clay networks about sand grains. Unlike other clay minerals, sodium montmorillonite, the active bonding agent in sodium bentonite, absorbs water within its crystal lattice and swells; as mentioned above, hydrated sodium bentonite can be deformed into long, thread-like plates. When sodium bentonite bonded sands were packed into a dense mass, the distorted clay plates, bonded to one another, form a continuous network throughout the molding sand. On drying, this strongly knit and continuous network is capable of resisting relatively high shear stresses.

#### Conclusions

During hydration of molding sands, double diffuse layers are formed by the adsorption of hydroxyl ions and positively charged counterions on free faces of clay and quartz particles; hydrated particles are called micelles. Green molding sand consists of hydrated clay and quartz particles (dipoles) firmly held at fixed distances apart in space; intermicellar voids are filled with extraneous water and the space between micelles depends upon the amount of free water. The bond between hydrated clay particles and between clay and quartz micelles is due in part to the net electrostatic forces between hydrated particles, and in part to the net intermicellar forces between micellar dipoles.

The types of ions (H+, Na+ and Ca++) absorbed on the surfaces of clay particles have a marked effect on the magnitude of bond between hydrated clay particles as indicated by viscosity measurements of various clay-water systems. Clay micelles whose surface-adsorbed ions are small in size and hydrate slightly develop relatively great bond strength. This is due to their forming thin double diffuse layers which favor low repulsion forces. The greater green bond obtained in molding sands bonded with calcium bentonite is attributed, in part, to this phenomenon.

Kaolinite, illite, and sodium and calcium montmorillonite clays behave very much alike in aqueous media; the net electrostatic and net intermicellar (dipolar) forces effect bonding between hydrated clay particles. Of these clays, sodium montmorillonite is unique for it absorbs water within its crystal lattice and swells. In this state, it is possible to distort sodium montmorillonite micelles, which form a continuous network throughout the molding sand. On this basis, sodium montmorillonite micelles should develop greater bond in green molding sand; but owing to the relatively low shear stress required to deform the swollen sodium montmorillonite lattice, calcium montmorillonite micelles are more effective in developing green bond. On drying, the continuous sodiummontmorillonite micellar network throughout the molding sand is strongly-knit and resistant to high shear stresses; it is due to this network that sodium bentonite develops stronger dry bond than the clays commonly used in the foundry.

Surface area markedly influences the magnitude of

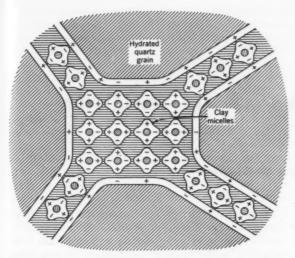


Fig. 15—Schematic illustration of disposition of hydrated quartz grains and clay micelles in green molding sand.

micellar and intermicellar forces; clays of large surface area (fine particle size) common to the bentonites are most effective as binders in molding sand.

The evaporation of extraneous intermicellar water in green sand effects a decrease in volume. The distance between micellar dipoles decreases and the average intermicellar force or bond between hydrated particles increases. Maximum bond or dry strength is attained when hydrated particles are a critical intermicellar distance apart. Additional drying does not effect an increase in bond even though more intermicellar water is evolved.

#### Acknowledgment

The authors are indebted to George E. Schmidt, Jr., Research Assistant, Metallurgy Dept., M.I.T., for preparing the illustrations used in this paper, and to others of our foundry colleagues for assistance of various sorts.

#### References

1. Hofman, Endell, and Wilm, "Crystal Structure and the Swelling of Montmorillonite," Zeitschrift fur Kristallographie, vol. 86, 1933, pp. 340-348.

- Gruner, "Densities and Structural Relationships of Kaolinite and Anauxites," American Mineralogist, vol. 22, 1937, pp. 850-860.
- 3. Grim, Bray, and Bradley, "The Mica in Argillaceous Sediments," American Mineralogist, vol. 22, pp. 813-829.
- 4. Endell, Reininger, Jensch and Csaki, "Ueber die Bedeutung der Quellfahigkeit toniger Bindemetall für Giessereisande," Die Giesserei, vol. 24, 20 November 1940, p. 465.
- Johnson and Norton, "Fundamental Study of Clay (II), Mechanism of Deflocculation in the Clay-Water System," Journal, American Ceramic Society, vol. 24, 1941, pp. 189-203.
- 6. Hauser, Colloidal Phenomena, McGraw-Hill Co., New York, 1939.
- Norton, "Some Notes on the Nature of Clay, Part II," Journal, American Ceramic Society, vol. 16, 1933, p. 86.
- 8. Bradley, Grim, and Clark, "A Study of the Behavior of Montmorillonite upon Wetting," Zeitschrift für Kristallographie, vol. 96, 1937, pp. 216-222.
- 9. Magdefrau and Hofmann, "Die Kristallstruktur des Montmorillonits," Zeitschrift fur Kristallographie, vol. 98, p. 299.
- Nagelschmidt, "On the Lattice Shrinkage and Structure of Montmorillonite," Zeitschrift fur Kristallographie, vol. 93, pp. 481-487.

## DISCUSSION

Chairman: C. C. Sigerfoos, Michigan State College, E. Lansing, Mich.

Co-Chairman: V. M. Rowell, Archer-Daniels-Midland Co., Cleveland.

Recorder: F. P. GOETTMAN, Standard Sand Co., Grand Haven, Mich.

O. J. MYERS: 1 How much of the bonding action is due to surface tension and how much is attributed to micellular forces?

MR. TAYLOR: This is impossible to determine. However, electrostatic forces are not considered to be the main bonding forces. Interparticle forces and surface tension are basic factors and additional research is required to further evaluate these factors.

D. C. WILLIAMS:² A paper recently published in France gives some indications that the double layer theory is not a true picture. Are you familiar with this work?

MR. TAYLOR: I am not familiar with this paper. However, it makes no difference as something is present. The present double layer theory has been the basis of a great progress in ceramic research. Thus if the double layer is not absolutely correct, we do know that interparticle forces and surface tension are still operating and are important. It is the purpose of this paper to direct attention to these fundamental forces as a foundation for bonding action.

J. R. SMALL:³ Has all this theory any practical application? MR. TAYLOR: Yes. With the basic knowledge obtained from a study of these fundamental factors one can adjust the pH of the water to the type of clay you use or add chemicals to develop desired properties.

¹Archer-Daniels-Midland Co., Minneapolis, Minn. ²Ohio State University, Columbus, Ohio. ²U. S. Naval Ordnance Test Station, China Lake, Calif.

## THE FOUNDRYMAN LOOKS AT AIR POLLUTION

By

N. H. Keyser* and H. P. Munger*

#### ABSTRACT

The air-borne contaminants from a foundry are classified as dust, fumes, and noxious gases. They originate at many points in the foundry. Particle size is the most important property of solid contaminants. Each foundry finds itself in a different situation because the nature of the operation differs from foundry to foundry and because location of the foundry is of great importance. The foundryman and the local air-pollution control authorities should cooperate in working out an individual program for collecting dusts and fumes, employing the knowledge and research methods at hand.

The technical methods for controlling air pollution are fairly well developed. These include dry inertial collectors, and wet scrubbers which have a wide range of efficiencies and also a wide range of costs. Filtering methods and electrostatic precipitators are among the most efficient dust collectors. Several newer methods are going through the production development stages.

In spite of a number of factors which partially offset the cost of air-pollution control, cleaning air rarely yields a net profit. The actual cost will depend largely on the amount of cleaning required by the local conditions. The public will eventually pay this cost

#### The Problem

The foundryman has a definite interest and a responsibility in preventing excessive discharge of waste into the atmosphere. But how much waste can be discharged into the air before we say the discharge is excessive or that the air is "polluted"? A few decades ago, dense clouds of smoke coming from the stacks of our industrial plants were regarded as a sign of prosperity, and nothing short of a dense cloud of noxious fumes would have been described as air pollution. On the other hand, refreshing country air contains wind-borne dust, pollen, and gasses from decaying vegetation.

The present trend is to define some concentration of contaminants between these two extremes as excessive. Air is regarded as polluted when the concentration of contaminants exceeds that which is injurious to property, to health, or to human comfort. For the most part, therefore, air pollution is of greatest concern in large centers of industrial activity. Overloading of the air with dirt may result from

numerous small sources of contaminants over a large area, or a single concentrated source, or any combination of these two conditions. Smoke and fly ash from industrial stacks, railroad locomotives, or the chimneys of homes and schools are the most obvious sources of contaminants. Burning trash and leaves in metropolitan areas, gob piles in coal-mining areas, and the exhaust from Diesel locomotives and trucks or even passenger automobiles also contribute their share of contaminants. It can be seen that the foundry, in most cases, is just one of the many sources of air pollution in a community.

Aside from the contribution of the foundry to the over-all burden of contaminants in an industrial area, the foundry, as an individual industrial plant, has a problem of community relations. When the neighbors see nothing but smoke pouring from the foundry, they get the impression that the foundry is a smoky, dirty place. They are inclined to blame their dirty washing on the smoke that they can see, even though the actual source of the dirt may be some distance away or perhaps just the general condition of that particular area.

One of the problems facing the foundry is that of attracting young men to work in that industry. Certainly a reputation for smoke and dirt does not help the foundry solve this problem.

## Each Foundry is an Individual Control Problem

To those who have studied air-pollution problems, it is evident that each plant presents an individual problem. The methods for reducing contaminants must be tailored to fit this individual situation. The municipal and county programs, in each case, form the backbone of air-pollution control. Each company within a local air-pollution control district should work out its own program in cooperation with the local air-pollution control engineer. Any program of air-pollution control should make the widest use of the knowledge that is available.

The first information required in setting up an individual program for the control of air pollution is a knowledge of the physical and chemical nature of the contaminants emitted into the air, especially

^{*}Battelle Memorial Institute, Columbus, Ohio.

their particle-size distribution. The concentration of the contaminants and the points at which they leave the plant also need to be known. Other important factors are the natural surroundings of the plant and locations of population centers and farms. Of equal importance are the atmospheric conditions under which the contaminants are found objectionable.

In the dispersion of air-borne contaminants, the nature of the country and the atmospheric conditions play major roles. Flat country, especially near the Great Lakes where strong winds are frequent, favors the rapid dispersal of dust, fumes, and noxious gases. Narrow, deep valleys furnish pockets in which contaminants may collect to form relatively high concentrations. Temperature inversions which form over pockets in the land stabilize the air and prevent the normal vertical dispersion of contaminants. A temperature inversion is a condition in which the air at ground level is colder than the air above it.

Los Angeles county is an outstanding example of the influence of the nature of the land surface and atmospheric conditions on the buildup of contaminants over an industrial area. In Los Angeles county, a large concentration of industry is situated in a basin surrounded on three sides by mountains. Temperature inversions, in effect, put a lid on this basin. Because they cannot get out of the basin, smoke and fumes and noxious gases frequently build up for several days. Fogs, rolling in from the ocean, combine with the concentration of man-made contaminants to create a powerful eye-irritating smog. It is because of this set of conditions, which are abnormal for most of the rest of the country, that the air-pollution standards for Los Angeles must be so rigid.

#### Devices for Analyzing Air-Borne Contaminants

There are well-established technical methods for determining what are the air-pollution problems of any individual industrial plant. These methods utilize devices for collecting and analysing the air-borne contaminants. Analysis of an individual foundry's contribution to air pollution can be made right at the source. For example, the amount and nature of emissions from a cupola can be determined. Measurements of dirt falling on the surrounding community also can be made. A new device, developed at Battelle, not only determines the amount of dirt fall but also establishes the direction from which dirt is coming.1 A knowledge of the direction from which the dirt is coming is of special interest in determining whether the contaminants are of local origin or have been blown in from another area or community.

Devices for measuring the influence of the surface of the land and the atmospheric conditions on the dispersion of contaminants are also in use. For studies just above ground level, instruments have been attached to fixed structures. A new technique uses a kite balloon to carry weather instruments and airsampling devices up to altitudes of 500 ft or higher. 1.2.3 In highly industrialized areas, these methods provide the tools for securing a three-dimensional picture of contamination of the air.

#### The Nature of Air-Borne Contaminants

An effective program for controlling air pollution from the foundry must be based on a knowledge of the origin and nature of the contaminants that are emitted into the air. It is customary to divide the contaminants into four principal classes: (1) smoke, (2) dusts, (3) fumes, and (4) noxious gases. Smoke is defined as the solid or liquid particles resulting from incomplete combustion of organic materials. Soot is an example. Dusts are solid particles of any nature which have been produced by mechanical processes and released into the air. Particles of iron and abrasive from the grinding of castings are typical dusts found in the foundry. Fumes are solid or liquid particles in the air, commonly generated by sublimation, vaporization, or chemical reactions. The white fumes liberated during the baking of cores, and the brownish fumes which arise from hot iron or steel are typical examples of fumes in the foundry. Sulphur dioxide and carbon monoxide are typical noxious gases with which the foundryman is familiar.

Probably the most important property of contaminants is their particle size. Particle size coupled with density and shape determines the distance contaminants will travel in air before they settle to the ground. Visibility of contaminants is determined by the number and size of the particles in the air and by the degree to which moisture condenses on the particles to form fog. Finally, the particle size of the contaminants determines what methods of sampling and analysis can be used for studying the air and what type of equipment can be used for separating the contaminants from air, and for their subsequent disposal.

In view of the importance of particle size in airpollution control, the foundryman should have an idea of the size of particles involved. The size of smoke, dust, and fume particles is commonly expressed in terms of the average diameter of the particles measured in microns. A micron is 1/1000 of a millimeter, or 1/25,400 of an inch. The size of particles in the air may range from over 1000 microns for raindrops to considerably less than 0.01 micron for sulphur dioxide and other gases. Several other examples which are useful in visualizing the size of the particles are: (1) a 50-mesh sand has an average diameter of 295 microns, (2) a 325-mesh silt has a diameter of 44 microns, (3) the smallest particle visible to the naked eye is 10 microns in diameter, and (4) the particles in tobacco smoke have an average of about 0.3 mic-

Particles below 1 micron in diameter tend to stay suspended in quiet out-door air almost indefinitely, whereas those larger than 1 micron tend to settle at a more or less rapid rate. In agitated air, such as found in foundries and other industrial plants, particles 10 microns or larger in diameter may stay suspended in the air for a considerable length of time.

The range of particle size for a number of industrial dusts and fumes is shown by means of bars in Fig. 1. Foundry dusts include such coarse particles as the cinders and other debris blown out the stack of a cupola and the finer dusts from the foundry

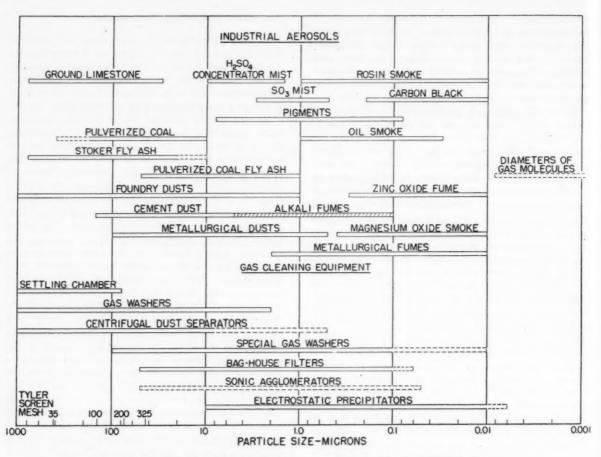


Fig. 1 -Particle Size of Industrial Aerosols and Equipment for Cleaning Gases

shake-outs and cleaning room. Metallurgical fumes comprise such contaminants as the brownish fumes coming off the electric or open hearth, and that part of the cupola and the non-ferrous-melting furnace emissions resulting from volatilization and oxidation of the metallic constituents of the charge.

It is apparent that the foundryman must contend with the entire range of particle sizes considered to be air contaminants. For the sake of economy in controlling the air contaminants, the foundryman should use the type of collection equipment that will do each job at the lowest cost.

The particle-size ranges for which broad classes of collecting equipment are effective are shown in the lower half of Fig. 1. Additional details on collecting equipment are given in a later section.

## Source of Contaminants

Air-borne contaminants come from many points in the foundry. In many cases, the melting department is a major source. In fact, the melting department has been the first to attract the attention of air-pollution control authorities in many cities. The electric furnace and open hearth for steel and iron melting emit clouds of reddish-brown fumes, especially during melt-down when the charge is not protected by the slag, and then again at the end of the heat when the metal is at its maximum temperature. The advent of the oxygen boil in steelmaking has aggravated the fume problem.

The open-top cupola belches forth a stream of hot gases clouded with solid particles. The coarsest particles are the size of coarse sand. The finest particles must be viewed with an electron microscope to distinguish them. Coarse particles settle on the foundry roofs, yards, and on the foundry's immediate neighbors. The finer particles drift away to settle at some distance from the foundry. Sulphur dioxide, formed from sulphur in the coke, and unburned fumes from oily scrap add their characteristic unpleasant odors.

A typical sample of the cupola stack-gas emissions will contain on the order of 35 per cent of its particles in the zero to 5-micron range and about 50 per cent of the particles will be over 44 microns (325 mesh). The remaining 15 per cent will be spread out over the range of 5 to 44 microns. The concentration of particles above 44 microns (325 mesh) is chiefly dust and debris consisting of broken coke, fine metal chips, rust, sand, and dirt, which are picked up by the rising gas stream and carried out the stack. The particles in the 0 to 5-micron range are chiefly fumes from the volatilization and oxidation of iron, silicon, manganese, and certain tramp elements by the intense heat of the melting zone. These fumes are

largely responsible for the opacity or color of the stack gases, primarily because there are so many particles in a given weight of fume as compared to a given weight of the coarse particles. It takes 1000 particles, I micron in diameter, to weigh as much as one particle 10 microns in diameter.

The core room also is a major contributor of fumes in the atmosphere in and around a foundry. Fumes given off from core oil and cereal binders during the baking of cores are pungent and irritating. A host of obnoxious compounds have been identified in the fumes given off by linseed oil during drying. These include acrolein, formic acid, acetic acid, acrylic acid, butyric acid, carbon monoxide, and a number of other volatile products.⁵ Presumably, many of these same compounds are generated during the baking of cores and also during casting and shake-out of molds containing cores.

No entirely satisfactory substitute for core oil has been developed which does not give off fumes and odors. In some cases, the odors of the substitutes were more obnoxious to many foundrymen than the odors from core oil.

The shake-out and cleaning room are also sources of large quantities of fumes, dusts, and noxious gases. In the shake-out, large quantities of hot sand are suddenly exposed to air. Particularly where many cores are used, large quantities of fumes and noxious gases are generated by the incomplete combustion of organic materials in the mold. Added to this burden are fine sand and binder materials thrown in the air by mechanical agitation used to shake the molding material off the casting. In the cleaning room, the problem is mostly one of dust generated by such operations as chipping, blasting, and grinding, although metal and other fumes do arise from welding operations.

Miscellaneous sources of air contamination in the foundry are open fires, handling of dry sand before it is tempered, parting agents, metal fumes from the ladle during transfer and pouring, and mold gases. The foundry pattern shop also has its dust-control problem. In this case, however, the materials involved are woodworking shavings, chips, sawdust, and, if metal patterns are used, there is the possibility of metal and abrasive dust.

## Methods for Control

The industrial hygienists have developed excellent methods for reducing the concentration of contaminants in the air within the foundry. They have approached the problem from several angles, among which are (1) to prevent or to reduce the production of dust, fumes, or noxious gases, (2) to reduce the use of materials which give rise to dust, fumes, and noxious gases, (3) to provide personal safety equipment to prevent inhalation of dust, fumes, or noxious gases, and (4) to extract the dust and fumes from the various points in the plant where the contaminants are generated.

The extracted dust- and fume-laden air may be exhausted directly into the atmosphere outside the

foundry. If the burden of dust and fumes of this air is great, the air must be passed through an air cleaner to avoid air-pollution problems. The foundryman is then faced with making a decision as to what type of dust collector to use.

Control of air pollution starts with the dust hood and ducts to convey the dirt-laden air to the collectors. Much good engineering has gone into dust hoods and ducts. Since the principles are well known, the subject will not be discussed here except to mention one of the problems encountered in the foundries that have applied dust control to existing plants. In most instances, space is at a premium, and dust-control equipment must be built into extremely limited space. This leads to a temptation to use hoods, piping, and collectors of inadequate size.

#### Available Types of Dust Collectors

The technical methods for cleaning dirty air and gases discharged into the atmosphere are fairly well developed. The proper selection and application of cleaning equipment depends on a knowledge of the types of collectors available and their performance. Dry inertial collection methods are the most economical but have definite limitations. These methods depend on gravity or the momentum of a particle to carry it out of the gas stream. In its simplest form, the dry collector is merely a settling chamber or an enlargement in the duct carrying the dust-laden gas. Practical limitations on the size of such chambers do not permit designs with low enough gas velocities to extract fine particles. The minimum particle size normally collectible in a settling chamber is on the order of 40 to 80 microns, depending on the design.

The cyclone collector is a widely used collector of the inertial type. By introducing the dust-laden gas tangentially into a circular chamber, a centrifugal motion is created which throws the dust particles to the side where they deposit, or drop to the cleanout in the bottom,

Decreasing the diameter of the cyclone makes the gas travel in a tighter circle and increases the centrifugal force for throwing the dust out of suspension. The small-diameter cyclones, therefore, are more efficient where particles in the smaller size range are encountered. On the other hand, because of rebound, of coarse particles, the collector may not be as efficient for removing particles of large size. Multiple units of small-diameter cyclones are required to handle a large volume of gas.

A large variety of makes and configurations of cyclone collectors is commercially available. In general, however, the collecting efficiency drops off rapidly for particles below about 20 microns in diameter for the large-diameter cyclone and below 10 microns in diameter for the small-diameter multiple cyclones. In addition, the collecting efficiency is practically nil for particles below 5 microns for simple cyclones and 2 microns for multiple cyclones. This, of course, limits the type of dust that may be collected. Efficiencies in the neighborhood of 97 per cent have been reported on coarse material, such as saw dust from planing mills.⁶ In contrast, they may

collect only between 10 and 25 per cent of the solid material in stack gases from a boiler.⁷

In many cases, where it is necessary to install additional exhaust fans, or where space is limited, a mechanical centrifugal collector will be found practical. This type of collector uses a high-speed impeller to provide the centrifugal force for throwing the dust out of suspension. The impeller also acts as a fan for drawing the dust-laden gas through the cleaner. At the point of greatest concentration of dust, part of the air is skimmed off and carried to a settling chamber. After depositing the entrained dust, the air is recirculated in the system.

Louver and dry skimmer collectors also depend on inertia for throwing the dust out of suspension. Inertia concentrates the particles of dust in the small end of a slotted metal cone. About 90 per cent of the air passes through these slots to the blower and stack, while the concentrated suspension passes into a small cyclone. The exhaust from the cyclone is recycled in the concentrator.

#### Centrifugal Separators and Louver Collectors

Both mechanical centrifugal separators and louver collectors have application for collecting particles in the size range collected by cyclone separators. Dry inertial collectors, as a class, are not suitable for collecting fine dust (particularly particles below 10 microns in diameter). They are satisfactory, however, as rough cleaning devices for reducing the load on high-efficiency final collectors with which they are frequently used. On the other hand, as final cleaners for coarse, non-toxic materials, such as wood or metal chips, they do an excellent job and are simple and inexpensive. These collectors can be built to handle relatively hot gases which makes them eminently suited for such applications in cupolas where merely the removal of the coarses particles from the stack gas is sufficient to satisfy local conditions.

When dry inertial collectors are inadequate, it is advisable frequently to go to one of the wet collecting methods. Wet collecting methods depend on a simple process of washing or scrubbing the dirt out of the gas and wetting the particles to increase their effective diameter so they may be removed by simple collectors, such as a cyclone. In addition to removing dust, wet collectors also can remove sulphur gases which may sometimes be important.

## Static Spray Washers

The simplest form of wet cleaner is the static spray washer which may be seen mounted on the top of many cupolas. This equipment can remove 50 to 75 per cent of the solid matter and a portion of the SO₂ in the cupola effluents. However, a large portion of very fine solid particles escape.

Wet scrubbers are more efficient in removing the smaller size particles than are the simple spray washers. Wet scrubbers depend on mechanical means of admixing a spray of water with the dust-laden gas. They employ the cyclone, rotary centrifugal collector, or packed tower to remove the wetted dust particles.

Roughly speaking, the efficiency of these scrubbers is related to the work applied to admixing the water and gas. Scrubbers can be very efficient and give very clean gas. Because of the high power requirements for the highly efficient scrubbers, it is customary where very clean gas is required, to use a scrubber of moderate efficiency, followed by other types of final cleaners which require less power. The wet cleaner reduces the load and, at the same time, cools the gas.

Cloth filters or bag houses are becoming a more familiar sight around the foundry. They operate on the same principle as the vacuum cleaner in the home. They have been applied to cleaning the exhaust from the shake-out and cleaning rooms. In similar applications, they have been found to operate with dust-collecting efficiencies of over 99 per cent.6 It has been only recently, with the development of the glass-cloth bag which will withstand temperatures up to about 500 F, that bag houses have been utilized for the difficult job of cleaning the hot gases from the cupola. They have been found to satisfy the difficult requirements of the Los Angeles County Air Pollution Control District.8 All material passing through the cloth filter is below 5 microns. Efficiencies are good on particles down to a fraction of a micron. Bag houses are limited, however, to applications where there are no condensable fumes or gummy materials present.

#### Electrostatic Precipitators

Electrostatic precipitators are probably the most efficient of all collectors. Collecting efficiencies are very high on particles down to a fraction of a micron. Although electrostatic methods do not remove gases, there seems to be no lower limit to the size of solid particles that they will remove.

Electrostatic precipitators employ up to 75,000 volts to induce a negative electrostatic charge on the surface of the dust particles. The dust particles are then repelled to the positive electrode, which is constructed to serve as a dust catcher.

Among the newer methods for collecting dust and smoke are the sonic agglomerators. 9,10 Sound waves cause the small particles of dust to vibrate, thereby increasing the number of collisions between particles. Upon collision, surface forces on the particles of certain types of materials cause them to adhere to one another. Through a series of collisions and adhesions, small particles agglomerate to such size that they can be removed from the gas with a cyclone separator. The method is limited to very small particles, as are found in smoke, and it does not work on all materials. Further developmental work is required before its possibilities in the foundry are known.

One of the latest improvements in wet collecting equipment is the Venturi scrubber. In this scrubber, water is broken up into a very fine mist. The mist and gas pass through the venturi portion of the collector where the very small droplets wet and absorb the very small particles of dust. A cyclone separator then collects the wetted particles and drops them into a tank for disposal.

#### Control of Combustible Fumes

Combustible fumes can be destroyed by oxidation or combustion. Where the ratio of combustible material to air is high enough, as in the cupola stack gases, combustion will be self-sustaining once it has been started. An ignitor is sometimes used for igniting the gases during periods when they do not ignite freely. Lean mixtures, which will not maintain combustion, must be heated to temperatures on the order of 1200 to 1400 F, to complete the oxidation. In some instances, fumes from the baking of cores have been reduced by recirculating the air in the oven through the burners.

One of the newer methods for oxidizing combustible fumes is the catalytic combustion process. Very dilute mixtures of combustible vapor or gas will oxidize at temperatures as low as 350 F, or less, in the presence of a catalyst. This principle has been applied commercially to the oxidation of a number of fumes and gases, including the fumes from the baking of cores. Of special interest is the fact that the heat obtained from the oxidation of these fumes is returned to the oven to reduce the amount of outside fuel required to keep the oven hot.

#### The Cost

The purchase cost of dust-collecting equipment ranges from something less than \$100 per 1000 cu ft per min treated for simple cyclone separators, to \$1000 or even more for certain of the very high efficiency collectors, such as bag houses, electrostatic, or disintegrator-type washers. 12,13 Power consumption may be as low as 0.1 kw per 1,000 cu ft per min for simple gravity settling chambers, to as high as 10 kws per 1,000 cu ft per min for high-efficiency scrubbers. These figures are for total power, including electrical energy, water-pumping power, power for necessary auxiliary equipment, and pressure drop through the apparatus.

One thing that must not be overlooked in estimating the cost of dust-control equipment is the actual volume of gas handled. The more gas handled, of course, the larger the size of collecting equipment required. In the design of dust hoods, for example, sufficient excess air must be drawn through the equipment to flush all the dust-laden air from the source. The equipment must clean all this air plus any excess that is taken in. Cupola stack gases are another example where the volume of gases emitted is greater than may seem at first. The volume of gases generated by a cupola may be 2 to 4 times the volume of air blown into the tuyeres. Burning of the gases in the stack and aspiration of air in the charging door may increase this volume to 5 to 7 times that of the tuyere air. Cooling of these gases before they are passed through the dust collectors, of course, reduces considerably the volume of gas that must be handled. Nevertheless, the volume that must be handled is more than the volume of air blown into the tuveres.

Who pays the cost of clean air? The answer to this is simple and one which everybody should know. The cost of clean air is eventually paid by the public. The degree that air pollution is controlled in the future will depend largely upon the amount the public is willing to pay for controls in terms of increase of cost of products.

The materials recovered in the air-cleaning operation seldom have much value. There have been figures published on the large savings in nonferrous smelting operation by collection of the dust from the flue gases. For example, it has been estimated that \$27,000,000 worth of metals have been recovered in the copper industry alone.¹⁴ Unfortunately, in the foundry industry, in general, the products recovered in the dust collectors are not nearly so valuable. For the most part, they may be classified merely as dirt which has to be hauled off to the dump.

Fly ash collected from the cupolas may have some value as a filler for concrete. However, it probably must sell for considerably less than concrete itself. It must be used close to the source and must be available in quantities large enough to be worth handling. Recently it has been found that fly ash makes a capable pipe eliminator of the non-exothermic or insulating type to apply to the surface of feeders on castings. This would permit the foundry to consume its own fly ash to some advantage.

For the most part, the small quantities of salvagable material involved make the cost of handling more than the products might be worth. It seems hardly worthwhile to develop a useful market for them.

Other opportunities to realize some return on the investment in dust collectors are difficult to evaluate in dollars and cents until the equipment is installed. The factors which may partially offset the cost of clean air are: (1) reduced maintenance, for example, less frequent cleaning of the roof, (2) better working conditions and workers' morale, and (3) better relations with the neighbors. The last two factors should help attract better workers into the foundry and reduce labor turnover.

The by-products and other benefits derived from gas cleaning rarely yield a net profit on the entire cost of the gas cleaning operation, but they may help materially to reduce the expense.

#### **Bibliography**

1. H. P. Munger, "The Engineering Approach to Air Pollution," presented at The American Public Health Association, San Francisco, California, November 2, 1951.

 H. P. Munger, "Present Status of Air-Pollution Research," Mechanical Engineering, vol. 73, no. 5, pp. 405–411, May, 1951.

3. H. P. Munger, "Meterological Methods for Studying Air Pollution," presented at the XII International Congress of Pure and Applied Chemistry, New York, September 10-13, 1951.

4. H. P. Munger, "The Spectrum of Particle Size and Its Relation to Air Pollution," presented at The First U.S. Technical Conference on Air Pollution, Washington, D.C., May 3, 1950.
5. W. M. Lord, "Foundry Cores and Moulds; Vapours and

 W. M. Lord, "Foundry Cores and Moulds; Vapours and Fumes Evolved From Organic Bonding Materials," *Iron and Steel* (London), vol. 23, January, 1950, pp. 21–25.
 R. Dennis, G. A. Johnson, M. W. First, and L. Silverman,

 R. Dennis, G. A. Johnson, M. W. First, and L. Silverman, "How Dust Collectors Perform," *Chemical Engineering*, vol. 59, no. 2, pp. 196–197, February, 1952.

7. A. J. Grindle, "Dust, Fume and Smoke Supression," presented to Iron and Steel Engineers, Cleveland, Ohio, September 26, 1950, and reprint by Whiting Corporation.

8. P. Siechert and H. B. Menardi, "Glass Bags Clean California's Air," The Iron Age, vol. 169, no. 4, pp. 78-80, January 24, 1952.

9. H. W. St. Clair, M. J. Sendlove, and E. V. Potter, "Agglomeration of Smoke, Fog, or Dust Particles by Sonic Waves," Industrial and Engineering Chemistry, vol. 41, pp. 2434-2438 (1949), U.S. Bureau of Mines, R. I. 4214.

10. H. W. Danser and E. P. Neuman, "Industrial Sonic Agglomeration and Collection Systems," Industrial and Engineer-

ing Chemistry, vol. 41, pp. 2439-2442 (1949).

11. H. F. Johnstone and M. H. Roberts, "Deposition of Aerosal Particles From Moving Gas Streams," Industrial and Engineering Chemistry, vol. 41, pp. 2417-2423 (1949).

12. C. E. Lapple, "Processes Use Many Collector Types," Chemical Engineering, vol. 58, no. 5, May, 1951, pp. 144-151.

13. O. T. Zimmerman and I. Lavine, "Chemical Engineering

Costs," Chapter 12, Dust Collecting Equipment, p. 209, Published by Industrial Research Service, Dover, N. H. (1950).

14. H. V. Welch, "Fume and Dust Problems in Industry,"

Transactions, AIME, vol. 185, p. 934 (1949).

15. "New Foundry Techniques; Part 2," Canadian Metals, vol. 15, no. 2, February, 1952, p. 27.

## DISCUSSION

Chairman: F. W. Shipley, Caterpillar Tractor Co., Peoria, Ill. Co-Chairman: J. R. Allan, International Harvester Co., Chicago.

Recorder: H. F. Scobie, American Foundryman's Society.

Chicago.

ORVILLE LAABS1: Which type of dust and fume collector is

preferable, the dry or the wet type?

MR. KEYSER: Foundrymen should use whichever type dust and fume collector is necessary to meet local air pollution ordinances. Use the dry type if possible because of its lower maintenance

K. M. Morse2: What data can you give us on the size spec-

trum of cupola emissions, Mr. Keyser?

Mr. KEYSER: About 50 per cent of the particulate matter emitted is dust larger than 44 microns. Approximately 35 per cent is metal oxides 5 microns and smaller. Very little matter between 5 and 44 microns is emitted.

¹Director of Personnel, Universal Foundry Co., Oshkosh, Wis. Industrial Hygienist, U. S. Steel Co., Pittsburgh.

# ALUMINUM SAND CASTING DEFECTS THEIR IDENTIFICATION, CAUSES AND CORRECTIONS

By

Donald L. LaVelle *

#### ABSTRACT

The first consideration in attempting to improve the quality of castings or to correct existing defective castings is the correct identification of the defects present, so that proper and effective corrective measures may be taken. This paper illustrates, by means of photographs of defective castings, a majority of the common defects experienced in aluminum sand foundries. The defects are described, the most probable causes are given and suggestions for their elimination are discussed.

The defects are classified as to type into seven main divisions, and each division covers the several sub-types and variations of the defect. The principal causes are discussed so that an understanding of the mechanism leading to the occurrence of the defect can be gained.

#### Introduction

In any foundry that is endeavoring to improve the general quality of their castings or is attempting to correct existing defective castings, the first and foremost consideration must be the correct identification of the defect so that the proper measures can be taken for its correction. Many foundries have often found that incorrect identification of defects has resulted in much wasted effort and money being expended on corrective measures which were not related to the defects whose correction was sought. Very often it is also found that in a particular foundry a wide difference of opinion will exist among various members of the foundry personnel regarding the nature of a defect and the proper remedies. The purpose of this paper is to illustrate a majority of the common defects experienced in aluminum sand foundries together with their identification and a number of suggestions for their elimination in subsequent production.

It is recognized that in many cases a defect can result from several causes and also that a number of remedies may be used to eliminate that particular defect. The descriptions of the defects will give the most prevalent or common name and the principal corrective measures will be listed and briefly discussed. In a considerable number of cases defects may appear to be similar when they are actually due to widely differing causes. In these cases where identification is difficult, experience must be the guide. When corrective measures have been applied and the defect still persists, the conclusion must be that the defect was incorrectly identified and the improper measures were used.

No attempt will be made here to establish standards of quality or limits of acceptability. Such standards can best be determined by agreement between foundry and customer. In many instances the defects listed are merely surface in nature and do not adversely affect the internal quality of the casting. Where castings are to be painted or where surface appearance is of little consequence, then many of these defects can be tolerated and the castings are quite acceptable for the customers' purposes. In other cases where a fine finish is desired on a casting, these same defects will be cause for rejection and steps should be taken for their correction.

#### I-Surface Defects

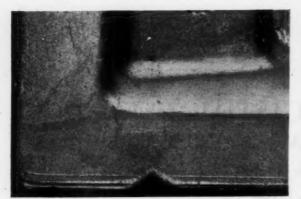
This category of imperfections will include only those which are evident on the surface of the casting and which, for the most part, are the result of causes existing outside the metal.

(a) Misruns—A misrun is an area in a casting where the metal has failed to completely fill the mold. It may be in the form of a hole or as an incompletely filled edge of a casting. An incompletely filled edge is illustrated in Fig. 1 while Fig. 2 illustrates a hole in the wall of a casting.

Misruns are most commonly caused by a low pouring temperature. However, many other factors can also cause this defect. Inadequate venting of the mold cavity is often encountered, which can include low permeability of the sand because of improper grain fineness distribution, hard ramming, or excessive moisture. It is well, in most cases involving thin-walled castings or large flat areas, to insure that the cavity is well vented by placing small holes through the cope surface from the mold cavity to allow the air and steam to escape rapidly. If the gas contained in, and generated in, the mold cavity is not vented, the metal must overcome a back pressure which can easily be sufficient to cause misruns. Inadequate

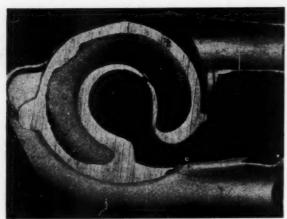
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^{*} Research Metallurgist, American Smelting and Refining Co., Barber, N. J.



Courtesy of Aluminum Co. of America

Fig. 1-A misrun at edge of a casting.



Courtesy of Aluminum Co. of America

Fig. 2-A misrun in wall of a casting caused by core shift.



Fig. 3—A form of misrun where portion of casting (the letters) did not fill because of very low permeability.

sprues, runners and gates are frequent causes of misruns; the metal may not flow fast enough or may be forced to flow too great a distance with the result that the metal is cooled to the solidification point before filling the mold cavity.

Figure 2 illustrates a hole in the wall of a casting which may be due to some of the above causes but is often the result of a misplaced or shifted core, which causes one wall to be considerably undersize with the result that the metal cannot fill the unusual thin section

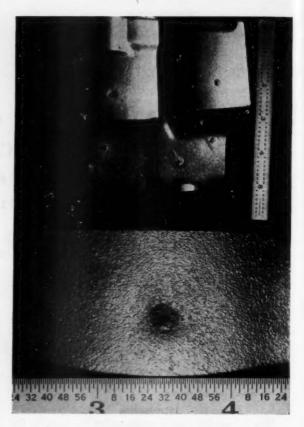


Fig. 4—Sand explosions caused by relatively large isolated foreign inclusions in the sand.



Fig. 5—A casting pitted by steam explosion of small clay balls in sand shown in Fig. 6.

A less common form of misrun is shown in Fig. 3 which illustrates some letters of the cope surface of a casting which did not fill in completely because of the extremely low permeability of the sand, which would not permit the escape of the steam generated on contact with the metal. This steam prevented the filling of the letter impression. This defect was eliminated by increasing the permeability of the sand.

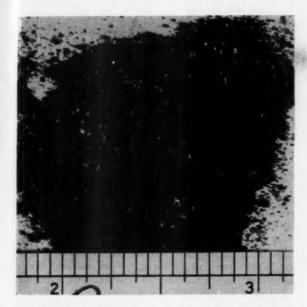


Fig. 6—Clay balls in sample of air-dried molding sand which caused surface pits on casting in Fig. 5.

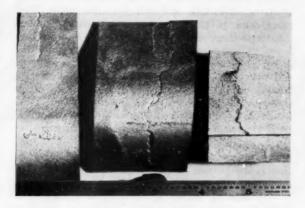


Fig. 7—Sand lines, or rat-tails, caused by buckling of sand at elevated temperatures.

Isolated cases of misruns can be the result of interruptions or fluctuations in the flow of metal during the pouring cycle. It is important in the pouring of any metal to maintain a smooth, uniform, and nonturbulent entry of the metal in the sprue system.

(b) Sand Explosions—This defect takes the form of small voids or craters on the surface of the casting which generally contain particles of sand or foreign material. They may be either in isolated occurrences or may be found generally distributed over the surface of the casting. Usually these two types are due to different causes.

The isolated type is usually caused by a fairly large inclusion of foreign material in the sand, which retains a large amount of water or volatile matter such as oil and is located on the mold surface. When the molten metal strikes this inclusion in the sand, steam or gas is rapidly generated with explosive force causing the crater in the casting. This type of



Courtesy of Aluminum Co. of America

Fig. 8—Sand buckles caused by expansion of sand before metal solidified.

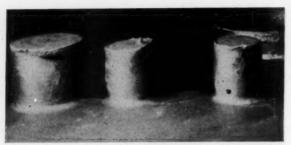


Courtesy of Aluminum Co. of America

Fig. 9—A sand scab caused by loosening of a loosely adherent layer of sand.

sand explosion is shown in Fig. 4. Complete elimination of this type of defect may be difficult since the size of the particles which cause the difficulty are most often smaller than the finest sand screens used in the average foundry. The best method is probably preventive in nature; that is, prevention of foreign material from entering the sand. If the occurrence of this type of defect becomes serious and it is impractical to completely screen the foundry sand through a fine sieve, then the sand had best be discarded.

The second type of this defect where the small pits or craters are more-or-less uniformly distributed over the surface of the casting, is illustrated in Fig. 5. This type of defect has usually been found to have been caused by an unusual sand condition where the clay content has separated from the usual uniform distribution throughout the main body of the sand and has collected into clay balls. Because of the fineness



Courtesy of Aluminum Co. of America

Fig. 10—A swelling of the casting caused by weak sand resulting from soft ramming.

of the clay particles in these clay balls, they tend to contain the greater part of the moisture content of the sand. When these clay balls are on the surface of the mold and are covered with the liquid metal, the sudden generation of steam causes the crater in the surface of the casting. Casual inspection of a sample of sand taken from the sand heaps will not usually reveal the presence of these clay balls. If this sand sample is spread out and allowed to dry for a short time, the clay balls will become clearly visible. Such a sample of dried sand containing clay balls is illustrated in Fig. 6, in which these balls are clearly evident. In this particular sample of sand the clay balls were removed on a 35 mesh screen and were found to comprise 14 per cent of the weight of the sand. The screened sand and the clay balls were then analyzed for A.F.S. clay content and it was found that the sand contained 1 per cent clay and the clay balls contained 10 per cent clay. The total sand contained 3.3 per cent clay, of which 45 per cent was contained in the balls. This analysis showed that nearly half of the clay was not available for contributing to the strength of the sand and hence the molder tended to use more water, which aggravated the occurrence of the defect.

These clay balls are most commonly found in foundries utilizing the heap sand system and less commonly found in synthetic sand systems. Their complete removal is difficult, if not impossible. The best way is to allow the sand to dry completely, when the clay balls become fragile and are easily broken up by any mechanical action on the sand. If this



Courtesy of Aluminum Co. of America

Fig. 11—Sand holes (upper) and a projection from a hole in a core.



Fig. 12—Rough, sandy surface (left) caused by rapid release of dissolved gas.

drying is not feasible, the most direct method of elimination of this defect is complete replacement of the sand.

(c) Sand Line, or Rat Tail—This is a defect which appears as an irregular line on the surface of a casting where the casting surface is at different levels on either side of the line. These lines are a surface phenomenon only and are not considered to be objectionable except from an appearance standpoint. They are not indicative of any internal or sub-surface unsoundness.

Sand lines are caused by the failure or buckling of the sand surface at elevated temperatures. The exact cause of this buckling is not clearly understood and there are many explanations of its occurrence. The appearance of typical examples of this defect is shown in Fig. 7. High moisture content and low permeability are believed to be closely related to the occurrence of this defect. Low green strength and low resilience of the sand are also believed to be related. They are most often found on large thin-walled areas of aluminum castings.

(d) Sand Buckle—This is a depression on the surface of a casting usually well defined with a sharp line at the bottom in the longitudinal direction. A common example is shown in Fig. 8.

This defect is caused by buckling or expansion of



Fig. 13—Coarse, crystalline, brittle fracture caused by iron contamination of the alloy.



Fig. 14-Dross and flux inclusions.

the mold surface at elevated temperatures. The properties of the sand mold leading to the occurrence of this defect are closely related to those causing the sand line and measures taken to eliminate either one will decrease the occurrence of the other.

(e) Sand Scab—This consists of excess metal in a localized area on the surface of a casting which is usually separated somewhat from the main body of the casting and is easily removed (Fig. 9). It is caused by the metal penetrating beneath thin layers of sand which have lifted from the mold surface. The sand lifting is caused by weak sand, hard ramming, high moisture content, and most often by patching of the mold which results in layers of loosely adherent sand which are easily dislodged by the action of the flowing metal or the sudden generation of steam. This defect is also occasionally related to the causes leading to sand lines and sand buckles.

(f) Soft Ram—This appears as a slight swelling or enlargement of the casting, which is illustrated in Fig. 10. This is caused by the sand in that particular area being unable to resist the hydrostatic pressure of the liquid metal. Its cause is primarily insufficient ramming of the mold or inadequate tucking of the sand into relatively inaccessible locations on the pattern surface, such as in between high bosses as shown

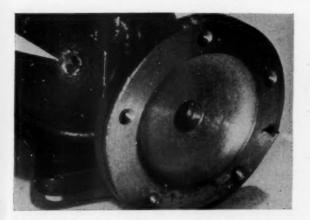


Fig. 15—Dross and flux inclusions which blistered after storage.



Fig. 16—Dross (oxide) inclusions in a casting resulting from excessive agitation in handling the molten metal.

in the illustration. Low flowability of the sand may also be related.

(g) Sand Hole—These are holes or cavities in the surface of the casting almost invariably in the drag surface. These are caused by the presence of loose sand on the mold surface. Loose sand has fallen to the drag portion of the mold because of a large number of reasons, such as, careless closing of the mold, or placement of cores; weak or dry molding sand, or soft ramming. Typical sand holes are shown in the upper portion of Fig. 11. The lower portion of this figure shows a projection of metal from the surface of the casting which was caused by a particle of sand having fallen from the surface of a core.

(h) Rough Sandy Surface—This defect occurs uniformly distributed over the entire cope surface of a casting and is intensified over heavy sections. It has the appearance of a rough, sand-papery surface to which sand more-or-less tightly adheres, as shown in Fig. 12. This is due to badly gassed metal. If the molten aluminum contains a large amount of dissolved gas when it fills the mold cavity, this gas is rapidly and strongly released from the metal as it solidifies. The bubbles of escaping gas breaking against the cope surface of the mold cause particles of sand to be entrapped in the surface of the casting. The causes and elimination of dissolved gas will be discussed later in the paper under heading VII (a) Gas (Hydrogen) Porosity.



Fig. 17-Dross inclusions in test bars.

#### II-Brittle Fracture

In Fig. 13 is illustrated one example of a brittle fracture in which the unusually crystalline appearance of the fractured surface is apparent. This casting was very brittle and difficult to machine. On analysis this casting was found to contain 3.5 per cent iron. The aluminum had been melted in iron crucibles which were improperly cared for in that the inside surface had never been coated with a refractory material to prevent the solvent action of the aluminum. In addition, the melting temperatures had been uncontrolled and had risen to high values. which markedly increased the attack of aluminum on iron. It was subsequently found that a large quantity of metal had become contaminated in this manner and had to be scrapped because of the re-use of gates and risers containing high iron. To guard against iron pickup in melting aluminum alloys in iron crucibles, it is essential that the crucibles be protected against the action of the molten aluminum. This is particularly true in sand foundries where the melting temperatures are higher than those used in die casting plants where iron crucibles are more prevalent.

This fracture is but one example of the contamination of alloys, which also may happen by accidental mixing of alloys of widely differing composition, which can also produce very brittle and unsuitable or unuseable metal.

#### III—Inclusions

It is possible for foreign inclusions of many different kinds to be present in aluminum castings. By far the most prevalent of non-metallic inclusions found in aluminum castings is aluminum oxide or a mixture of aluminum and magnesium oxides. These appear on a fracture to be gray to black hard particles which may appear to be in powder form in some instances, and in other cases appear to be in the form of a film. In other cases, these inclusions are mixtures of these oxides and flux. If a flux is used and is stirred into the metal without allowing for sufficient settling time for separation of the dross and flux, the dross and flux will find their way into the casting.

Figure 14 illustrates an aluminum sauce pan which appeared to be satisfactory immediately after polishing but on standing for a period of time was found to contain a large number of blisters on the surface. Examination of these blisters chemically and microscopically showed them to be mixtures of dross and flux. Figure 15 illustrates a small casting which was satisfactory when machined and painted but when taken from stock some time later exhibited the two blisters shown. In this case again, the inclusions were found to be mixtures of dross and flux which had reacted in time to produce the effect shown. If flux must be used, it is not always desirable to stir it into the body of metal, but if this is done, sufficient time should be given to allow the excess flux to float to the surface where it can be removed from the metal. As a general rule, most fluxes should be used only to assist in the removal of dross and should

not be introduced into the body of the molten atum-

The most common type of inclusion is that which consists of aluminum oxide, or in the cases of alloys containing magnesium, the inclusions are mixtures of aluminum and magnesium oxides. These oxides appear in many shapes such as those shown in Fig. 16 and Fig. 17. This type of inclusion is quite detrimental to the quality of the casting since if they are encountered in machining, they are of sufficient hardness to dull or break the edge of the hardest tool and also are a source of weakness in the casting. The properties obtained on the test bars shown in Fig. 17 were exceptionally low through no fault of the alloy quality but entirely due to the presence of the oxides.

It may be safely said that the primary source of all dross inclusions in aluminum castings is the presence of agitation somewhere in the handling of the metal between the ingot and the casting. It is essential in the production of high quality aluminum castings that all turbulence and agitation be eliminated as far as possible. The metal should not be stirred while in the melting furnace; all transfer operations should be performed with as short a stream between the two vessels as possible; and finally when pouring aluminum into sand molds the crucible or ladle should be held as close to the sprue opening as possible and the sprue should be kept full at all times to minimize turbulence, agitation, and air entrapment at this point. One of the primary sources of oxide and dross inclusions in aluminum castings is improper pouring technique.

Also of importance in the production of dross and oxide free castings is the design of the gating system. These systems should be studied with care to insure that the runner and gating system introduces the metal into the mold cavity with a minimum of disturbance.

#### IV-Blows

(a) "Normal" or Open Blows—This type of blow generally appears as a smooth-walled, rounded or globular void or cavity usually open to the surface of the casting. Blows take many shapes and forms and may occasionally be difficult to identify as such. Some typical examples are shown in Fig. 18.

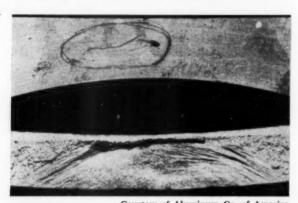
Blows are caused by sudden locally developed gas pressure, usually steam. Most blows in the vicinities of chills are caused by cold or damp chills. When the molten metal strikes this cold damp surface, the steam generated causes a blow into the casting. Other causes are moisture from wet sand, low permeability of the sand and paste or wet mud in the mold. Blows also may be caused by excessive core gas or gas from cores that are inadequately vented. Improperly baked cores are a source of excessive gas production. Too liberal use of a swab especially in conjunction with sand of low permeability, is a source of blows. Chills should be warm, clean and dry before placing in the mold. Often it is necessary to warm the chills in the mold with a torch, just before closing the mold and pouring, to prevent condensation of steam on them during the flow of metal through the mold.



Courtesy of Aluminum Co. of America

Fig. 18—Blows resulting from excessive moisture (above)

and a chill (below).



Courtesy of Aluminum Co. of America Fig. 19—A collapsed blow shown on machined surface (above) and as opened by tracturing (below).



Courtesy of Aluminum Co. of America Fig. 20—A hot crack,

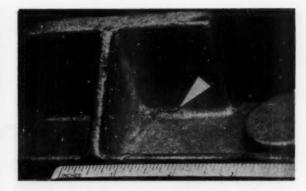


Fig. 21—A crack resulting from shrinkage at junction of light and heavy sections.

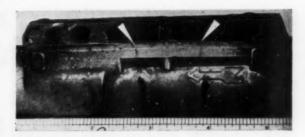


Fig. 22—Cracks resulting from use of cold quench water following solution heat-treatment.

(b) Collapsed Blows—Blows which are formed as described above are occasionally closed again by further movement of the molten metal. In such cases the blow appears to be a fine line or crack on a machined surface. A fracture through such an area will usually reveal a typical structure such as that shown on Fig. 19, indicating that it is a collapsed blow.

#### V-Cracks

Cracks in castings can be among the most difficult defects to precisely identify as to their cause. In many cases laboratory microscopic examination is necessary to establish the exact cause of a crack. The mechanical type of crack resulting from mishandling of castings will not be discussed since its cause is usually obvious.

(a) Hot Crack or Solidification Shrinkage Cracking—This crack occurs during the solidification of the alloy or while the casting is in the temperature range immediately below the solidification point. It is caused by restraints on the casting which prevent the casting from shrinking the amount required by thermal contraction. Aluminum alloys differ markedly in their resistance to this type of cracking. Some alloys are highly prone to this defect and caution must be used to insure that the casting is not restrained during cooling. Other types of alloys, notably high silicon content types, are resistant to this defect and are rarely troubled with it.

The cause of this crack is restraint of the casting during cooling. This may be caused by hard cores or cores of insufficient collapsibility. Hard ramming



Fig. 23-Quench crack resulting from cold quench water.



Fig. 24—Shrinkage in a thin casting resulting from use of one gate.

can also cause hot cracks. If hot cracking of a casting persists in spite of corrective measures, then it is quite possible that the design of the casting is inherently unsuitable for an alloy subject to hot cracking. Figure 20 shows an example of a hot crack. There are no outstanding visual characteristics of this type of crack. It is usually true, however, that there is no evidence of shrinkage on the surface in the neighborhood of this type of crack. The effect of gating on the occurrence of hot cracking should not be overlooked when attempting to eliminate this defect. If the number of gates is inadequate or their location is not the best, then too much metal may flow through one gate and one area of the casting may become overheated. This area will then be the last to freeze, causing the contraction stresses to be concentrated at this point with the result that a crack is formed. Aluminum should enter the mold cavity at as many points as practical, with some thought given to the equal and uniform distribution of the metal.

(b) Shrink Crack—This type of crack (Fig. 21) is the result of severe surface shrinkage which results in the parting of the surface of the casting to form a crack. Its cause is usually obvious because of a depression and sponginess in the surface of the casting which indicates shrinkage. The remedies for this type of crack are the usual ones taken to reduce shrinkage such as the use of chills, risers, and gating changes to improve the distribution of hot metal.

(c) Quench Crack—This type of crack results from

stresses imposed on the casting by the cooling during the quenching process following solution heat treatment. These stresses exceed the tensile strength of the alloy in the area of the crack which results in a fracture. Two examples of this type of crack are shown in Figs. 22 and 23. The principal cause of quench cracks is the use of cold quench water. The temperature of quenching water should be above 150 to 180 F. It is possible for the design of the casting to be such that even with boiling water the stresses are still sufficiently severe to crack the casting. The safest procedure to insure that quenching cracks are minimized is to use boiling quench water.

## VI-Shrinkage

The subject of shrinkage in aluminum castings is very broad and could well be the subject of an extensive article in itself. Since it encompasses the entire subject of gating, chilling and risering of castings, the discussion here will be necessarily limited to covering a few of the basic principles with several illustrations of the common types of shrinkage. When aluminum alloys solidify, there is a decreasing change in volume on solification of the metal. If this decrease in volume is not compensated for by adequate feeding, then shrinkage will be present in the casting in the form of finely dispersed porosity, localized voids in the interior of the casting, or as surface holes or depressions. Occasionally shrinkage is classified into two types as surface shrinkage and interior shrinkage, but this is not believed to be a valid classification because the two types cannot be differentiated as to cause. Shrinkage defects will be presented here under three broad classifications according to cause.

(a) Too Few Gates—Figures 24 and 25 illustrate surface shrinkage on two castings which was primarily caused by the use of one gate to feed the entire casting. Unless the casting is of great simplicity, the use of one gate to feed a casting will often result in the occurrence of shrinkage in the vicinity of the gate. This is caused by too much metal passing over one area in the mold, with the result that the sand becomes overheated, causing that area of the casting to feed last with the result shown in the photographs. Almost every case of this type of shrinkage can be easily eliminated by the use of two or more gates. The use of a larger number of gates is also beneficial in that a lower pouring temperature can usually be used with an attendant improvement in the appear-



Fig. 25—A second example of shrinkage caused by use of one gate.



Fig. 26—Surface shrinkage caused by too high a pouring temperature.



Courtesy of Aluminum Co. of America

Fig. 27—Surface shrinkage caused by high pouring temperature.

ance of the casting. This is a very common cause of shrinkage in aluminum castings.

(b) Too High a Pouring Temperature—Figures 26 and 27 also illustrate surface shrinkage on castings where the gating was probably adequate but an excessively high pouring temperature resulted in the appearance shown. Shrinkage is often aggravated by high pouring temperatures, and when it is suggested to the foundryman that the pouring temperature be lowered, the usual reply is that the casting will not run. Usually in cases like this the casting will not run with a moderate pouring temperature because of inadequate venting, low permeability sand, or inadequate gating. If some of these other factors are corrected, then the pouring temperature can be lowered and the shrinkage difficulty will disappear.

(c) Inadequate Feeding—Two examples of surface



Fig. 28—Surface shrinkage cavity resulting from lack of a riser on heavy section.



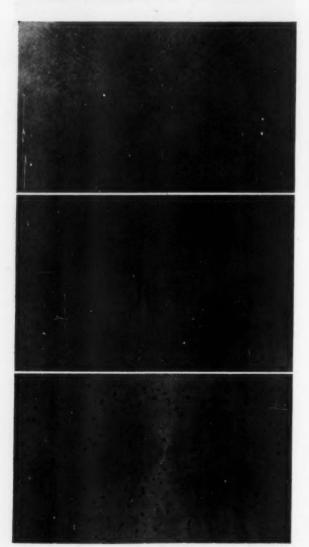
Courtesy of Aluminum Co. of America

Fig. 29—Shrinkage cavity resulting from an inadequate riser.

shrinkage cavities caused by inadequate feeding are shown in Figs. 28 and 29. In both cases the shrinkage occurred at the junction of a light and a heavy section, and the risering was insufficient to feed the shrinkage occurring in the heavy section. The shape, location and size of risers govern the degree to which they will perform the job of feeding the casting. Small errors in any of these factors often cause great differences in the results. A common error is too small a neck connecting the riser to the casting, which results in this neck freezing first before the casting has solidified and hence the riser cannot feed. The riser is often located too far from the section it is supposed to feed and the liquid feeding metal cannot reach the shrinkage area. The shape of the riser is important-tall thin risers are generally inadequate and ineffective. Generally speaking the riser should



Fig. 30—Shrinkage at gate resulting from use of a combination sprue, gate and riser which did not feed.



Courtesy of Aluminum Co. of America

Fig. 31—Various degrees of pin hole porosity resulting from dissolved gas as they appear on machined surfaces.



Fig. 32—Pin hole porosity in saw cut surface, and trapped air bubble porosity on sand blasted surface.



Fig. 33—A severe case of pin hole porosity as revealed on a saw cut section.

be short, broad and closely connected to the section which is to be fed. Gating can often affect the efficiency of a riser since a riser must contain hot metal and remain liquid after the casting has solidified. Where possible it is generally good practice to gate through a riser into the casting, causing the riser to be the last portion of the mold to be filled. Risers which are located on the side of a casting opposite the gates are usually ineffective because they will then contain the coldest metal in the mold and cannot adequately feed the casting.

Chills are a very effective tool in the control of shrinkage. One of the functions of a chill is to equalize the rate of solidification between a heavy section and an adjacent thin section so that they will freeze at uniform rates thereby minimizing the possibility of shrinkage. Chills are also useful in promoting directional solidification so that a riser can more effectively act. When attempting to feed an exceptionally heavy section of a casting, it is often found that a riser cannot feed because the casting does not solidify quickly and directional solidification toward the riser cannot be obtained. In this case chills on the heavy section of the casting will facilitate feeding by the riser. The use of insulating riser sleeves and ex-

othermic riser compounds are very effective in improving the feeding characteristics of risers and allowing their size to be decreased while maintaining ade-

quate feeding.

Figure 30 illustrates an internal shrinkage defect discovered when the combination gate, riser and sprue was removed from the casting. A slight surface depression was visible on the casting before trimming. In this case a separate sprue should have been used with a runner to the casting so that an adequate riser could have been placed on the gate to feed the heavy section in the casting.

## VII-Internal Porosity

None of the defects to be discussed in this section are apparent on the surface of the casting when it is removed from the sand. The porosity is only revealed when the casting is grit blasted for cleaning or given any type of surface treatment which removes some metal such as, wire brushing, sawing or machining. The sources of internal porosity are several in number and they will be discussed separately. The type of internal porosity resulting from well dispersed shrinkage will not be covered here, although it is often quite difficult to establish in a particular case whether the porosity is due to dissolved gas or to

shrinkage.

(a) Gas (Hydrogen) Porosity—This porosity is found in the form of small pin holes generally uniformly distributed throughout the casting in thick and thin sections. The pin holes are usually well defined and more or less rounded voids but may not always be in the form of spheres since their shape is often affected by the alloy, the pouring temperature and the degree of gassing. Figure 31 illustrates several degrees of gas porosity as they appear on machined surfaces. Fine gas porosity is also shown on the saw cut surface of the casting illustrated in Fig. 32. A severe case of pin hole porosity caused by dissolved gas is shown in Fig. 33. Examination of a small area or a few pin holes cannot give sufficient evidence to determine whether the porosity is shrinkage or gas. If the porosity is localized in a relatively small area, then it is probable that it is due to shrinkage. If the porosity is found throughout the casting, it is probably the result of gas which dissolved in the melt during processing.

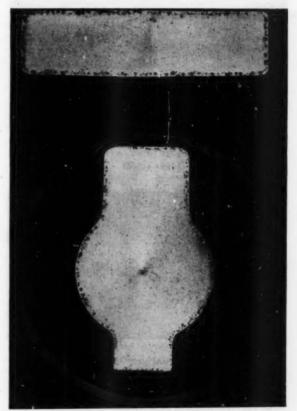
It is generally conceded that hydrogen gas is the principal source of porosity from dissolved gas in aluminum alloys. The hydrogen is picked up by the aluminum from any source of moisture or hydrocarbons from which hydrogen can be derived. Aluminnum reacts readily with moisture in any form with the production of aluminum oxide and hydrogen gas, which then dissolves in the aluminum. The solubility of hydrogen in aluminum increases rapidly as the temperature of the molten metal rises. One of the most common causes of gassed aluminum castings is improper temperature control during melting where the aluminum is allowed to become very hot. If the aluminum that has been over-heated is allowed to cool slowly down to a low pouring temperature, the excess gas will be released from the metal and the castings will often be sound. However, this purging

action is not reliable and quite often the castings will be found to be porous.

When aluminum containing dissolved gas is poured into the mold and solidifies, there is an abrupt change in the solubility of the gas since it is relatively insoluble in solid aluminum. The dissolved gas comes out of solution forming the voids as shown in tillustrations.

In oil- or gas-fired furnaces the character of the combustion gases is important. If the flame is reducing in nature, that is, if there is insufficient air for complete combustion, the presence of unburned hydrocarbons are a source of hydrogen which can be dissolved in the aluminum. From the standpoint of minimizing hydrogen porosity, the combustion gas should be neutral or slightly oxidizing. The relative humidity of the air used for combustion is also a measure of the degree to which a melt of aluminum can be gassed. High humidity air is especially bad if the metal is over-heated. The use of damp tools, damp flux, or the charging of damp or corroded metal, are other possibilities that should be checked when attempting to eliminate pin hole porosity.

Dissolved hydrogen can be removed from molten aluminum by several methods, the most common of which is bubbling a gas such as nitrogen (dry, oil pumped) through the metal for about 5 min per 100 lb of metal. The bubbling of chlorine gas through molten aluminum is generally conceded to be the



Courtesy of Aluminum Co. of America

Fig. 34—Reaction porosity found just beneath surface on all surfaces of the castings.



Fig. 35—Porosity resulting from steam evolution in mold.

Note how porosity follows contours of blades.



Fig. 36—Severe porosity from steam generation in a wet mold.



Fig. 37—Porosity resulting from air and dross entrainment in metal stream.

most effective method of removing dissolved hydrogen but its use is not so common since it entails the removal of the unpleasant fumes which are corrosive on iron and steel equipment. Dry fluxes which evolve

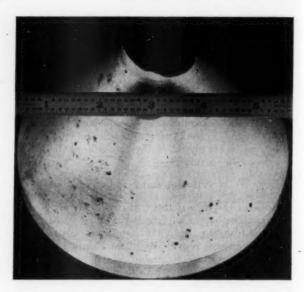


Fig. 38—Air and dross porosity from severe agitation in pouring.

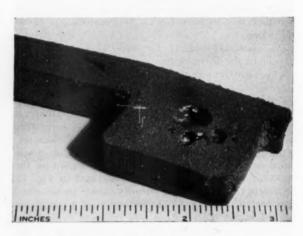


Fig. 39-Air bubbles adjacent to the gate.



Fig. 40-Air bubbles and dross inclusions in gate area.

gases when plunged beneath the surface of molten aluminum are also used but they are not considered to be as effective as the use of nitrogen or chlorine. In addition, it is essential that the fluxes be dry or their use is nullified by the moisture which is introduced in the metal with them.

- (b) Reaction Porosity—This type of porosity is found as a concentration of small spherical voids immediately beneath the surface of the casting and almost invariably on all surfaces-cope, drag and sides. Cross-sections through two castings having this type of defect are shown in Fig. 34. The principal cause of this type of defect is reaction of the metal with excessive moisture in the molding sand. The two distinguishing characteristics are that it is immediately beneath the surface and is found on all surfaces of the casting.
- (c) Gas (Steam) Porosity—This porosity is caused by the generation of large volumes of steam in the mold as it is being filled with molten aluminum and is the result of excessive moisture in the molding sand. This type is distinct from that discussed above in that here the porosity is caused by bubbles of steam rising through the molten metal, whereas the reaction porosity is caused by a reaction with the moisture in the sand resulting in the small voids present uniformly beneath the surface of the casting. The steam type of porosity discussed here is invariably found beneath the cope surface of the casting and not on the drag or bottom surface of the casting. Two examples of this type of porosity are shown in Figs. 35 and 36. In both of these cases the metal immediately adjacent to the porosity is sound indicating that there was a minimum of reaction between the steam and the body of the metal. The excessive use of a swab around the edges of a mold cavity can result in the type of porosity shown in
- (d) Entrained Gas (Air) Porosity-This type of porosity is one of the most prevalent defects found in aluminum sand foundries. The porosity appears only on the cope surface of the casting or the upper surface of an internal portion of the casting. It appears in the form of rounded to very irregular voids which may be clean or may be associated with dross. They are caused by agitation in the flowing metal, probably the major portion of which occurs in the sprue. They are also caused by agitation in the gating system or as the metal flows through irregularities within the mold cavity. Two examples of small voids partially filled with dross and oxides, are shown in Figs. 37 and 38. Two other examples where the voids are larger and are more obviously air-bubbles are shown in Figs. 39 and 40. Figure 32 in addition to showing dissolved gas porosity on the saw cut surface, also shows entrained gas porosity on the surface of the casting surrounding the saw cut. This porosity was revealed by a normal sand blasting operation. The area shown in Fig. 40 is that immediately above a gate in a very large heavy casting.

This type of defect is caused primarily by one of the most neglected phases of aluminum foundry practice-the pouring operation itself. It has been observed that aluminum foundrymen will control and carefully check all other phases of the casting of aluminum and then allow the metal to be poured in a most haphazard and unsupervised manner. Because

of the low density of aluminum (as compared to brass or bronze), bubbles (and associated dross) introduced into the metal stream are not readily broken and are easily carried in the moving stream of metal to find their way into the casting where they rise to the surface, giving the defect illustrated. It is most essential that the sprue be kept full of metal from the beginning of the pour through to the end. The crucible or ladle must also be held within an inch or two of the top of the sprue to prevent the entrainment of air into the metal stream because of the force of the metal entering the sprue. If the pouring operation is correctly done the oxide skin on the surface of the metal in the ladle, on the stream, and on the surface of the full sprue will be continuous and stationary, with the liquid metal moving beneath it. When this condition exists it is not possible for air to enter the metal stream at this point. The design of the gating system should be such that the metal flows smoothly without turning sharp corners or tumbling over irregularities and is introduced into the mold cavity without splashing. Often a gate is at a high point on a casting and the metal is forced to fall vertically several inches or a foot within the mold. The resultant splashing can easily lead to the types of defects discussed. Application of the principles demonstrated in the A.F.S. motion picture, "Fluid Flow in Transparent Molds" will do much towards the elimination of defects such as discussed in this section.

In conclusion, it is hoped that this discussion has served to illustrate the most common defects experienced in the production of aluminum alloy sand castings. Because of space limitations it has not been possible to completely discuss each type and the many variables that enter into its cause. Also, it has not been possible to completely explain all of the corrective or preventive measures that could be used for their elimination. The correct identification of the defects so that the proper remedies are used, remains the most important factor in improving the quality of castings. Increased knowledge of the types of defects and their causes should enable foundrymen to more easily improve the quality of aluminum sand castings and thereby benefit the industry as a whole.

## DISCUSSION

Chairman: A. CRISTELLO, American Light Alloys, Inc., Little

Co-Chairman: H. E. Elliott, The Dow Chemical Co., Bay City, Mich.

Recorder: A. CRISTELLO.

HIRAM BROWN: 1 I should like to commend the author of this paper not only for the excellent technical content and his contribution to the industry, but for his courage in presenting such work. This is a highly debatable subject. It has taken the author several years to collect the necessary samples and photographs required for this work. I am sure that the paper will be an important contribution to A.F.S. knowledge, and it is hoped that this will become a part of our revised Recommended Practices for the Light Metal industry. R. C. Военм: ² Mr. La Velle has presented a fairly complete

report on the common aluminum sand casting defects. However,

² Chief Metallurgist, Solar Aircraft Co., Des Moines, Iowa,

we have recently encountered a condition which I believe merits some discussion, namely pinhole porosity on the drag side of alloy 355 castings. This defect has also been observed on alloy 356 parts, although not as pronounced. The severity of the porosity appears to be related to the section size, the voids generally are larger and more numerous at the more massive areas of the casting.

The addition of about 1 per cent boric acid to the sand mixture was found to eliminate the surface voids. Furthermore, no adverse effect was observed on castings of alloys 48, 356, 220, or 195. This is the opposite to the results reported earlier in the discussion that pinhole porosity on the cope surface of alloy 356 castings was obtained when small amounts of boric acid was accidentally introduced into the sand.

C. F. ZABRISKIE: I should like to submit as an additional cause of pinhole (surface) porosity in Alcoa 355 and 356 aluminum alloy castings, that as little as 0.05 per cent of boric acid in aluminum molding sand as an impurity picked up from magnesium molding sand, can cause severe pinhole porosity particularly in heavy sections of castings.

This condition was verified by pouring heavy test castings, 3 in. x 3 in. x 6 in. at 1400 F in new molding sand of the following composition as control:

Grade "E" Washed Silica Sand (92 A.F.S. Fineness) 6% Bentonite (50% Western, 50% Southern)

4% Moisture

Then to some of the above sand 0.05 per cent boric acid was added and well mulled. Molds from the same pattern were then poured also at  $1400~\rm F.$ 

Those castings poured in new sand were entirely free of pinhole porosity, while those poured in boric acid treated sand contained severe surface pinhole porosity extending to a depth as great as 3/32 in. This was verified by slicing a ½ in. section across the long axis of the casting and X-raying it to show the depth of porosity.

Where aluminum molding sand regularly becomes contaminated with boric acid, the following action may be taken to correct the ill effects of same:

- 1. Periodically replace the entire contaminated heap or system sand.
- 2. Isolate the aluminum floor entirely from the magnesium floor including separate core shake-out facilities.
- 3. When contamination reaches the point where porosity commences to appear, face all heavy molds with new molding sand and back up mold with the contaminated sand. This will prevent formation of pinhole porosity from this source.

MEMBER: If aluminum alloys are cast in sand used for magnesium alloys the boric acid in the sand caused surface defects in the aluminum castings.

M. E. BROOKS: A I know of one foundry where No. 220 alloy was cast in magnesium foundry sand with good results. Perhaps the inhibitors in magnesium foundry sand affect some aluminum alloys but not others. What was the alloy referred to above?

MEMBER: The alloy referred to was No. 256 alloy.

² Chief Metallurgist, Wellman Bronze & Aluminum Co., Cleveland.

 $^{^{\}rm a}$  Foundry Metallurgist, Sperry Gyroscope Co., Div. of The Sperry Corp., Great Neck, Long Island, N. Y.

⁴ Foundry Engineer, The Dow Chemical Co., Bay City, Mich.

## BASIC REFRACTORIES FOR CUPOLA SERVICE

By

M. W. Demler*

#### ABSTRACT

Since the early development of the basic-lined cupola for commercial iron melting, a number of very informative papers have been written on the subject. These papers have dealt with the differences in metallurgy involved between acid and basic practice, the water-cooled cupola versus the conventional cupola, as well as many of the other complicated problems connected with basic cupola operation.

It is the furpose of this paper to briefly review the refractories used for acid practice as compared with those required for basic practice, and to discuss the application of basic refractories for cupola linings.

In acid cupola practice, slag has generally been treated merely as a necessary accessory. As long as the slag has been sufficiently fluid and of the lowest possible volume, little interest has been shown except in its disposal. Since the limestone charge is held to a minimum, and the coke ash is acid in character, an alumina-silica refractory such as the usual fireclay brick of the several commercial classes is suitable for the linings

The acid slag results from the fluxing of coke ash (SiO²+Al₂O₃), sand (SiO₂), the firebrick lining (Al₂O₃+SiO₂) and the limestone. The calcium-silicate slag has low basicity with limited desulphurizing properties. In order to desulphurize in the cupola, it is necessary to increase the basicity of the slag by increasing the percentage of limestone and sometimes other fluxing agents in the charge. This causes excessive lining erosion of the conventional alumina-silica refractories. It was evident that, if desulphurizing was to be accomplished in the cupola, refractories that would resist the reaction from the basic slags were a necessity.

## Physical Properties of Basic Lining Materials

Several decades ago only two general classes of basic refractories were available, namely, prefired magnesite and chrome brick. Today, many more kinds are manufactured regularly and used extensively. While there is considerable overlapping in their many uses, the distinctive properties of each render it particularly suitable for certain conditions and purposes.

*Technical Sales Dept., Harbison-Walker Refractories Co., Pittsburgh, Pa. Table 1 shows in general terms the comparisons in the physical properties of the more or less standardized classes of basic refractories now manufactured in the United States.

In addition to the important properties of basic refractories such as chemical composition, spalling resistance and high temperature strength, for use in cupola linings, thermal expansion and conductivity are of particular significance in connection with the most advantageous installation.

Figure 1 shows the reversible thermal expansion of several refractory materials. Alumina-silica brick expand at the rate of approximately ½6 in./ft, while magnesite brick, as shown in the curve, expand approximately three times as much or ¾6 in./ft over the operating temperature range of the cupola. Provision must be made for this increased expansion. This is done by placing cardboard strips or other readily combustible material between the brick around the perimeter of the lining, as well as between the vertical courses of basic brick. As the temperatures increase, the cardboard strips will burn out as the brick expand.

Some exception is taken to this in the well of the cupola, since initial temperatures are not sufficiently

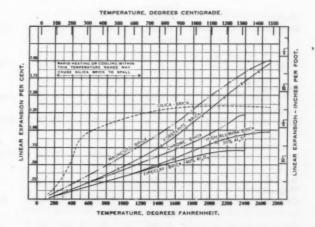


Fig. 1-Reversible thermal expansion of refractory brick.

high to absorb this extra expansion allowance, and leakage of metal might occur. It has been found satisfactory to allow expansion only in the melting zone

proper of the cupola.

Figure 2 illustrates the thermal conductivity of several refractory materials. This is important when considering basic linings, since the heat conductivity through a lining composed of magnesite refractory is approximately double the amount of heat transfer through a high-duty fireclay lining at a temperature of 2800 F. In order to prevent damage to the cupola shell, it is recommended that a course of high-duty fireclay brick be placed between the inner basic lining and the shell.

## Refractory Lining Materials

Experience has shown that refractory linings in basic cupola operation are subject to so many operating variables that predictions based on theory or analogy regarding their life in service are likely to be wide of the mark. Results achieved from a given refractory material at any one plant under the conditions there prevailing are generally an insufficient basis for definite conclusions as to the probable behavior of the same material in another plant where operating conditions are different. It must be remembered that, if a basic lining is used, a basic slag must be maintained by increasing the limestone charge over that previously used on the acid operation.

Hard-burned magnesite brick, containing 88 per cent magnesia are most generally being used for lining basically operated cupolas. This is the same type of refractory that has been used successfully through

TABLE 2—COMBINATIONS OF SHAPES REQUIRED FOR VARIOUS BASIC LINING DIAMETERS

Diameter	Pieces per Course Required for Cupola Linings (see Fig. 3)		
Inside Lining, in.	Shape SW-6-1	Shape SW-6-2	
27	10	18	
32	16	15	
37	21	14	
42	27	11	
45	30	10	
48	33	9	
54	39	7	
60	47	4	
66	53	2	
72	59	_	

the years for basic open-hearth steel furnaces, basic electric melting furnaces, and in non-ferrous refining and melting furnaces.

A somewhat newer class of magnesia brick than the usual burned basic magnesite refractory has been found satisfactory for certain locations where both the metal and basic slag impose more severe service on the refractories. This refractory, accurately termed periclase, is higher in magnesia content and lower in other oxides such as silica, lime, ferric oxide and alumina than the conventional magnesite brick. Periclase brick, containing approximately 92 per cent MgO, have a definite application for tap-holes, lining troughs of front slagging cupolas, and forming the slag knife.

High-fired chrome-magnesite refractories are being used successfully for the breast as well as for slag-hole

TABLE 1-PROPERTIES OF BASIC REFRACTORIES

Туре	Bond	Principal Minerals	Apparent Porosity,	Weight, lb/cu ft	Modulus of Rupture, psi	Resistance to Spalling	Stability of Volume at High Temperatures	Strength at High Temperatures	Thermal Expansion at 2600 F, %
Chrome	Fired	Chrome- Spinel	18-21	190-200	1200-2000	Fair	Fair	Fair	1.2
		Chrome-							
Chrome-		Spinel							
Magnesite	Chemical	Periclase	18-22*	192-202	750-1200	Excellent	Excellent	Excellent	2.1
Chrome- Magnesite	High- Fired	Chrome- Spinel Periclase Forsterite	20-24	184-194	1100-1500	Good	Excellent	Excellent	1.2
Magnesite 83-88 MgO	Fired	Periclase	18-22	165-175	2000-3000	Fair	Fair	Fair	2.1
Periclase									
92-93 MgO	Fired	Periclase	20-26	167-177	2200-2600	Good	Good	Excellent	2.1
Magnesite- Chrome	Chemical	Periclase Chrome- Spinel	18-22*	177-187	1000-1400	Excellent	Good	Good	2.1
Spinel- Bonded Magnesite-	1	Periclase Chrome- Spinel							
Chrome	Fired	Magnesia-	21-24	170-180	800-1100	Excellent	Excellent	Excellent	1.5
		Alumina- Spinel							
Forsterite	Fired	Forsterite	21-24	157-167	750-1000	Good	Good	Excellent	1.8

^{*}Porosity determined after heating the brick to remove non-permanent bonding ingredients from pores.

[†]The strength at high temperatures is measured by the temperature of failure in the load test at 25 psi. "Fair" corresponds to a temperature of failure between 2400 and 2600 F; "Good," 2600 to 2850 F; "Excellent" above 2850 F.

construction on a rear-slagging cupola. Some of the desirable physical properties of this refractory are attributable to the unusually high temperature at which the brick are fired. It follows that its constancy of volume in high temperature soaking heats is exceptionally good. Spalling resistance is higher than that for other prefired basic refractories. In strongly reducing atmospheres at high temperatures, the oxides of iron do not affect seriously the chrome-magnesite composition.

Burned chrome-magnesite brick have not been used for lining the melting zone of the basic cupola since some metallurgy precludes the use of chrome due to the danger of metallic chrome pickup in the metal.

While forsterite is one of the newest types of basic refractories, it is well established for use as a furnace lining material in many applications. The results of some few trials of forsterite brick in a basic cupola melting zone lining indicate that further investigation is warranted. The mineral forsterite (2MgO.SiO₂) has the composition corresponding to 57.3 per cent MgO and 42.7 per cent SiO₂. The melting point of the pure mineral is 3470 F. Forsterite refractories are characterized by their unusually high strength under load at high temperatures. This property, together with its high refractoriness and chemical composition, commended it for consideration in the melting zone of basic cupolas.

With the development of chemically bonded magnesite brick having high resistance to spalling and some other especially desirable properties, the use of a basic lining of this type also deserves investigation.

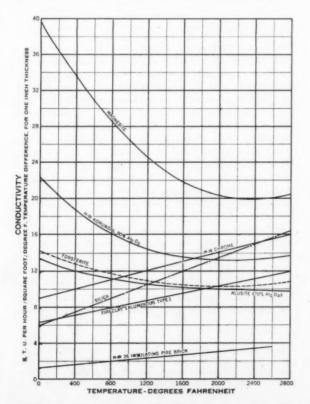
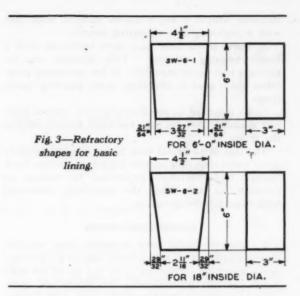


Fig. 2-Approx. thermal conductivity of refractory brick.



According to practice as most widely established at present, the basic lining is installed from the bottom of the cupola to a height of approximately 4 ft 6 in. above the tuyeres. Experience, so far, indicates that it is not economical to carry the basic lining for the entire height to the charging door.

Figure 3 illustrates shapes SW-6-1 and SW-6-2 which are 6 x  $4\frac{1}{2}$  x 3-in. key brick, and are laid to form a 6-in. lining thickness.

Table 2 indicates the number of pieces of each of the above shapes necessary to form a given inside diameter of lining. If a 12-in. total lining is used in the acid operation, an inner 6-in. lining of basic brick backed up with a 6-in. lining of regular highduty fireclay cupola blocks is recommended.

A 2 or 3 in. thick high-duty fireclay brick course may also be used back of the 6-in. basic lining, depending upon the total thickness of lining desired.

The basic brick are advantageously laid with dipped joints of air-setting high magnesia content mortar.

In the melting zone expansion is provided for by placing cardboard strips between the brick, both around the perimeter and vertically, allowing for an expansion rate of approximately  $\frac{3}{16}$  in./ft.

For the slag-hole and tap-hole, the high-fired high-magnesia or burned chrome-magnesite refractories, as well as basic ramming mixes, are reasonably satisfactory, although results are somewhat variable. Some operators have used a cast bronze water-cooled slag tap which serves the purpose and lasts for several heats.

#### Refractories for Daily Maintenance

In some cases it has been found that, if a basic slag is maintained, the consumption of the refractories in a basically operated cupola is much less than is the case in acid operation. Three general methods are being used for maintenance by patching. These are as follows:

1. Allow the lining to be slagged back after several days of operation until there is sufficient space for

patching with  $9 \times 41/2 \times 21/2$ -in. straight brick laid with a high-magnesia air-setting mortar.

2. Hand patch after each day's operation with a basic ramming mixture. This mixture can be rammed to a dense monolith. It has air-setting properties which work to advantage in the patching operation.

3. By means of an air-placement gun, various basic gun mixtures may be used for daily maintenance of the lining.

The cost of an initial basic lining is approximately three times that of a regular high-duty fireclay brick lining. It is therefore important, for reasons of economy, to fully study the metallurgy connected with basic cupola operation.

## **General Considerations**

1. After the initial basic lining has been installed, it is generally agreed that a brush coat of a high magnesia air-setting mortar, or 1/2 to 11/2 in. of the basic batch provides protection to the lining and minimizes wear during the first heat.

2. The same bottom sand which is used for acid operation may also be used for basic operation, except that from ½ to 1 in. of a basic ramming material is usually tamped in place on top of the sand.

3. Gas or oil burners should be placed in the cupola after the basic lining has been installed. A minimum of 8 hr is required to completely dry the lining, to establish satisfactory temperature equilibrium through the basic brick, and to allow the initial

expansion to occur. This is a proper safeguard against thermal spalling of the basic refractories when the cupola is initially operated. If ramming mixes are used for slag and tap holes, a separate burner should be used to completely dry them and obtain a ceramic set.

## DISCUSSION

Chairman: R. A. WITSCHEY, A. P. Green Fire Brick Co., Chi-

Co-Chairman: S. F. CARTER, American Cast Iron Pipe Co., Birmingham, Ala.

Recorder: R. A. WITSCHEY and S. F. CARTER.

W. R. JAESCHKE: 1 Will you please expand your comments relative to the application of Forsterite?

MR. DEMLER: Yes, Forsterite is extensively used in glass tank checkers and in the basic copper converter with satisfactory results. Forsterite brick are slightly cheaper than other basic brick. Investigations are being continued on the development of maintenance materials using olivine. We have it in a couple cupolas in test panels now and initial observations are encouraging.

in test panels now and initial observations are encouraging.

A. H. Thomson: ^a Have you tried steel plates between brick, like metal casing of brick?

MR. DEMLER: We did not try this as far as I know.

Mr. Thomson: This practice has been considered in Canada for that section immediately above the tuyeres for the same reasons that apply in electric furnace and open hearth practice.

CHAIRMAN WITSCHEY: Mr. Roudabush, has your company had any experience with metal case brick in cupola operation?

N. W. ROUDABUSH: ⁸ No, we have not. We would be concerned over the iron oxide slag reaction. We too are investigating the field of patching materials and have tried unburned chemically-bonded magnesite chrome.

Whiting Corp., Harvey, Ill.
 Canadian Refractories, Ltd., Montreal, Que., Canada.
 General Refractories Cσ., Philadelphia.

## APPLICATION OF CHILLS TO INCREASING THE FEEDING RANGE OF RISERS

By

E. T. Myskowski,* H. F. Bishop* and W. S. Pellini**

#### ABSTRACT

The length of steel plate and bar castings which can be made sound when directional solidification is aided by the use of adequate steel chills has been determined radiographically and related to the thermal gradient conditions in both the casting and chill. It was found that for steels containing 0.25 to 0.35 per cent carbon the maximum distance in inches (D) which can be made sound with one riser follows the empirical rules given by the formulae D (Plates) =  $4\frac{1}{2}T + 2$  in. and D (Bars) =  $6\sqrt{T} + T$  where T =casting thickness. Chills placed between risers promote soundness for a distance equal to 2D (1D in each riser direction).

Similar studies made with high conductivity copper and water cooled chills show that the added improvement in soundness obtainable by such chills is negligible.

The basic effects of chill behavior as related to heat flow and solidification mode are discussed.

## Introduction

Previous reports1,2 have shown that the distance which plate and bar castings of uniform thickness can be fed to soundness is a function of the gradients produced by the heating effect of the risers and the chilling effects of casting edges. It was shown also that the different geometries of plate and bar sections influence feeding-barely measurable longitudinal temperature gradients of 1 to 2 F per in. are sufficient to provide feeding for plates while gradients of at least 6 to 12 F per in. are necessary for feeding bar sections. The extent of the gradients in solidifying plate sections up to 4 in. in thickness is such that a plate can be fed to complete soundness when its length, as measured from the riser perimeter to the edge, is equal to 41/2 times the plate thickness (41/2T). Of this distance 21/2 times the plate thickness is made sound as a result of gradients originating from the casting edge. Bar sections between 2 and 8 in. in thickness can be fed for a distance equal to 6 times the square root of the bar thickness  $(6\sqrt{T})$  and the sound zone resulting from casting edge gradients varies from 4 in. for the 2-in. bar to 11 in. for the 8-in. bar.

Later work³ showed that a bar casting completely surrounded with chills will complete its solidification in 20 to 25 per cent of the time required for a similar casting in a sand mold. This suggested that chills suitably placed at casting edges would increase the rate of solidification and cause an increase in the expanse of a bar or plate which could be made sound with a single riser.

Plate and bar castings were accordingly made as in previous studies with the riser located at one end but

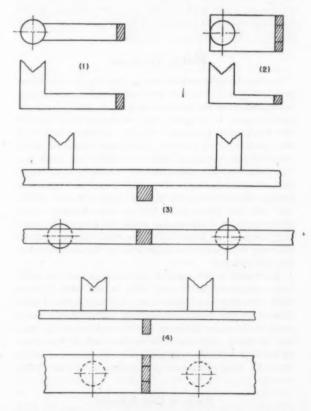


Fig. 1—Chill locations in experimental plate and bar castings.

Opinions set forth in this paper are those of the authors and do not necessarily reflect the opinions of the Navy Dept.

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Metallurgist and **Head, Metal Processing Branch, Metallurgy Division, Naval Research Laboratory, Washington, D.C.

with the sand at the casting end opposite the riser replaced with medium carbon steel chills. Chill tests were also made on castings which were sufficiently long to require two risers, in which case the chill was located on the drag surface of the casting midway between the risers with the aim of developing gradients which would extend toward each riser. The effectiveness of high conductivity chills such as copper was also explored. The soundness of the resultant castings was determined by means of transverse radiography of 1-in. thick strips removed from the center of the castings.

In selected cases thermal studies were made on the chill material and the solidifying casting in order to relate the degree of soundness to the gradient conditions. The thermometry techniques employed are described in detail in earlier papers.1,2,3 Briefly, the casting temperatures were determined by means of suitably insulated and protected Pt, Pt-Rh (13%) thermocouples inserted into the mold cavity so that their junctions were at the casting center line. Chill temperatures were obtained by means of chromelalumel couples, flash welded into the bottom of 3/32in. diam holes which had been drilled at the desired distances from the interface along the chill center line. The temperatures were recorded on 16-point automatic recorders. All castings were poured at 2950 F±25 F with induction melted steel containing 0.25 to 0.35 per cent carbon, 0.50 to 0.80 per cent manganese, and 0.30 to 0.50 per cent silicon. Final deoxidation was accomplished by the addition of 0.10 per cent of aluminum to the ladle.

#### Molding Techniques

Various lengths of bar castings having cross-sections of 2 x 2, 3 x 3, 4 x 4, 6 x 6 and 8 x 8 in. and plate castings having thicknesses of 1, 2, 3 and 4 in. with widths equal to 5 times their thickness were cast in the horizontal position. Earlier studies indicated that plates of 5 to 1 width to thickness ratios behaved essentially as plates of infinite width—i.e., solidification from the side edges did not interfere with longitudinal solidification. The bars were fed by risers having diameters equal to  $1\frac{1}{2}$  times the bar thickness and the plates were fed by risers having diameters equal to 3 times the plate thickness; the riser heights were at least  $1\frac{1}{2}$  times their diameters. In all cases the risers were sufficient to eliminate underriser shrinkage.

The chills at the ends of the castings had the same cross-section as the casting, Fig. 1, and were located such that they were extensions of the castings. Chills located at the center of castings between two risers were of the same shape as in the single riser castings, but in these cases the chill surface which had been in contact with the casting end was now in contact with the drag surface of the casting as is also shown in Fig. 1.

#### Study of Chill Behavior

For purposes of this investigation it was desired that the chill would promote more extensive directional solidification by developing a continuing rapid rate of heat removal throughout the entire solidification period of the casting. Momentary and localized effects such as obtained by the conventional use of chills at hot spots where the chill is only required to balance out the hot spot condition were not considered.

Previous investigations³ have shown that the rate of heat removal which is maintained by a chill is determined primarly by the thermal capacity of the chill, which is a function of its thickness, and secondarily by the formation of an air gap at the chill metal interface, which acts as a barrier to heat flow. Thin chills of inadequate thermal capacity become saturated with heat before the casting completes its solidification, after which time, from a thermal standpoint, the chill may be considered essentially as an extension of the casting. Very thick chills of over-adequate heat capacity do not become appreciably heated at their outer regions during the freezing of a casting; while such chills maintain a continuing rapid rate of heat removal and serve the purpose for which they are intended, generally they are not practical. The practical and technically adequate chill is characterized by reaching and maintaining approximate temperature equilibrium at the interface while the outer regions continue to rise in temperature during the intermediate and late stages of solidification. A continuing, rapid rate of heat removal is thus maintained by such a chill.

The first experiments were aimed at determining (1) the minimum chill thickness which could be considered adequate, and (2) the existence and effect of

TABLE 1—SUMMARY OF RADIOGRAPHIC DATA FOR DETERMINATION OF ADEQUATE END CHILLS

Heat No.	Casting Thickness, In.	Chill Thickness	Distance to be Fed, In.	Sound End Zones, In.	Remarks
		PL	ATES		
48	1	1T	10	31/2	Shrinkage
72	1	2T	10	31/2	Shrinkage
101	2	1/4 T	14	6	Shrinkage
101	2	1/2T	14	61/2	Shrinkage
101	2	IT	14	71/2	Shrinkage
53	2	1T	14	71/2	Shrinkage
59	2	2T	14	71/2	Shrinkage
81	4	1/2T	20	10	Shrinkage
57	4	ÎT	22	12	Shrinkage
65	4	1T	24	13	Shrinkage
77	4	2T	20	_	Sound
57	4	2T	22	13	Shrinkage
		В	ARS		
30	4 x 4	1/4 T	24	81/2	Shrinkage
30	4 x 4	1/2T	24	101/2	Shrinkage
55	4 x 4	1/2 T	20	11	Shrinkage
30	4 x 4	1T	24	101/2	Shrinkage
30	4 x 4	11/2T	24	11	Shrinkage
102	6 x 6	1/4T	22	11	Shrinkage
102	6 x 6	1/2 T	22	131/2	Shrinkage
83	6 x 6	1/2 T	22	13	Shrinkage
83	6 x 6	1T	22	13	Shrinkage
31	8 x 8	1/4T	32	131/2	Shrinkage
37	8 x 8	1/2T	32	17	Shrinkage

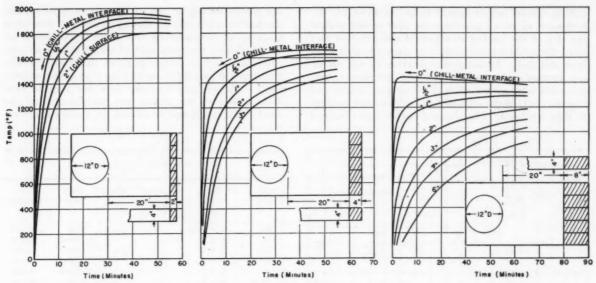


Fig. 2—Temperature cycles at various distances from the interface in ½T (left), 1T (center) and 2T (right) steel chills located at the end of 4-in. thick plate castings.

an air gap. To determine adequate chill sizes, chills having thicknesses between 1/4 and 2 times the casting thickness (1/4T and 2T) were placed at the ends of plates and bar castings having lengths in excess of the critical size which could be made sound. Chill effectiveness was evaluated by measuring the length of the sound end zones from transverse radiographs made of the castings. Thermal measurements were also made in one series of chills of varying thicknesses which had been indicated radiographically to range from inadequate to over-adequate. At this point in the experimental work, the effect of the air gap was not known. To eliminate possible variations due to air gap formation, 1/4-in. diameter steel rods, threaded over their entire length, were screwed into each chill so that about an inch of the rod protruded into the mold cavity thus effectively bolting the chill in intimate contact with the casting following the formation of a solid skin.

The radiographs indicated that 1/2T chills were fully effective for bars and 1T chills were similarly adequate for plates. Chills of greater thicknesses caused no significant increase in the length of the sound end zone over the length noted for the adequate chills, Table 1.

It should be noted that the subject application requires chills of greater thickness than would be required for uses entailing a short time effect only. Troy⁴ has shown that when chills are used on bosses to prevent localized shrinkage, smaller chills are adequate since they must be effective through only a portion of the solidification process. In fact, as Troy points out, oversize chills in such cases can actually be deleterious since they may block feeding to regions in the casting beyond the chills.

The thermal cycle studies, Fig. 2, which were made on ½T, 1T and 2T chills at the ends of 4-in. thick plate castings corroborated the radiographic findings. It can be noted from Fig. 2 that the interface of the

overadequate 2T chill reaches a maximum temperature within 3 min and thereafter its temperature falls gradually. The temperature 1/2 in. from the interface also begins to decline before the end of solidification, while at greater distances from the interface temperatures are constantly rising. These temperature cycles indicate that at late stages of solidification the remote regions of this chill absorb heat from interface regions at a faster rate than the interface receives it from the casting. All locations in the 1/2T chill reach a maximum temperature between 40 and 50 min after pouring indicating heat saturation at this point. During the last 15 min of solidification this chill maintains an essentially constant thermal condition indicating that it is passing but not absorbing heat; this chill thus acts as an extension of the casting rather than a heat extracting agent during this period. The 1T chill, which was shown to be adequate by the radiographic tests, reaches a maximum temperature at its interface in approximately 30 min, after which the interface temperature remains essentially constant during the remainder of the solidification process. The chill regions beyond the interface continued to rise in temperature until the end of solidification, indicating continuing absorption of heat by the chill in addition to heat passage.

A limited number of radiographic studies were made to determine if the chills used in the single-riser plate-systems were adequate for use in plate-systems with chills between two risers. A small and possibly not significant difference was noted between the formerly described adequate 1T chill and the overadequate 2T chill. It was accordingly decided as a precaution to use chills of twice the thickness used for the single riser systems. This procedure was followed for both plates and bars for all tests of chills placed between risers.

Normal molding procedures for horizontal castings with no provisions for maintaining intimate chill casting contact would appear to be favorable for the formation of an air gap since the riser would anchor the end of the casting to which it was attached and cause the cooling and contracting casting to pull away from the chill. To determine whether or not an air gap formed under normal molding procedures, temperatures were measured at the interfaces of free and secured chills at the ends of  $4 \times 4 \times 20$ -in. bar castings as they solidified. The thermal curves thus obtained

are shown in Fig. 3. An air gap forms between the unattached chill and its casting at about 2 min after pouring as indicated by the sharp temperature reversal; during the remainder of the solidification process the interface temperature of this chill remains between 150 and 200 F lower than that of the secured chill. The air gap however does not impede heat flow sufficiently to impair the effectiveness of the chill. This was shown by casting, in duplicate, several

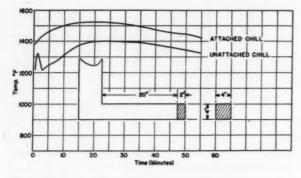


Fig. 3—Temperature cycles at the interface of secured and unsecured chills located at the end of 4x4-in. bar castings.

TABLE 2-SUMMARY OF RADIOGRAPHIC TESTS ON END CHILLED PLATES (1T CHILLS)

Heat No.	Plate Thickness, In.	Distance to be Fed, In.	Sound End Zone, In.	Chill Type	Remarks
48	1	5	4	Secured	Sound
48	1	6	_	Secured	Sound
48	1	7	_	Secured	Sound
59	1	7	_	Free	Sound
48	1	8	31/2	Secured	Shrinkage
59	1	8	31/2	Free	Shrinkage
79	1	8	3	Free	Shrinkage
48	1	9	3	Secured	Shrinkage
59	1	9	4	Free	Shrinkage
79	1	9	31/2	Secured	Shrinkage
72	1	10	31/2	Secured	Shrinkage
48	1	10	31/2	Secured	Shrinkage
79	1	10	4	Free	Shrinkage
88	1	12	41/2	Secured	Shrinkage
		Aver.	31/2		0
58	2	10	_		Sound
58	2	11	_		Sound
58	2	12	-		Sound
53	2	12	8		Shrinkage
53	2	14	71/2	Secured	Shrinkage
101	2	14	71/2	Secured	Shrinkage
59	2	14	71/2	Secured	Shrinkage
59	2	16	81/2	Secured	Shrinkage
59	2	18	71/2	Free	Shrinkage
	1	Aver.	71/2		
75	3	15	_	Free	Sound
93	3	15	_	Secured	Sound
73	3	17	91/2	Secured	Shrinkage
72	4	20	_	Free	Sound
77	4	20	_	Secured	Sound
77	4	20	-	Secured	Sound
57	4	22	12	Secured	Shrinkage
65	4	24	13	Secured	Shrinkage
58	4	26	14	Secured	Shrinkage
		Aver.	13		

TABLE 3-SUMMARY OF RADIOGRAPHIC TESTS ON END CHILLED BARS (1/2T CHILLS)

Heat No.	Bar Size, In.	Distance to be Fed, In.	Sound End Zone, In	Туре	Remarks
52	2 x 2	9	_	Free	Sound
40	2 x 2	9	-	Secured	Sound
52	2 x 2	10	-	Free	Sound
40	2 x 2	10	_	Secured	Sound
35	2 x 2	10		Secured	Sound
32	2 x 2	10	-	Free	Sound
52	2 x 2	11	51/2	Free	Shrinkage
40	2 x 2	11	5	Secured	Shrinkage
35	2 x 2	11	51/2	Secured	Shrinkage
32	2 x 2	11	41/2	Free	Shrinkage
52	2 x 2	12	51/2	Free	Shrinkage
40	2 x 2	12	5	Secured	Shrinkage
35	2 x 2	12	41/2	Secured	Shrinkage
32	2 x 2	12	5	Free	Shrinkage
35	2 x 2	13	51/2	Secured	Shrinkage
32	2 x 2	13	5	Free	Shrinkage
35	2 x 2	14	41/2	Secured	Shrinkage
32	2 x 2	14	41/2	Free	Shrinkage
34		Aver.	5	rice	Sillinkage
85	3 x 3	12	_	Free	Sound
74	3 x 3	12		Secured	Sound
85	3 x 3	13	7	Free	Shrinkage
85	3 x 3	14	8	Free	Shrinkage
74	3 x 3	15	7	Secured	Shrinkage
74	3 x 3	18	71/2	Secured	Shrinkage
*-		Aver.	71/2	0000100	
33	4 x 4	14	_	Free	Sound
34	4 x 4	14	_	Secured	Sound
33	4 x 4	16	-	Free	Sound
33	4 x 4	18	-	Free	Sound
34	4 x 4	18	garrents .	Secured	Sound
43	4 x 4	20	111/2	Free	Shrinkage
43	4 x 4	20	111/2	Secured	Shrinkage
55	4 x 4	20	101/2	Free	Shrinkage
55	4 x 4	20	11	Secured	Shrinkage
33	4 x 4	20	101/2	Free	Shrinkage
34	4 x 4	20	111/2	Secured	Shrinkage
36	4 x 4	21	12	Free	Shrinkage
36	4 x 4	22	12	Free	Shrinkage
36	4 x 4	23	101/2	Free	Shrinkage
30	4 x 4	24 Aver.	101/2	Secured	Shrinkage
89	6 x 6	18	_	Free	Sound
89	6 x 6	19	_	Free	Sound
82	6 x 6	20	-	Secured	Sound
102	6 x 6	22	131/2	Secured	Shrinkage
83	6 x 6	22	13/2	Secured	Shrinkage
84	6 x 6	24	13	Free	Shrinkage
01	0 1 0	Aver.	13	1100	- III III Mage
42	8 x 8	24	_	Secured	Sound
45	8 x 8	25	18	Secured	Shrinkage
66	8 x 8	26	18	Secured	Shrinkage
66	8 x 8	26	17	Free	Shrinkage
37	8 x 8	32	17	Secured	Shrinkage
		Aver.	171/2		

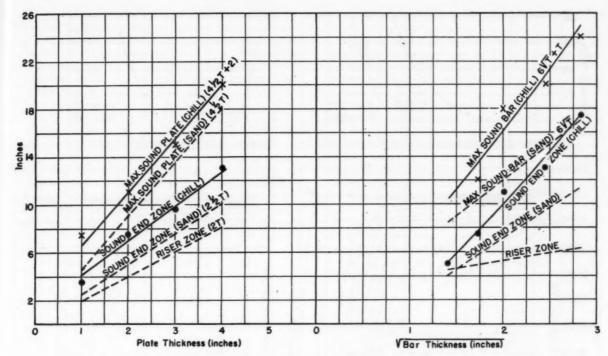


Fig. 4-Soundness relationships for chilled and unchilled plates (left) and bars (right).

plate and bar castings of different sizes chilled with adequate chills both free and secured. The radiographs of these castings showed close agreement, Tables 2 and 3.

Troy⁴ also has shown that chills placed on cope surfaces of castings where conditions are favorable for air gap formation are equally as effective as drag chills.

#### Feeding Distance with Chills at Casting Edge

A summary of the radiographic tests made on endchilled plate and bar castings which were fed with single risers are presented in Tables 2 and 3. In all cases the casting lengths were measured from the periphery of the riser to the end of the casting. In castings which were unsound there was a comparatively long sound zone adjacent to the chilled end of the casting and a short sound zone adjacent to the riser, with shrinkage located between them. As in the case of unchilled castings increasing the casting length resulted in a like increase in the length of the shrinkage zone leaving the sound zones essentially unchanged. The average lengths of the sound end and riser zones of incompletely fed castings and the maximum lengths of the castings which can be made completely sound are plotted in Fig. 4 along with similar data reported previously for unchilled castings.

The effect of chilling the edge of plate castings is to increase the length of the sound end zone by approximately 2 in. which results in approximately a 2-in, increase in the total length of plate which can be made sound. Hence, the maximum plate length in the range of 1 to 4-in, thickness which can be completely fed with this chilling and risering practice is

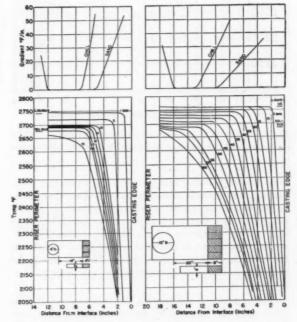


Fig. 5—(Bottom) Gradients along the centerplane of chilled 2-in. and 4-in. thick plates and (top) Effect of chills on the gradient slopes at the time of complete solidification of the various positions.

represented by the empirical formula, D max =  $4\frac{1}{2}$ T + 2 in., where T equals casting thickness in inches.

Similar data for end-chilled bar castings, Fig. 4, indicate that geometry differences cause the bar cast-

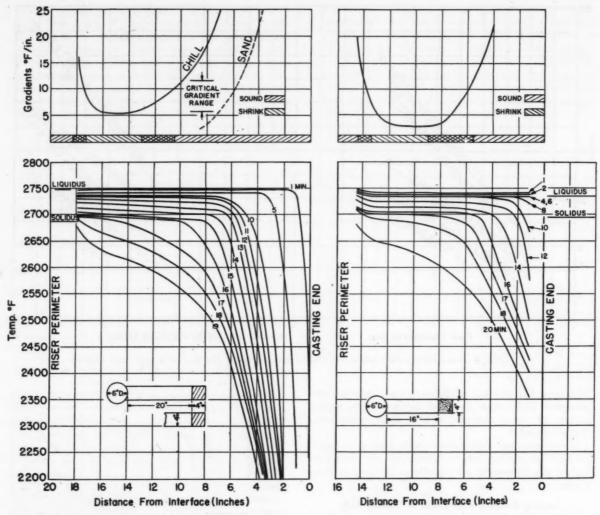


Fig. 6—(Bottom) Gradients along the centerline of chilled and unchilled bar castings and (top) Effect of chills on the

gradient slopes at the time of complete solidification at the various positions.

ings to react differently to chilling than do plate castings. The sound end zone in bars of various thicknesses is not increased by a fixed amount as in the plates but by a distance which is more nearly equal to the casting thickness. As with the plates the increased end effect is reflected in the total lengths of bars which can be made sound, thus making the critical bar length approximately equal to D max =  $6\sqrt{T} + T$ .

The relation of the feeding distances in end chilled plate and bar castings to the thermal gradient conditions is shown in Figs. 5 and 6. The bottom graphs illustrate the gradient conditions while the upper depict the slope of the thermal gradient at the time of complete solidification (drop to solidus temperature) of various points along the casting center line. The thermal gradient conditions reported previously^{1,2} for unchilled castings are also presented for comparison.

The gradient curves show that the critical gradient conditions required for soundness are the same as determined previously for unchilled castings. The previous studies indicated that shrinkage develops in plates when the length of the near-isothermal (1 to 2 F/in. gradients) region at the solidus temperature is greater than 1T. It is observed, Fig. 5, that the 2-in. thick, 14-in. long plate shows a near-isothermal region in excess of 1T; this plate developed shrinkage. The 4-in. thick, 20-in. long plate represents a case of the maximum distance which can be fed to soundness; it is observed that the limiting 1T length of near-isothermal region is developed in this plate. Figure 6 shows that the gradient slopes in the 20-in. long, 4 x 4-in. bar, which was shown to contain slight shrinkage by radiographic tests barely reach the minimum value of 6 to 12 F/in. reported to be necessary for soundness in bar castings.

The added feeding distance obtained by the use of chills results from the movement of the edge gradients toward the riser for greater distances before these fall to the limiting condition. It can be noted that in the chilled 2 and 4-in. plate castings, Fig. 5, the crit-

ical gradient condition extends into the castings for distances approximately 2 in. greater than in the unchilled castings. Likewise the minimum gradient at the solidus temperature of 6 to 12 F/in. is shown to be reached, Fig. 6, in the chilled bar at a distance from the casting edge which is approximately 5 in. greater than for the unchilled bar.

## Feeding Distance with Chills Between Risers

It has been shown in previous reports that risers promote soundness for a fixed distance regardless of the presence or absence of casting edge effects. Moreover, this distance is relatively short, amounting to 2T in the case of plates and varying from 2T to 0.8T for bars as the section size increases from 2 to 8 in. Thus, regions of castings which are located between risers and therefore do not benefit from casting edge effects can be fed only for distances equal to the sum of the contributions of the two risers.

The introduction of a chill at the midpoint between two risers develops an artificial casting edge condition which results in a marked increase in the distance which can be fed, Table 4. Chills located between risers in effect separate a long casting into two shorter castings and create soundness in each direction from the chill for a distance essentially equal to that produced by a chill located at the end of a casting of like cross-section. Thus while two risers can feed a 3-in. thick unchilled plate to complete soundness for a maximum distance of 12 in. (6 in. per riser), a chill located between them permits complete soundness to be obtained when they are separated by as much as 30 in. (15 in. per riser). Similarly two risers on an unchilled 3 x 3-in. bar can be separated by only 10 in. (5 in. per riser) if shrinkage is to be avoided but a chill midway be-

TABLE 4—SUMMARY OF RADIOGRAPHIC TESTS ON CENTER CHILLED PLATES AND BARS (2T CHILLS FOR PLATES AND 1T CHILLS FOR BARS

		ng Distance	Per	Sound	Remarks
	In.	Fed, In.	Riser, In.	Zone, In.	
			PLA	TES	
63	2	14	7	-	Sound
67	2	20	10	deman	Sound
91	2	20	10	_	Sound
94	2	22	11	_	Sound (Knock off Risers)
100	2 2	22	11	-	Sound
100	2	22	11	-	Sound
90	2	24	12	14	Shrinkage
89	2	28	14	151/2	Shrinkage
69	2	28	14	15	Shrinkage
103	2	28	14	17 '	Shrinkage
93	3	30	15	_	Sound
96	3	33	161/2	18	Shrinkage
			ВА		
64	2 x	2 16	8	K3	Sound
91	2 x	_	9	_	Sound
74	2 x	-	10	11	Shrinkage
89	2 x	_	10	81/2	O O
93	3 x	3 24	12	_	Sound

tween them permits this distance to be increased to 24 in. (12 in. per riser).

Figure 7 shows the thermal effect of a chill between two risers separated by 24 in. on a bar casting having a cross-section of 3 x 3 in. An essentially isothermal region is changed to a region of steep gradients.

It is fortunate that a marked increase in feeding distance is obtained by the use of chills in the case of multiple risers which normally represents the most difficult condition of feeding. The more extensive feeding distances which may be obtained by this simple use of chills should result in a net increase in casting yield inasmuch as fewer risers are required to provide the necessary coverage.

## **Evaluation of High Conductivity Chills**

The relatively small improvement in feeding distance which was obtained by the use of steel chills at casting edges prompted an investigation of more drastic chills which might be expected to intensify

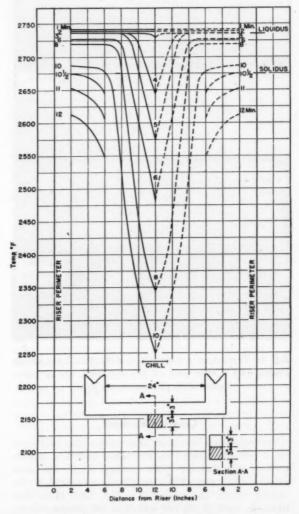


Fig. 7—Thermal gradients along centerline of centerchilled 3x3-in. bar casting.

the rate of heat transfer from the casting edge. Two types were investigated: (1) copper chills which, by virtue of the higher conductivity of the copper, would be expected to distribute the heat accepted from the castings more uniformly and rapidly through the chill thickness and therefore maintain a lower interface temperature, and (2) water-cooled steel chills which would be expected to maintain a lower interface temperature by virtue of continuous heat removal by the water.

This study was conducted on 4 x 4 x 24-in. bar castings with secured 4 x 4 x 8-in. (2T) copper chills and secured 4 x 4 x 8-in. water-cooled chills having 1/4-in. thick walls. Neither chill resulted in a significant improvement in the distance which could be made sound over that obtained by the conventional steel chill. The similarity in the conditions of soundness of the bars chilled by steel and water-cooled chills can be noted from the radiographs, Fig. 8.

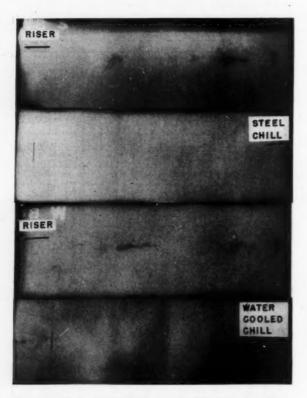


Fig. 8—Transverse radiograph of 4x4x24-in. bar castings chilled with a steel chill (2 top sections), and a water-cooled chill (2 bottom sections).

## General Discussion

Previous investigations³ have shown that chills exert a very great effect on the mode and rate of progressive solidification from castings walls. In these prior studies the chills completely surrounded the casting which was a long 7 x 7-in. bar. Heat flow from the ends of the bar which would produce directional solidification was essentially eliminated thus providing a condition of heat passage in only the

lateral or chill directions. It was observed that with chills solidification occurred in a narrow band-like fashion in contrast to a broad freezing band resulting in a mushy liquid plus solid condition through the casting section obtained with sand. Also, final solidification at the center of the chilled casting occurred in one-fifth of the time required for the sand casting. In fact the very great effect on the rate and mode of solidification from side walls prompted the present investigation on the effect of chills on directional solidification. However, the limited effect of chills when employed to intensify directional solidification in sand castings seemingly is in conflict with the previous studies. In addition the much greater effectiveness of chills when employed between risers as compared to casting extremities also may appear anomalous at first consideration. However, due consideration of the heat flow interactions which are involved in those various uses of chills shows clearly the basis for the differences.

A proper understanding of chill effects in the various applications requires exact distinction between the features of progressive (lateral) and directional (longitudinal) solidification. Progressive solidification entails wall growth which starts from mold walls and moves toward the center of the casting. Directional solidification towards risers or away from casting ends entails a gradation of rates of progressive solidification along the casting such that center line positions near casting edges freeze at a faster rate than positions near the riser-thus a "tapered" feed channel condition exists during the solidification of the casting. Figure 9 schematically illustrates the features of progressive and directional solidification. It is apparent that directional solidification is not due to extensive wall growth from the end of a casting.

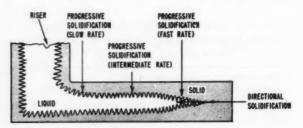


Fig. 9—Features of progressive and directional solidification.

Progressive solidification, or wall growth, from the casting end exists only during early stages of solidification. The end of this phase of solidification for the casting edge is marked by the meeting of wall growth from the side surfaces and the end surface at a common point at the casting center line. If a common mold material is in contact with side and end surfaces the meeting point is approximately ½T from the end surface. In the case of unbalanced heat flow as occasioned by the use of chills at the casting end this point is moved inward since progressive solidification is faster from the chill surface.

Following the completion of progressive solidification from the casting end, a race is established be-

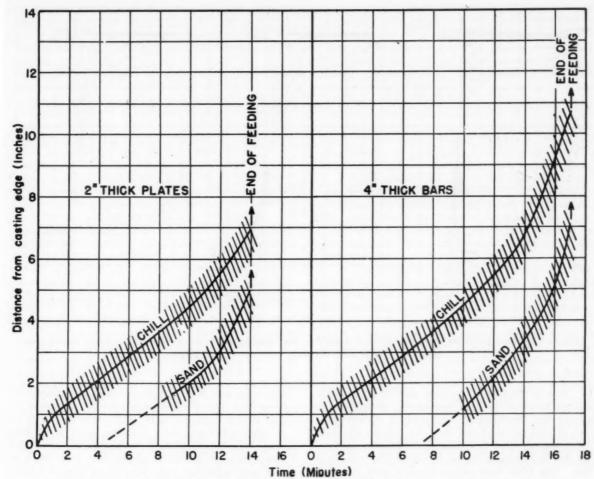


Fig. 10—Progression of the "end of freeze" point from the extremities of chilled and unchilled plates (left), and bars (right).

tween the movement of the rapid solidification zone (knee region of fast wall growth) along the casting center-line and the gradual completion of the lateral solidification cycle at positions near the riser. When lateral progressive solidification at near-riser positions nears completion the flow of feed metal from the riser is cut off and shrinkage develops.

In the various investigations of the solidification of bars and plates, the maximum distances which could be made sound were related to a minimum, critical gradient condition along the casting center line. The gradient condition implies that solidification is directional and that "tapered solidification channels" required for movement of liquid feed metal are present.

It is apparent that if chills at casting ends are to be effective, they must either (1) develop longer gradients and thereby lengthen the tapered channel condition or (2) move the tapered channel region (solidification knee) faster along the length of the casting towards the riser thus covering a greater distance before the completion of lateral progressive solidification at near riser positions. Comparison of the gradient conditions of sand cast plates and bars with those of end-chilled castings shows clearly that the

additional feeding distance results from the second of the two effects listed above. The faster movement of the solidification knee occurs primarily during the period of progressive solidification from the casting end. This sets up a distance differential in comparison to unchilled castings which is essentially retained throughout the remaining cycle of directional solidification. This may be observed from Fig. 10 which presents data of the progression of the point of complete solidification along the casting center line of representative plates and bars.

It may also be noted from Fig. 10 that during the period of progressive solidification from the casting ends the approximate five-fold faster rate of solidification observed in the side wall solidification of the chilled and unchilled 7-in. x 7-in. bar castings is in fact developed. For example, a 1-in. thick solid skin at the casting ends is developed at approximately 1½ min in the chilled 2-in. plate and 4-in. bar castings, while from 7 to 10 min are required in similar unchilled castings.

The significant features of the thermal behavior of sand and chill systems which illustrate the described mechanism of chill action are shown in Fig.

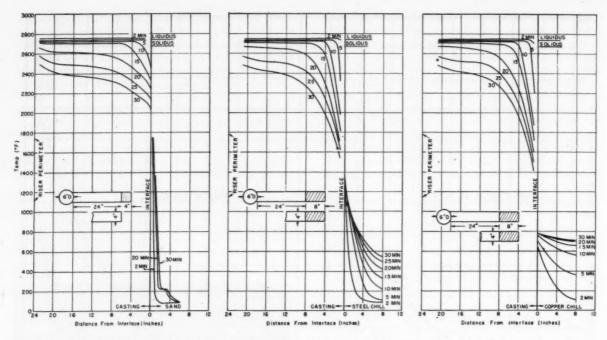


Fig. 11—Thermal gradients along centerline of 4x4x24-in. bar castings and in the mold material at the casting edge; sand (left), steel chill (center), and copper chill (right).

11. It is noted that all systems show a gradientless plateau between the riser and casting extremity signifying the development and gradual growth of a wall of uniform thickness over the plateau distance. All systems develop knee-like gradients near the casting end signifying a region of very active solidification. The knee zone moves forward with time encroaching upon the gradientless plateau with the movement being much faster for the chill systems at early times thus establishing the condition for the moderate increase in feeding distance developed by the use of chills.

It is further noted that the alternate mechanism required to increase feeding distance-i.e., the development of solidification gradients which extend deeper into the casting (wider knees), does not result from the use of chills. In fact this should not be expected inasmuch as the basic effect of chills is to decrease rather than increase the width of the solidification knee. This effect was clearly demonstrated in the studies of progressive solidification of the 7-in. x 7-in. bar castings cited previously. Thus, instead of developing a broader solidification knee it is noted from Fig. 11 that the knee region is approximately cut in half as may be observed from the fact that the knee of the 10-min gradient curve is approximately 1 in. wide for the case of the chills and at least 2 in. wide for the sand case.

The question of why the chill effect is pronounced at early stages but not at intermediate and late stages of directional solidification requires consideration of the changing nature of heat transfer conditions with time. In a recent NRL publication⁵ heat transfer calculations were made showing the cumulative amounts of the various heats (liquid and solid spe-

cific heats and heat of fusion) which are transferred from steel bars to chill walls during solidification. The studies indicated that while chills having thick or "over-sufficient" walls absorbed more total heat than chills having walls of "sufficient" thickness, this added heat was primarily specific heat from the metal already solidified; very little being heat of fusion. Thus, the lower casting interface temperature developed by the thicker chill indicates that it absorbs a proportionally greater amount of heat to accomplish a given amount of solidification than is required of the thinner chills.

It is apparent that these observations apply also to solidification from the end of a plate or bar casting since "oversufficient" chills or chills having very high conductivities have a negligible effect on the extent of soundness produced. The added heat removed by these chills during solidification is confined primarily to cooling further the already solid metal near the interface. The low interface temperature developed by a chill results in an increased heat burden on the chill which is further increased as the distance of the solidification knee zone from the chill is increased. The low interface temperatures signify that heat is in fact flowing at a faster rate from the casting to the chill but as discussed above only a fraction of this heat which is transferred is heat of solidification. At early stages of directional solidification the fraction is high and it is observed that the solidfication knee moves forward at a very rapid pace in the case of the chills as compared to sand; at later stages the fraction diminishes due to increased thickness of wall which is liberating specific heat, and the more rapid movement of the solidification knee of the chill casting gradually falls off. The relationship

in this respect of steel chills to sand is the same as that of copper and water-cooled chills to steel chills in that the lower interface temperatures results in an additional specific heat removal burden on the chills

The above provides an explanation of the ineffectiveness of higher conductivity chills and the decrease in chill effect as the chill is made to act over extensive distances at intermediate and late stages of solidification. The differences in the heat flow characteristics between casting ends and side walls provide an understanding of the difference in the efficiency of chills placed at side walls or between risers as compared to casting edges. It has been shown³ that in sand cast bars of sufficient length to eliminate axial heat flow (no end effects) the interface temperature during solidification remains near the solidus temperature. This implies a heat saturation of the sand adjacent to the casting. It is not until solidification is complete and the casting begins to cool that the interface temperature drops as the result of heat travel outward into the colder sand at a faster rate than it is being received from the casting. It was shown for example that the side wall interface of a long casting having a cross section of 7 x 7 in. was only about 40 F below the solidus at the end of solidification. Solidification and the accompanying gradient conditions from the end of a sand-cast bar, however, are entirely different. As shown in Fig. 11 the interface temperature of a 4-in. x 4-in. unchilled bar is approximately 450 F below the solidus when solidification is completed. It was determined that as soon as the cycle of progressive solidification from the end wall is completed, the temperature begins to fall at the casting end and the adjoining sand. Thereafter the end portion of the casting which has solidified acts as a heat pipe (may also be considered as a saturated chill) no longer emitting heat of solidification of its own but transporting and distributing heat of solidification from the interior reaches of the casting. This feature coupled with that of heat flow from a very large surface area (end, four adjacent surfaces and corners) places a smaller burden of heat transfer on the large volume of sand which surrounds the end portion of the casting than that which is being forced on the sand adjoining the side wals. Thus, at the side wall locations in the casting where end effects are absent only negligible gradients can exist while at the end of the casting the gradients are comparatively steep. Because of the presence of steep gradients at casting ends the effect of chills is relatively small. In the case of side walls the effect is very great since the gradients established by the sand are relatively mild and subject to considerable improvement by the application of a chill.

## Conclusions

1. A chill applied to the extremity of a simple plate or bar casting for the purpose of increasing riser feeding distance will be of an adequate size to produce the maximum improvement when its cross-section is the same as that of the casting and its thickness is 1T in the case of plates and ½T in the case of bars.

2. Empirical formulae denoting the section length in inches (D) which can be made sound by one riser with the aid of chills are D (Plates) =  $4\frac{1}{2}T$  +

2 in. and D (Bars) =  $6\sqrt{T} + T$ .

3. Chills placed at casting extremities cause only a minor improvement in length of sound zone (2 in. for plates and T for bars) over that normally present in sand castings, but when the chills are placed at locations where normal sand end effects are absent (as between two risers on a long section) an increase in the length of sound zone in excess of 100 per cent is obtained. In this case soundness is promoted for a distance equal to 2D (1D in each riser direction).

4. High-conductivity copper and water-cooled chills do not cause a measurable improvement in soundness over that obtainable with steel chills. Thermal studies show that while these chills remove heat at a faster rate than steel chills, this added heat is primarily specific heat from the solid metal rather than

heat of fusion.

5. The minimum gradient slopes found to be necessary for soundness in sand castings (1 to 2 F/in. for plates and 6 to 12 F/in. for bars) are also necessary for soundness in chilled castings. Chills, however, move the point at which the gradients fall below these critical values to a greater distance from the casting extremities, thus increasing feeding distance.

6. During the early stage of solidification (progressive) from the casting edge the rate of wall growth from the end of a chilled casting is four to five times that developed in a similar unchilled casting. During the later stages of solidification (directional) the rates are essentially the same for both sand and chill castings. The increased feeding distance obtained by the use of chills results from the distance differential established during the stages of progressive solidification from the casting edge.

#### Bibliography

1. H. F. Bishop and W. S. Pellini, "Contribution of Riser and Chill Edge Effects to the Soundness of Cast Steel Plates," Transactions, A.F.S., vol. 58, pp. 185–197 (1950).

2. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "The Con-

 H. F. Bishop, E. T. Myskowski and W. S. Pellini, "The Contribution of Riser and End Effects to the Soundness of Cast Steel Bars," Transactions, A.F.S., vol. 59, pp. 171–180 (1951).

3. H. F. Bishop, F. A. Brandt and W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," Transactions, A.F.S., vol. 59, pp. 435–450 (1951).

4. E. C. Troy, "Controlled Solidification by Gradient Heat Ex- traction," Steel Foundry Facts, Feb. 1943, pp. 2-13.

5. H. F. Bishop, F. A. Brandt and W. S. Pellini, "Solidification Mechanism of Steel Ingots," *Journal of Metals*, AIME, Jan. 1951.

#### DISCUSSION

Chairman: C. B. Jenni, General Steel Castings Corp., Eddystone, Pa.

Co-Chairman: DALE HALL, Oklahoma Steel Casting Co., Tulsa, Oklahoma.

Recorder: L. H. Hahn, Sivyer Steel Castings Co., Chicago. J. B. Caine: ¹ The authors have explored another phase of steel risering and have placed it on a firm factual foundation rather than the shaky opinions it rested on in the past.

The extension in feeding distance of single risered castings is disappointingly low, and it is doubtful if chilling such castings will be practical except in unusual cases. However it would seem "that if the casting section is circular and there is no casting end, a strategically placed chill will double the feeding distances" even with one riser. Do the authors agree with this

extrapolation of thought?

MR. BISHOP: A circular section could be considered a plate without edges. Mr. Caine's concept is correct.

S. A. ELIOT: ² We poured castings using chills which we buried

1/4 to 1/6 in. in the sand. There was a considerable improvement in solidification. We used zircon instead of silica sand; this indicated a higher heat transfer by higher chill temperature. Would pouring "up-hill" cause any difference? Would you recommend use of tapered chills along the surface of the casting?

MR. BISHOP: We have not tried this idea, but it seems worth-

while as in casting bronze bushings.

R. A. Willey: ⁹ Would you recommend use of chills between spokes of a six-spoked gear?

Mr. Bishor: This would be satisfactory with alternate risers plus head on the hub. T-section feeding can be facilitated by chilling if within critical length.

Consultant, Cincinnati.
 General Electric Co., Schenectady, N. Y.
 Commercial Steel Casting Co., Marion, Ohio.

## INTERNAL POROSITY IN GRAY IRON CASTINGS

By

John C. Hamaker, Jr.,* William P. Wood,** and Franklin B. Rote†

#### ABSTRACT

The influence of phosphorus and alloys on internal shrinkage porosity in a 3.25 per cent T.C., 2.25 per cent Si gray iron was determined by precision density measurements on unfed 4-in. cube castings. Molybdenum additions produced the greatest internal shrinkage tendency, concentrating it in visible porous regions, while chromium additions resulted in more widespread microporosity of lesser magnitude. Nickel reduced the microporosity of plain iron slightly but had little effect on the shrinkage tendencies of molybdenum or chromium irons. Visible porosity appeared in a Ni-Cr-Mo iron when the phosphorus content was raised from 0.04 to 0.06 per cent, and increased steadily in magnitude to over 2 cc per 4-in. cube at 0.18 per cent P. Variations in carbon or silicon of 0.25 per cent above or below the base content had a negligible effect on the magnitude of shrinkage voids.

From an analysis of cooling curves and the variables affecting internal porosity, it was concluded that the liquid and solidification contraction of the low melting network of phosphide eutectic in the solidified iron could provide the only logical mechanism for this type of void formation. Metallographic and analytical studies revealed that both molybdenum and chromium segregate strongly to the phosphide eutectic, approaching saturation contents of approximately 8.3 per cent Mo and 2.4 per cent Cr in the eutectic. This segregation greatly increases the volume of phosphide eutectic and changes the phosphide solidification from the stable Fe-Fe₃P graphite form to the metastable Fe-Fe₂P-Fe₃C form, with a drastic increase in its contraction. Nickel shows no tendency for segregation to the phosphide eutectic and does not change its properties appreciably.

Measurements of the liquid, solidification, and solid contractions of synthetic phosphide eutectic compositions permitted the calculation of theoretical 4-in. cube void volumes produced by phosphide contraction. These values gave an excellent quantitative check with the measured void volumes.

Most forms of unsoundness in gray iron castings have been quite completely catalogued as to their source and means of elimination. The foundryman has learned to recognize gas defects due to improper melting, poor gating design, or insufficient core or mold venting, and take appropriate steps for their prevention. Similarly, shrinkage defects due to the liquid and solidification contraction of the iron-carbon eutectic have been eliminated by improved gating, risering, or casting design based on the concept of controlled directional solidification.

However, with the adyent of high test irons, increased use of alloy additions, and more demanding soundness specifications, a new type of defect has come into recognition. Commonly called internal porosity or sponge, it will unexpectedly appear as a region of small voids in the oil lines of an engine block or at critical points in a pressure casting or highly stressed part, causing leakers and fatigue failures in service. Most baffling has been the fact that it will frequently occur throughout a foundry in a number of castings which have been satisfactorily made for years, and with no marked change in the routine analysis of the iron.

Industrial and laboratory investigations^{1,2,3,4,5} soon revealed that this type of defect was distinctly different from any previously encountered. Although the voids had a well defined dendritic structure typical of shrinkage, they could not be removed by improvements in gating and risering which would normally eliminate shrinks or piping due to iron-carbon eutectic contraction. In order to prevent void formation, it was found necessary to employ oversize risers which would move every thermal center or hot spot completely out of the casting. This would be extremely impractical on most engineering castings with their numerous bosses and changes in section. Normal risering practice involving the simple feeding of liquid iron to a heavy section merely shifted the porous region a very short distance from its original position, and frequently increased its size.

Further studies indicated that internal porosity appears when the phosphorus content exceeds a definite level and seems to be greatly magnified by certain alloy additions or build-up of alloy residuals in the return scrap. It was found to be quite independent of carbon and silicon content over wide ranges, although very low carbon irons (below 2.90 per cent) showed a reduced tendency for void formation.

The correlation with phosphorus would immediately indicate that internal porosity might be associ-

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ated with the solidification of the low melting phosphide eutectic. However, the expansions observed by Turner⁶ in his contraction studies of high silicon "phosphoretic" pig iron and the appearance of small beads of high phosphorus liquid exuded from the shrink pipes of castings after primary solidification have always led metallurgists to believe that the phosphide eutectic expands on solidification and thus would tend to increase rather than decrease internal soundness. Consequently, the source of these voids and their association with phosphorus and certain alloys have never been satisfactorily explained.

Accordingly, a research investigation was undertaken to determine the mechanism for this type of void formation and more accurately fix the composition levels at which it will appear. These findings were then checked by further studies in several production foundries.

#### I. Internal Porosity Study

The bottom-gated, unfed 4-in. cube shown in Fig. 1 was selected as a standard test specimen for this investigation. In industrial studies, the presence of voids in the center section of the cube has correlated closely with the appearance of internal porosity in a wide range of castings, and the specimen has been used as a standard quality test in several foundries. The cube is cast entirely in baked core sand to avoid the external shrinkage or piping variations due to mold wall movement observed by Bohm,⁷ Sanders, and Sigerfoos⁸ in their studies of green sand.



Fig. 1—Photo of typical 4-in. cube casting.

The base compositions chosen for this study were those normally employed in the production of automotive castings, namely 3.00-3.40 T.C., 2.10-2.40 Si, 0.75 Mn, and 0.10 S. These were modified by single and combined additions of up to 2.00 per cent nickel, 1.00 per cent molybdenum, and 1.00 per cent chromium, and the phosphorus content was varied between 0.04 and 0.30 per cent. The heats were melted in a 250-lb Detroit indirect arc furnace and a 200-lb induction furnace from charges of wash metal obtained by melting 100 per cent steel in the cupola, ferro-85silicon, ferro-80-manganese, and coke. After superheating to 2850 F, the iron was inoculated with 0.50 per cent of the total silicon to promote uniformity of graphitization, and the cubes were poured at 2600 F as measured by a platinum-platinum rhodium immersion thermocouple.

In order to obtain a quantitive comparison of the porous regions produced in the center of the cube by

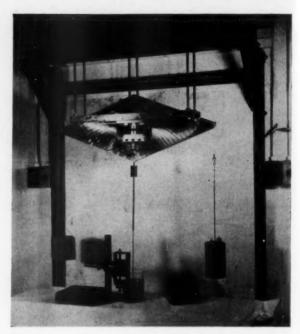


Fig. 2—High capacity precision balance for void volume determinations.

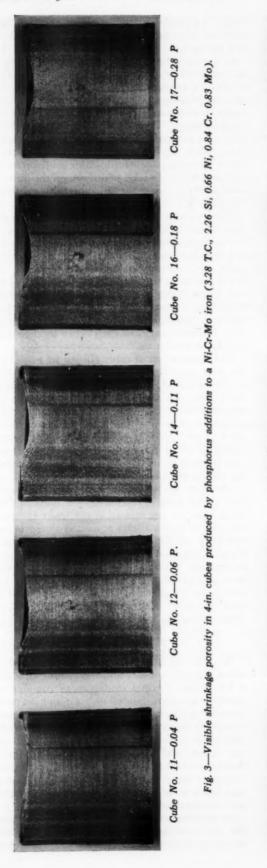
various phosphorus and alloy contents, a high capacity precision balance, shown in Fig. 2, was constructed. This balance, based on the flexure plate principle and capable of weighing 16 lb to 5 mg, was devised and employed by Timmons⁵ in a previous investigation of internal porosity in 4-in. cubes. The total volume of internal voids may be simply evaluated by Archimede's principle. The sandblasted cube is weighed in air and water to determine its volume, and four 1-in. cubes cut from the bottom corners and deeply ground to remove any chill layer are similarly weighed to determine the density of sound iron in the cube. Dividing the weight of the entire cube in air by the density of sound iron gives the theoretical volume of the 4-in. cube if it were perfectly sound, and subtracting this value from its actual volume gives the void volume.

After weighing, the total volume of the top sink or pipe was measured by filling it level with water, the cube was sectioned along the vertical plane of symmetry for visual examination, and metallographic specimens were cut from various points to study structures and microporosity.

## Results

The void volumes, visible porosity, and microporosity obtained in representative series of cubes are summarized in Table 1, and photographs of some corresponding cube sections are shown in Figs. 3 and 4.

It will be noted that Cube No. 11 (Table 1), which exhibited no visible or microporosity, gave a computed void volume of 0.35 cc. This may be explained by the slight variation in density of sound iron produced by the difference in cooling rates between the corner cubes and the interior of the casting. Metal-



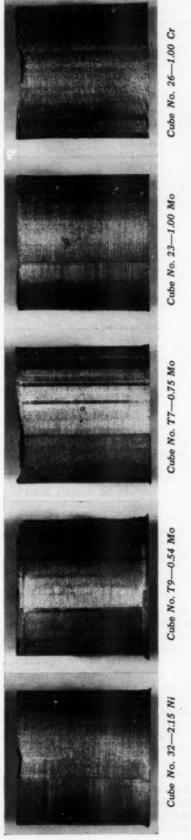


Fig. 4—Visible shrinkage porosity in 4-in. cubes produced by alloy additions to a 0.13 P base iron (3.25 T.C., 2.25 Si, 0.13 P).

Table 1 — Void Volumes, Visible Porosity, and Micro-Shrinkage in 4-in. Cubes Containing Various Phosphorus and Alloy Additions (Base Composition: 3.24-3.32 T.C., 2.23-2.29 Si, 0.77-0.81 Mn, 0.09-0.11 S)

					Void	Visib	ole*
Cube					Vol.,		
No.	% P	% Ni	% Cr	% Mo	CC	sity	Voids
11	0.04	0.66	0.84	0.83	0.35	0	None detected
12	0.06	0.66	0.84	0.83	1.15	5	Many at shrink center
13	0.08	0.66	0.84	0.83	1.31	7	25
14	0.11	0.66	0.84	0.83	1.53	7	91
15	0.14		0.84				99
16	0.18	0.66			2.29	9	**
17	0.28	0.66	0.84	0.83	2.13	10	**
21	0.13		(Plain)		0.30	0	1 large, 3 small
22	0.13			1.50	1.96	10	Many at shrink center
23	0.13			1.00	1.51	10	29
24	0.13			0.46	0.49	0	3 large, 5 small
26	0.13		1.00		1.57	2	Many, scattered
27	0.13		0.50		0.65	0	5 large, 7 small
31	0.13	1.00			0.32	0	1 large, 2 small
32	0.13	2.15			0.13	0	2 very small
33	0.13	2.00		0.75	1.43	7	Many at shrink center
34	0.13	2.00	0.25	0.75	1.52	7	**
35	0.13	2.00	0.75		1.15	1	Many, scattered
36	0.13	2.00	0.75	0.25	1.23	2	. "
37	0.13	2.00	0.50	0.50	1.07	2	29

Estimate of relative magnitude from visual examination;
 1 = smallest, 10 = largest.

• Observed in polished center section 1/2-in. square.

lographic examination revealed that the first Ni-Cr-Mo series, the straight Mo irons, and the balanced irons in the last series all contained about equal quantities of ferrite in the center while the corners were fully pearlitic. Therefore, the computed void volumes for these cubes should be directly comparable and may be assumed to be approximately 0.35 cc greater than the actual volumes. The straight nickel irons, on the other hand, exhibited less section sensitivity, necessitating a smaller correction factor, while the 1.00 per cent Cr cube contained a few free carbides and undoubtedly required a larger correction.

The first Ni-Cr-Mo series clearly illustrates the consistent increases in internal porosity with increasing percentages of phosphorus. It is noteworthy that visible porosity persisted at phosphorus levels as low as 0.06 per cent, while no visible or microporosity appeared at 0.04 per cent P. In his study of a 0.70 Mo, 0.30 Cr, 0.16 Ni iron of the same base composition, Timmons⁵ also noted a marked change in internal shrinkage tendencies in this range, obtaining the following 4-in. cube void volumes:

% Phosphorus	Void Volume (cc)
0.015	0.35
0.036	0.43
0.073	2.12
0.104	2.62
0.123	2.96

The slight reduction in void volume as the phosphorus was increased from 0.18 to 0.28 per cent corresponds with the findings of Rote,9 who obtained a

slight, but consistent reduction in the void volume of Ni-Mo irons as the phosphorus was increased above 0.20 per cent.

In the second and third series, with the base phosphorus level set at 0.13 per cent, the plain iron showed a very small amount of microporosity, and increasing additions of molybdenum or chromium produced correspondingly greater volumes of visible and microporosity. The shrinkage in the molybdenum irons was concentrated entirely in visible spongy regions, while the chromium irons showed very little visible shrinkage and considerable widespread microporosity. Straight nickel additions, on the other hand, consistently reduced the microporosity exhibited by the plain iron.

The balanced alloy irons in the third series showed approximately the same internal porosity as would be expected from their molybdenum and/or chromium contents, indicating that balancing additions of nickel are relatively ineffective for reducing the internal shrinkage tendencies of molybdenum or chromium irons. Combined additions of chromium and molybdenum consistently produced more porosity than corresponding straight additions, indicating that these elements augment each other in their internal

shrinkage tendencies.

Additional series of cubes alloyed in a similar manner were poured from base irons ranging from 3.00 to 3.40 total carbon and 2.10 to 2.40 silicon. In every case, the size and appearance of the porous region corresponded with the phosphorus and alloy content, regardless of the base iron composition. At the 0.13 per cent P level, visible porosity consistently appeared when the molybdenum content exceeded 0.50 per cent or the chromium content exceeded 0.55 per cent in straight additions.

Although the size and shape of the top sink or pipe varied widely, extending over one-third the distance from the top surface in the lower carbon irons, the porous region always remained along the vertical centerline of the cube slightly above the horizontal plane of symmetry. Similarly, no relationship could be detected between internal porosity and the presence of exuded high phosphorus beads in the top sink, as shown in Fig. 1. The beads appeared at random on cubes ranging from completely sound to those containing a large amount of internal porosity.

#### Thermal Studies

In order to investigate possible relationships between internal porosity and the cooling conditions in a 4-in. cube, thermal studies were conducted on a number of cubes poured in conjunction with the porosity specimens. Chromel-alumel thermocouples, encased in porcelain insulators and tipped with alundum cement, were cemented into the molds at various points and connected by compensating leads through a switch box to a precision potentiometer. Successive temperature readings were taken every 0.1 min from immediately after pouring until the completion of austenite transformation.

Representative cooling curves at several points in the 4-in. cube are illustrated in Fig. 5. As might be expected in a casting of this section, a major part of

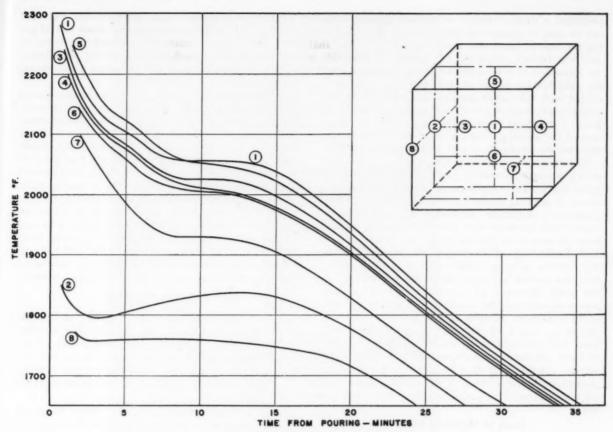


Fig. 5-Cooling curves at various points in 4-in. cube.

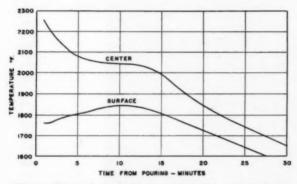


Fig. 6—Thermal gradients in cube with normal ingate area (½ sq. in.).

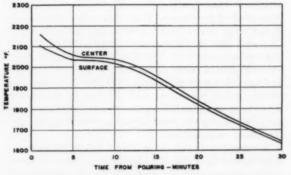


Fig. 7—Thermal gradients in cube poured simultaneously with reduced ingate area (1/4 sq. in.).

the iron remained liquid for approximately 8 min and then underwent nearly simultaneous iron-carbon eutectic solidification at various degrees of undercooling. Further studies with thermocouples placed 1/8, 1/4, and 3/8 in. in from the side, bottom, and top surfaces indicated that a solid shell is formed almost immediately on pouring. It attains a thickness of slightly less than 1/4 in. on the side surfaces, 1/4 to 3/8 in. in the bottom and corners, and does not form on the top surface until the center is midway through solidification. The temperature of the top pipe remained only 30 to 60 F below that of the

center during its eutectic solidification. Similar cubes poured from the various base iron and alloy compositions showed no significant change in these thermal gradients, thus indicating that internal porosity must not be associated with a change in cooling conditions.

This was quite conclusively confirmed by a further experiment in which a pair of cubes with the ingate area reduced to one-half its normal size were poured from a common pouring basin with a pair of standard cubes. One cube of each pair was equipped with thermocouples and the other was used for internal porosity determination. The reduction in gate area

produced a drastic change in thermal gradients, as can be seen by comparing Figs. 6 and 7. By decreasing the rate at which the mold was filled, the mold wall was heated sufficiently to prevent the immediate formation of a solid surface layer. Consequently, the solidification of the one-half gate area cube proceeded with very small temperature gradients, which might be expected to greatly reduce the magnitude of internal shrinkage voids. However, careful examination of the shrinkage cube sections from several sets revealed no significant change in the porous regions.

The influence of pouring temperature on internal shrinkage and temperature gradients was also investigated for several compositions. In all cases, an increase in pouring temperature from 2600 to 2700 F produced a detectable increase in the magnitude of voids, while a reduction to 2500 F resulted in slightly reduced void areas. These results are in accordance with the findings of Timmons,5 who concluded that the increase in void volume with pouring temperature was caused by larger temperature gradients in the mold. However, an examination of the cooling curves showed that the opposite was true in the case of the 4-in. cube. Pouring at 2700 F heated the mold surface to such an extent that a solid skin did not form and produced much smaller gradients, similar to those obtained by reducing the gate area. Conversely, pouring at 2500 F resulted in much larger gradients, the temperature of the surface layer reaching a maximum of only 1750 F during final solidification.

## Study of Meehanite Metal

This investigation was later extended to various grades of Meehanite metal, which represent specially processed irons vastly different from the automotive irons studied previously. Four-inch cubes of the same design were poured from production ladles of GA,* GC, and GE Meehanite melted in a 54-in. cupola. Various quantities of ferro-24-phosphorus and ferro-60-molybdenum were added to 50- and 90-lb ladles and the cubes were poured at 2400-2600 F as measured by an optical pyrometer and checked by an immersion thermocouple. The cubes were sandblasted and sectioned along the vertical plane bisecting the gate for visual examination.

A summary of the observations on visible porosity is presented in Table 2 and photographs of representative cube sections are shown in Fig. 8. In all cases, phosphorus additions to the various grades of Mechanite metal, which represent widely different carbon and silicon contents, produced visible voids at a definite phosphorus level. This level, 0.24-0.25 per cent P, corresponds closely with the results of West,⁴ who obtained visible porosity in a variety of castings, ranging from T-bars and drilled blocks to automotive cylinder heads, when the phosphorus content exceeded about 0.25 per cent in an unalloyed automotive iron.

Molybdenum additions to Meehanite metal reduced the "critical" phosphorus content to approximately the same level obtained earlier in the investigation. Visible porosity first appeared in GA plus 0.48 per cent Mo at 0.142 per cent P, corresponding closely with the values of 0.50 per cent Mo and 0.13 per cent P observed in the automotive irons. This alloy showed consistent increases in porosity with increasing phosphorus at levels below the critical content for unalloyed GA, and raising the molybdenum content to 0.75 per cent produced a marked increase in void area consistent with the results obtained previously. Similarly, 0.75 per cent Mo additions to GC and GE produced visible porosity in the base alloys at phosphorus levels considerably below unalloyed "critical content."

#### Influence of Risering

In order to study the effect of feeding on internal shrinkage in the 4-in cube, the patterns were modified to include a 51/2-in. diameter riser with a neck 11/2 in. in diameter attached near the center of the top surface. The neck and 1 in. of the riser proper were included in the core sand mold; the remaining riser

TABLE 2—VISIBLE POROSITY IN 4-IN. CUBES PRODUCED BY PHOSPHORUS AND MOLYBDENUM ADDITIONS TO MEEHANITE METAL

Cube No.	Meehanite Grade and Additions	% Mo*	%P*	Visible Porosity**	Pouring Temp., F
23	GA Base		0.10	0	2500
42	GA+.05 P		0.15	0	2460
43	GA+.10 P		0.20	0	2520
44	GA+.15 P		0.245	6	2500
26	GA+.18 P		0.282	8	2590
8	GA+.24 P		9.34	10	2540
25	GA+.75 Mc	0.75	0.106	0	2460
47	GA+.48 Mo	0.48	0.10	0	2520
48	GA+.48 Mo+.04 P	0.480	0.142	2	2560
49	GA+.48 Mo+.08 P	0.48	0.18	4	2570
50	GA+.48 Mo+.14 P	0.48	0.24	6	2560
24	GA+.75 Mo+.14 P	0.75	0.24	10	2510
3	GC Base		0.17	0	2580
21	GC Base		0.173	0	2470
34	GC Base		0.16	0	2440
35	GC+.05 P		0.21	0	2420
36	GC+.10 P		0.261	5	2450
22	GC+.18 P		0.364	8	2570
37	GC+.75 Mo	0.75	0.16	1	2470
19	GE Base		0.201	0	2400
46	GE Base		0.20	0	2520
31	GE+.05 P		0.240	2	2540
32	GE+.10 P		0.29	4	2500
20	GE+.14 P		0.357	9	2475
33	GE+.75 Mo	0.75	0.20	7	2460
1	Special GE Base		0.338	9	2570
2	Special GE+.14 P		0.47	10	2520

Three digit numbers are analyzed compositions; two digit numbers are calculated compositions.

West found that this porosity consistently increased in magnitude as the phosphorus was raised to 1.00 per cent. It is noteworthy that the consistent increases in the magnitude of voids with increasing phosphorus completely overshadowed any variations with pouring temperature.

Estimate of relative magnitude from visual examination:
 1 = smallest, 10 = largest.

[•] The symbols correspond to the following Meehanite metal specifications:

GA—Minimum tensile strength = 50,000 psi GC—Minimum tensile strength = 40,000 psi

GE—Tensile strength = 25-30,000 psi

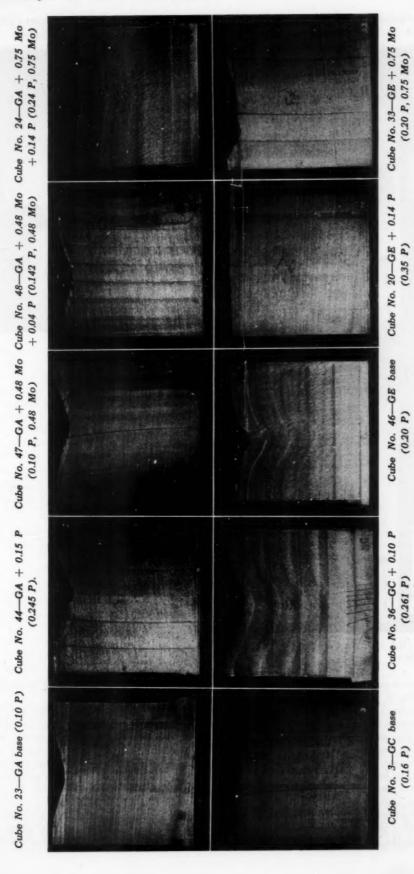
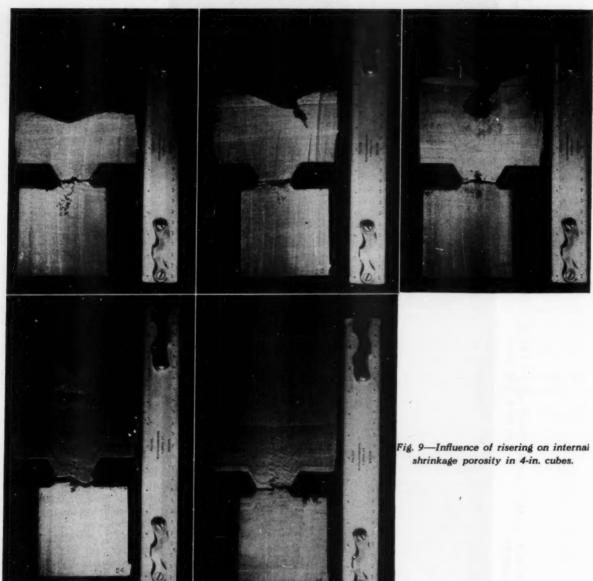


Fig. 8—Visible shrinkage porosity in 4-in. cubes produced by phosphorus and molybdenum additions to various grades of Meehanite metal.

Cube No. 51—15-lb riser (GA + 0.14 Cube No. 52—21-lb riser (GA + 0.14 Cube No. 53—26-lb riser (GA + 0.14 P + 0.48 Mo)

P + 0.48 Mo)

P + 0.48 Mo)



Cube No. 54—13-lb riser (GA + 0.14 Cube No. 16—44-lb riser (GA + 0.14 P + 0.48 Mo) P + 0.75 Mo)

height was molded in green sand. The neck and riser design may be seen in Fig. 9.

The results obtained by employing various riser heights on a porous composition are summarized in Table 3 and shown in the sequence of photographs, Fig. 9. Although a  $3\frac{1}{2}$ -in, height provided sufficient feeding to completely remove all traces of a top sink or pipe, the height had to be extended to over  $4\frac{3}{4}$  in, to move the internal shrinkage from the cube proper to the riser neck. Thus, over 31 lb of riser was required to remove the porosity from the 16-lb, 4-in, cube. It will be noted that the porous region always maintained its position at the hot spot or center of the

total mass, regardless of the feeding indicated by the pipe in the riser.

It also is interesting to note in Table 3 that the relatively small amount of porosity in the unfed Cube No. 50 (about the same as shown on Cube No. 44, Fig. 8), was greatly magnified by the presence of a riser on Cube No. 51, which was poured from the same ladle. A relatively large volume of voids is apparent in all of the risered cubes in this series poured from GA plus 0.48 per cent Mo, which would be expected to exhibit rather weak shrinkage tendencies. This effect has been observed in many industrial studies where the addition of a riser to a small porous region would actually

TABLE 3—INFLUENCE OF RISERING ON INTERNAL POROSITY IN 4-IN, CUBES

Cube*	Riser Size (Diam x Height)	Riser Weight, lb	Visible** Porosity	Distance† From Top, in.	Pouring Temp., F
24	0	0	10	2	2510
50	0	0	6	2	2560
51	51/2×21/2	15	10	11/2	2560
52	51/2×31/2	21	10	3/4	2520
53	51/6x4	26	10	3/4	2480
54	51/2×43/4	31	10	1/4	2500
16	51/2x61/2	44	10 (riser)		2510

 $^{^{\}rm o}$  Base metal for all cubes: GA+.14 P (0.24% P) . Cubes No. 24 and 16 contained 0.75% Mo; cubes No. 50 to 54 contained 0.48% Mo.

** Estimate of relative magnitude from visual examination:

1 = smallest, 10 = largest.

magnify the void area.

Similar series of cubes were poured from GA, GC, and GE with sufficient phosphorus additions to produce porosity in the unfed cube. In every case, a 5½-in. diameter riser, 6½ in. high, was sufficient to move the porosity completely into the riser as shown in the last photograph of the sequence. Despite the widely varying degrees of liquid and solidification shrinkage produced by the different carbon contents, the porous region always appeared in exactly the same position in the lower part of the riser.

The impracticability of attempting to completely remove internal shrinkage porosity from a commercial casting by risering is readily apparent from this study. It would necessitate very large risers at every center of mass or hot spot, even though these points may be adequately fed by liquid metal at the iron-carbon eutectic temperature. Porosity in long uniform sections, for example, the fillets on a circular flange, would be practically impossible to remove. Therefore, the appearance of internal porosity in a foundry should be attacked at the point of origin, namely, in the composition of the metal poured.

#### Summary

A study of internal shrinkage porosity in 4-in. cubes poured from alloyed automotive irons and various grades of Meehanite metal produced the following findings:

(1) In unalloyed gray iron castings, visible internal porosity tends to appear in every hot spot or center of mass whenever the phosphorus content exceeds about 0.25 per cent and consistently increases in magnitude as the phosphorus is raised above this level. Decreasing quantities of microporosity are obtained as the phosphorus is reduced below 0.25 per cent, appearing as only a few small voids at 0.13 per cent P. This behavior has been observed in a wide variety of castings and appears to be quite independent of carbon and silicon content, melting practice, or special processing treatments.

(2) Molybdenum additions to gray iron tend to greatly magnify the visible or microporosity at a given phosphorus content and, accordingly, reduce the "critical phosphorus level" for visible shrinkage porosity. The extent of the reduction in the critical phosphorus level is somewhat dependent on the molybdenum content. At 0.13 per cent P, visible porosity appeared with a 0.50 per cent Mo addition, increasing considerably in magnitude as the percentage was raised to 0.75, 1.00, and 1.50 per cent Mo. Molybdenum tends to concentrate shrinkage in visible spongy areas at the hot spots of a casting with little or no microporosity appearing outside these regions.

(3) Chromium additions to gray iron tend to increase visible and microporosity in a manner similar to molybdenum, but to a lesser degree. In a 0.13 per cent P iron, a small amount of visible porosity appeared at 0.55 per cent Cr, but additions up to 1.00 per cent Cr did not increase its magnitude to any great extent. However, chromium irons exhibit more widespread microporosity which tends to decrease their general soundness.

(4) Nickel additions reduce the internal porosity in plain irons to some extent, but balancing additions to molybdenum or chromium irons have little or no effect on their internal shrinkage tendencies.

(5) Combined chromium-molybdenum additions tend to augment each other in their internal shrinkage tendencies to produce voids of greater magnitude than corresponding individual additions. At lower alloy contents, this effect appears to be nearly additive, with or without balancing additions of nickel.

(6) Gray irons containing fairly large amounts of chromium-molybdenum or nickel-chromium-molybdenum may produce visible shrinkage porosity at phosphorus levels as low as 0.06 to 0.07 per cent, increasing in magnitude as the phosphorus is raised to 0.20 per cent. The porosity in these irons was completely removed when the phosphorus was decreased to 0.04 per cent, and a slight reduction in magnitude was also indicated when the phosphorus level was raised above 0.20 per cent.

(7) Normal variations in carbon and silicon content, melting practice, or processing appear to have a negligible effect on the shrinkage tendencies of molybdenum or chromium irons. However, previous investigations have indicated that very low carbon irons (below 2.90 per cent) may possess reduced shrinkage tendencies.

(8) The use of inadequate risers to eliminate internal porosity actually appears to increase the magnitude of the voids. A 51½-in. diameter riser, over 43½ in. high and weighing more than 31 lb, was required to completely remove the porosity from a 16-lb, 4-in. cube. This is accomplished by moving the hot spot or center of mass completely out of the casting—not by a simple feeding action.

(9) Since it is impractical to riser every center of mass in a casting to this extent, control of internal porosity should originate in the composition of the metal. If porosity of this type is encountered, check for high phosphorus or alloys in incoming charge materials or a build-up of alloy residuals in returns.

(10) The results of this study indicate that processed irons should exhibit greater freedom from internal shrinkage tendencies at a given phosphorus level than gray irons alloyed with molybdenum or chromium for similar properties.

The findings outlined above were checked against

[†] Deepest penetration of porosity below top surface of cube.

experience with casting defects in several production foundries. In nearly every case, internal porosity of this type was removed from critical castings by reducing the phosphorus below the critical level specified by the alloy content, and reappeared whenever the phosphorus again exceeded this amount.

## Shrinkage Concepts

A clue to the mechanism for the formation of internal shrinkage porosity may be obtained from a consideration of the general concepts of shrinkage void formation.

Internal voids may be formed whenever a mass of liquid metal is entrapped within a solidified shell. The magnitude of the voids depends on the quantity of liquid metal entrapped and its contraction relative to that of the solid shell. Although these factors are interrelated, a separate consideration of the variables affecting each should provide the clearest understand-

ing of the problem.

The quantity of liquid metal entrapped depends upon how quickly the solid shell is completely formed which, in turn, is determined by the freezing range of the alloy and the temperature gradients set up in the casting. An alloy with a short freezing range would be expected to complete the solidification of a solid shell in the shortest time. In like manner, large temperature gradients indicate that the metal adjacent to the mold surface has cooled rapidly from the pourin temperature and is prone to form a solid shell quickly.

The temperature gradients in a casting are influenced by a number of factors, including the mass of metal in the section under consideration, the thermal conductivity and specific heat of the mold material, the gating and risering, and pouring speed and temperature. A greater mass of metal serves to reduce the gradients by heating the mold material to a greater extent before solidification begins. Similarly, the proximity of gates or risers to the section in question will preheat the mold material and reduce the gradients in their direction. Mold materials of high thermal conductivity and specific heat will abstract heat more rapidly to increase temperature gradients.

The effect of pouring temperature is dependent to some extent on the section size relative to the properties of the mold material. A thin section molded in a fast heat-abstracting material would solidify rapidly before the mold had been heated to any extent. Consequently, a higher pouring temperature would tend to increase the temperature gradient by providing a hotter center layer between the rapidly solidified surface layers. On the other hand, heavy sections heat the mold considerably before they proceed with solidification and higher pouring temperatures may reduce the gradient in this case by providing more heat to the mold surface. In general, slower pouring speeds should reduce temperature gradients by preheating the mold surface more slowly and thoroughly before the casting is filled.

After the various factors outlined above have exerted their influence and a complete solid shell has formed, the liquid in the center is cut off from all further feeding and goes through its transformations independently from the rest of the casting. If the solid

shell is complete before the entrapped liquid has reached its eutectic temperature, solidification first proceeds by the separation of primary dendrites onto the surrounding surface. This produces a decrease in the volume of the liquid, withdrawing it from the top surface by gravity, and restricting further solidification to the side and bottom surfaces. If no further volume changes occurred, this would result in an internal cavity having primary dendrites extending down from the top surface.

When the liquid reaches the eutectic temperature, solidification proceeds from various nuclei throughout the melt in addition to building up on the shell walls. In the case of gray iron, the solidification of the ironcarbon eutectic may result in a contraction or exparsion, depending on the amount of graphite formed. The data of Ash and Saeger¹¹ permits the computation of this volume change from the graphitic carbon analysis. An iron containing 2.80 per cent graphitic carbon, or approximately 3.40 to 3.50 per cent total carbon with a pearlitic matrix, exhibits no volume change on solidification. Higher graphitic carbon contents result in a net expansion, while lower graphitic carbons produce a solidification contraction. The magnitude of this expansion or contraction may be approximately evaluated by the formula:

% Volume Increase = 2.0 x (% Graphitic Carbon — 2.80%)

While the entrapped liquid is undergoing the volume changes described above, the surrounding shell is also cooling and contracting in the solid state. Depending on the relative rates of contraction of the liquid and the shell, one of several phenomena may occur. If the liquid contracts more rapidly than the shell, a vacuum would be set up in the interior. Depending on the thickness of the shell and the strength of the metal at these temperatures, the hottest part of the shell with either bow in under atmospheric pressure to relieve the vacuum condition, or resist deformation to result in the formation of internal voids. In the unfed 4-in, cube, the hottest part of the shell is the top pipe or sink, since liquid contraction has pulled it away from the cooling sand before the beginning of solidification.

If the shell contracts more rapidly than the entrapped liquid, the liquid in the center would be subjected to high pressures. This phenomenon would be particularly expected in gray irons, where the shell is experiencing a relatively high rate of solid contraction while the center liquid is undergoing slow eutectic solidification with slight contraction, or even an expansion, due to graphite formation. In order to relieve this pressure, a small amount of liquid met: might be squeezed out through the weakest part of the shell wall, usually the pipe, and solidify as sma' beads adhering to the surface. This mechanism for the formation of high phosphorus exuded beads, commonly called liquation, has been strongly favored by recent investigators,5,12 and does not involve any relationship with the volume changes of the phosphide eutectic.

Liquation has been observed from a wide range of compositions, indicating that pressure conditions of this nature are a common occurrence in the solidification of gray iron castings. The presence or absence of beads on castings of similar design and composition is undoubtedly related more closely to the presence of a weak spot in the solid shell than to a drastic difference in the pressure conditions during solidification. The incidence of high pressures of this nature would, of course, remove any shrinkage voids which could have formed earlier in the solidification process.

The results of previous investigations have indicated that liquation occurs quite late in the solidification process, probably during the final stages of iron-carbon eutectic solidification in the center, or possibly shortly after its completion. Since the phosphide eutectic solidifies at temperatures approximately 350 F below the iron-carbon eutectic, it is present as a liquid network in the solid shell wall and is the first metal to be forced out by the pressure in the center. In low phosphorus irons, the small amount of phosphide eutectic is undoubtedy not sufficient to completely relieve the pressure, and any remaining iron-carbon eutectic is also forced out to appear in the beads. This would explain the variations in phosphorus content observed in analyses of the beads.

Upon the completion of iron-carbon eutectic solidification at 2000 to 2050 F, the casting has attained its final physical form and consists of solid iron interspersed by a network of liquid phosphide eutectic outlining the iron-carbon eutectic cells. As the casting cools further, the phosphide eutectic undergoes liquid contraction until it reaches its solidification temperature of 1700 to 1750 F, at which it is commonly believed, a solidification expansion occurs. Due to the temperature gradients still existing in the casting, this solidification proceeds progressively from the surface to the center, as each layer reaches the eutectic temperature.

When the phosphide eutectic has completed solidification, no more liquid is present in the casting and further cooling results in a fairly uniform contraction in the solid state. There is a slight expansion at approximately 1400 F due to the gamma-alpha transformation, but the temperature gradients in this range are usually very small and any stresses set up by the transformation would be negligible in most cases.

It can be seen from a consideration of the above analysis that a large number of factors could influence the formation of internal shrinkage voids during the solidification of a casting. All of the theories which have been proposed to explain the increase of internal shrinkage tendencies by phosphorus and certain alloys have been based on factors associated with void formation at or above the iron-carbon eutectic temperature. It has been suggested that these elements (1) increase the volume of liquid entrapped by shortening the freezing range or increasing the temperature gradients, (2) increase the solidification contraction by reducing the quantity of graphic carbon formed, or (3) increase the hot strength of the metal, thereby reducing the piping action which relieves the internal vacuum.

#### **Anomalies**

However, during the course of the investigation a number of anomalies have appeared which cannot be satisfactorily explained on the basis of shrinkage voids forming in the iron-carbon eutectic range:

(1) In both industrial and laboratory investigations, improvements in gating and risering which normally remove shrinkage defects formed at or above the iron-carbon eutectic temperature have been singularly unsuccessful with internal porosity associated with phosphorus and alloys.

(2) Internal porosity exhibits no apparent relationship to external shrinkage, consistently appearing at the hot spot or center of mass of a casting and riser system, regardless of the indicated feeding. In the unfed 4-in. cube, the voids always maintained the same position and magnitude despite wide variations in the size, shape, and position of the top sink, which remained only 30 to 60 F below the center during iron-

carbon eutectic solidification.

(3) Internal porosity does not exhibit the close relationship to temperature gradients expected of voids formed in the iron-carbon eutectic range. In fact, an increase in pouring temperature actually increased porosity in the 4-in. cube, although it reduced thermal gradients drastically. This is also indicated by the appearance of porosity at definite composition levels in a wide variety of castings, despite marked variations in thermal gradients and cooling rates.

- (4) Variations in the carbon content between 3.00 and 3.45 per cent do not produce a detectable change in the magnitude of internal voids, while low carbon irons containing up to 2.90 per cent appear to show a decreased tendency for visible porosity. Using Ash and Saeger's data11 and assuming 0.65 per cent combined carbon in the pearlite, a variation in total carbon from 3.00 to 3.45 per cent would change the ironcarbon solidification contraction from approximately 0.90 per cent to 0 per cent. If only one-fourth of the iron in a 4-in. cube were involved in void formation, this would produce a net change in void volume of over 2.2 cc, or approximately the magnitude of the largest voids observed. Similarly, a decrease in porosity with lower carbon contents would necessitate a very drastic change in some other shrinkage factor to more than compensate for the greater solidification contraction.
- (5) Exuded beads are frequently found in the top pipes of castings exhibiting internal shrinkage porosity. These beads are believed to be formed from liquid metal forced out of the casting by high internal pressures which would tend to produce maximum soundness. Since this phenomenon occurs quite late in the solidification process, it is extremely difficult to conceive of a sufficiently rapid reversal in pressure conditions to form internal voids before the completion of iron-carbon eutectic solidification.
- (6) Another investigation, involving tensile tests on direct cooling from the liquid state, showed quite conclusively that phosphorus sharply decreases the strength of gray iron in the range 2000 to 1700 F and thus would tend to improve soundness at the ironcarbon eutectic temperature by piping more readily under atmospheric pressure.

It is apparent from this analysis that a mechanism based on occurrences at or above the iron-carbon eutectic temperature cannot possibly explain the marked differences between normal shrinkage and

that associated with phosphorus and certain alloys. All of the observations have pointed toward an internal shrinkage phenomenon which occurs after the iron-carbon eutectic has completed its solidification and the casting has assumed its final physical form. The formation of internal voids under such conditions would be quite independent of the various factors affecting iron-carbon solidification and would produce porosity in a hot spot regardless of the cooling conditions.

## II. Phosphide Eutectic Properties

Accordingly, an investigation was undertaken to reexamine the properties of the phosphide eutectic and, more specifically, to determine: (1) the influence of alloying elements on its composition and mode of solidification, (2) the liquid and solidification volume changes of these compositions, and (3) if the volume changes could produce void volumes of the magnitude measured in the 4-in, cubes.

History. The constitution of the various systems of phosphorus with the elements in gray iron have been studied primarily by German investigators^{13,14,15,16,17}, ^{18,19,20}. In the binary iron-phosphorus system, they established the existence of the compounds Fe₃P and Fe₂P at 15.5 and 21.7 per cent respectively, with an iron-Fe₃P eutectic appearing at 10.2 per cent P and 1922 F. The solubility of phosphorus in alpha iron at room temperature was set at about 1.2 per cent and approximately 0.6 per cent P was found to close the gamma loop. The diagram as drawn by Vogel¹⁷ is shown in Fig. 10.

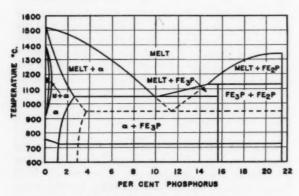


Fig. 10-Iron-Phosphorus System (Vogel).

In the ternary iron-carbon-phosphorus system, an iron-Fe₃C-Fe₃P eutectic was found at approximately 2.0 C, 6.9 P, and 91.1 Fe, with a solidification temperature of 1747 F. The addition of up to 2.8 per cent silicon to this system did not change the eutectic composition or solidification temperature appreciably. The ternary diagram of Goerens and Dobblestein¹⁹ as modified by Von Keil and Mitsche²⁰ is illustrated in Fig. 11.

One important factor which has been frequently overlooked by previous reviewers is the metastability of the phosphide eutectic in the presence of silicon. In other words, it may solidify in one of two forms, as an iron-Fe₃P-Fe₃C eutectic or as an iron-Fe₃P-graphite

eutectic, in a manner similar to the double iron-carbon system. This has been quite conclusively demonstrated by previous investigators 19,20,21,22 by means of chemical analyses and photomicrographs. The iron-Fe₃P-Fe₃C eutectic produces a coarse structure consisting of long Fe₃C and Fe₃P needles in a two-phase background of fine "shot-like" austenite transformation product in a Fe₃P matrix. The iron-Fe₃P-graphite eutectic produces a much finer, more homogenous structure with no Fe₃C needles and a pseudo-binary appearance. The latter structure is most frequently shown as representative of the phosphide eutectic or steadite in gray iron. It is believed that more of the graphite in this eutectic separates out onto existing graphite flakes as the eutectic temperature is approached.

In a study of the phosphide eutectic in gray iron, Bardenheuer and Kunkele^{21,22} found that the metastable iron-Fe₃P-Fe₃C modification solidifies at 1744 F, while the stable iron-Fe₃P-graphite form completes solidification at 1760 F. Their results indicate that the graphitization of this eutectic is much more sensitive to carbide stabilizing factors, such as cooling rate or silicon content, than the corresponding iron-carbon eutectic. Either or both forms of phosphide solidification were frequently observed in normal gray irons. A large increase in the base phosphorus content was found to promote graphitization, thereby explaining the predominance of the graphite form in published photomicrographs which usually show the relatively large fields of steadite in high phosphorus irons.

Incidentally, this metastability may explain the inconsistencies in reports of the carbide stabilizing influence of phosphorus on gray iron. If the phosphide eutectic were solidifying in the carbide form, higher phosphorus would reduce graphitic carbon by simply tying up more carbon in the steadite phase. However, when the phosphorus was increased sufficiently to promote graphitization of the metastable eutectic, an increase in graphitic carbon would be noted. On the

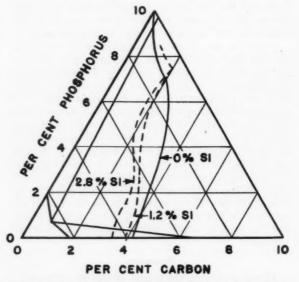


Fig. 11—Iron-Carbon-Phosphorus-Silicon System (Von Keil & Mitsche).

other hand, if the eutectic were originally solidifying in the stable graphite form, no variation in graphitic carbon would be observed with increasing phosphorus.

The solubility of phosphorus in gray irons and pig irons has been variously reported, with values ranging from 0.04 to 0.12 per cent. Stead²³ noticed an apparent increase in the solubility with slower cooling rates which was later explained by Vogel¹⁷ from a phase rule study of the iron-carbon-phosphorus system confirmed by quenching experiments. He demonstrated that phosphorus has a greater solubility in ferrite than in austenite, a peritectoid reaction causing part of the Fe₃P to redissolve on cooling through the gamma-alpha transformation range. Thus, the closer approach to equilibrium produced by slow cooling would tend to make this re-solution more complete, and smaller quantities of steadite would appear on metallographic examination. The function of cooling rate undoubtedly explains the wide discrepancies in reported solubilities, since they were nearly always obtained from metallographic studies.

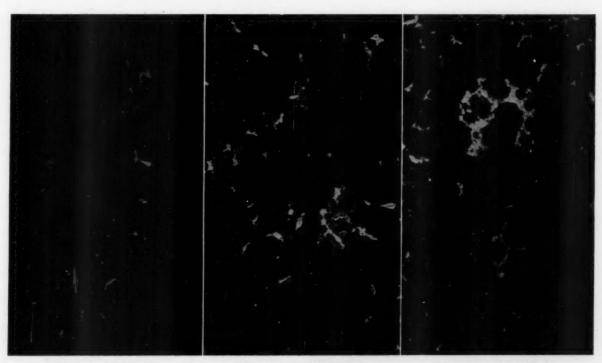
Very few investigators have studied the properties of the phosphide eutectic in the presence of alloying elements. Rote and Wood¹² obtained the outstanding findings in this field from a study of exuded beads from nickel-molybdenum gray irons. Chemical analyses of the beads revealed a strong tendency for molybdenum to segregate to the phosphide eutectic. High phosphorus 2.65 C, 2.25 Si, 2.50 Mo, 1.75 Ni irons consistently yielded phosphide beads containing 2.10 C, 1.10 Si, 4.40 P, 5.85 Mo, and 1.65 Ni, leading to the conclusion that phosphorus in the eutectic re-

moves 1.3 times as much molybdenum from the matrix

In similar base irons containing 0.95 to 1.60 P and only 0.75 Mo, nearly all of the molybdenum in the matrix segregated to the phosphide eutectic because there was not enough present to satisfy this ratio. Metallographic and thermal analysis of reheated specimens showed quite conclusively that this molybdenum segregation occurs only in the cooling range between the iron-carbon and the phosphide eutectic, apparently with a high diffusion rate. Nickel exhibited no tendency toward a segregation of this type, distributing about equally between the phosphide and the matrix.

## Phosphide Eutectic in Cubes

Metallographic studies of the phosphide eutectic in 4-in. cubes produced several interesting findings. All cubes containing molybdenum or chromium consistently showed larger areas of steadite at a given phosphorus level, indicating that both alloys must segregate to the eutectic and thereby increase its quantity. A comparison may be seen in Fig. 12, which shows representative areas in the centers of cubes alloyed from the same base heat. The specimens were etched with an arsenic solution developed by Schneidewind and Harmon²⁴ which darkens all microconstituents except the phosphide eutectic. It will be noted that the molybdenum iron shows a marked tendency to form continuous phosphide networks, while the chromium iron produces more widely scattered phosphides similar to its scattered micro-



Cube No. P7 (Plain)

Cube No. P2 (0.75 Cr)

Cube No. P4 (1.00 Mo)

Fig. 12—Increase in quantity of phosphide eutectic in center of 4-in. cubes produced by chromium and molybdenum additions to a 0.13 P base iron. Arsenic etch. Mag. 250x.

TABLE 4-CHEMICAL ANALYSES OF EXUDED BEADS FROM 4-IN. CUBES

	Composition of Cube*		1		Con	position of	Beads			
No.	. P	Alloy	P	Ni	Cr	Мо	T.C.	Si	Mn	S
T-1	0.13	(Plain)	0.91				2.83	1.97	0.55	0.04
T-2	0.13	2.00 Ni	0.48	1.62			2.75	1.65	0.57	0.04
T-7	0.13	0.75 Mo	0.55			2.21	3.06	1.41	0.30	0.04
T-4	0.13	0.97 Cr	0.65		2.24		3.49	1.70		
C-11	0.04	Ni-Cr-Mo**	0.10							
C-12	0.06	Ni-Cr-Mo	0.25							
C-13	0.08	Ni-Cr-Mo	0.41		1.58	2.52		1.87		
C-15	0.15	Ni-Cr-Mo	0.46	0.54	1.58	2.40	3.08			
C-17	0.28	Ni-Cr-Mo	0.70							
		180000000000000000000000000000000000000								

Base Composition: 3.24-3.32 T.C., 2.23-2.29 Si, 0.77-0.81 Mn, 0.09-0.11 S.

** 0.66 Ni, 0.84 Cr, 0.83 Mo.

porosity. It should be remembered that the quantity of phosphide eutectic visible in the photomicrographs is undoubtedly considerably less than that present at its solidification temperature, due to the peritectoid gamma-alpha reaction.

 $\gamma + Fe_3P \rightarrow a + Fe_3C$ .

A comparison between surface and center specimens from each cube revealed no marked segregation of phosphorus toward the center in any of the irons. This was confirmed by chemical analyses taken from several points. In some cases, the surface phosphorus would be reported 0.01 per cent higher than the center; in others it would be reversed. An absence of segregation from surface to center would be expected in view of the nearly equilibrium solidification conditions observed in the thermal studies.



Fig. 13—Photomicrograph of phosphide eutectic in center of Ni-Cr-Mo cube, showing typical iron-Fe₃P-Fe₃C eutectic structure. Etch—2% Nital. Mag.—500x.

A study of the steadite structures at high magnifications revealed that the presence of molybdenum or chromium produces a drastic change in the solidification of the phosphide eutectic. All irons containing over 0.50 per cent of either or both of these elements showed the coarse, needle-like phosphide structures characteristic of the iron-Fe₃P-Fe₃C eutectic, even in the presence of sufficient nickel to produce balanced alloy additions. A representative structure of this type is illustrated in Fig. 13. The white carbide needles, brought into relief by deep etching, may be seen extending completely across the eutectic area. As the molybdenum or chromium content was reduced below 0.50 per cent, increasing quantities of the fine pseudo-binary iron-Fe₃P-graphite eutectic were evident. The plain and straight nickel cubes exhibited only the pseudo-binary graphite type at all composition levels.

## **Exuded Beads from Cubes**

The exuded beads which frequently appeared in

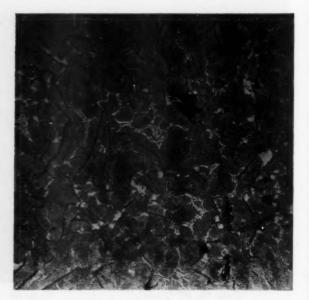
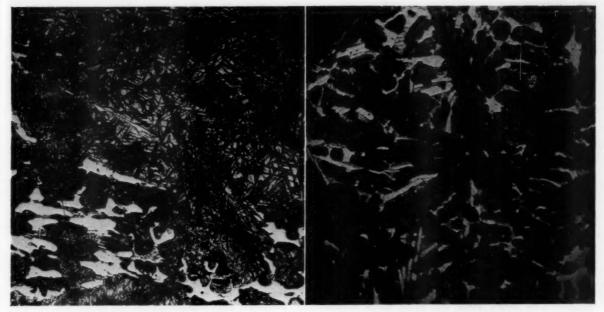


Fig. 14—Photomicrograph of bead from Cube No. T2 (2.00 Ni), showing typical structure of plain and nickel phosphides. Etch—2% Nital. Mag.—100x.



Bead from cube No. 11 (0.04 P)
Bead analysis—0.10 P
Etch—2% Nital Mag.—200x

Bead from cube No. 17 (0.28 P)
Bead analysis—0.70 P
Etch—2% Nital Mag.—100x.

Fig. 15—Photomicrographs of beads from Ni-Cr-Mo cubes, showing typical structure of chromium and molybdenum phosphides.

the top sink of the cubes, as was shown in Fig. 1, yielded additional information on the properties of the phosphide eutectic. Chemical analyses of typical beads are presented in Table 4, together with the base cube compositions. Since all of the cubes were poured from relatively low phosphorus irons, the beads contained considerable quantities of iron-carbon eutectic in addition to phosphide eutectic.

The first series of beads was obtained from cubes poured from the same 0.13 per cent P base iron with various alloy additions. Both molybdenum and chromium indicated strong tendencies for segregation to the phosphide, the alloy contents of the "impure" beads being over twice that of the base irons. Nickel showed no tendency for segregation, distributing slightly less than its base content to the phosphide. Sulphur and manganese were both strongly rejected from the high phosphorus liquid, due to the high melting point of manganese sulphide. The tendency for molybdenum and chromium to carry higher carbon contents into the phosphide eutectic is noteworthy and will be discussed further in the next section.

The second series of beads was obtained from the Ni-Cr-Mo cubes having various phosphorus contents. Since the liquid metal was exuded from the cubes near the end of iron-carbon eutectic solidification, these results provide a very accurate measure of the solubility of phosphorus in austenite in the range of the phosphide eutectic temperature. Any enrichment of the remaining liquid would indicate that the solubility is lower than the phosphorus content of the base iron, since austenite would be expected to separate out with its full phosphorus content. The slight

enrichment of the liquid in the 0.04 per cent P iron shows quite conclusively that the austenite solubility is slightly below this level in the presence of the alloys. The higher phosphorus irons exhibited increasing percentages of phosphorus in the beads, as would be expected.

Photomicrographs of representative beads from these series are shown in Figs. 14 and 15. Even at the fairly rapid solidification rates experienced by the small drops of liquid, both the plain and straight nickel beads exhibited the well developed gray iron structures and typical pseudo-binary phosphide eutectic areas shown in Fig. 14. The molybdenum and chromium beads, on the other hand, all revealed the typical iron-Fe₃P-Fe₃C type of structure illustrated in Fig. 15. It will be noted that the phosphide areas in these beads are all associated with large amounts of massive carbide. The small amount of phosphide eutectic in the bead from the 0.04 per cent P cube may be seen in areas joining the massive carbides.

#### **Exuded Beads from High Phosphorus Irons**

In order to estimate the compositions of the complex alloy-phosphide eutectics, exuded beads of greater "purity" or higher phosphorus content were required. For this purpose, a series of high phosphorus heats of base composition 2.60 C, 2.20 Si, 0.95 Mn, 0.90 S, and 0.92-1.50 P were heavily alloyed and cast into cylinders 3½ in. in diameter and 5 in. high, molded in baked core sand. The cylinders were found to produce a maximum yield of beads in preliminary trials, and previous investigations have indicated that low carbon compositions are conducive to bead formation, probably due to their wider freez-

TABLE 5—CHEMICAL ANALYSES OF EXUDE D BEADS FROM HIGH PHOSPHORUS IRONS

	Composition of Casting®		Casting* Composition of Beads								
No.	P	Ni	Cr	Mo	P	Ni	Cr	Mo	T.C.	Si	Mn
R16-2	1.45		(Plain)		5.57					0.80	0.66
R27-47	1.30		(Plain)		4.84				2.10	1.12	0.72
R17-6**	1.50		,	2.50	5.26						
R27-44	1.30			3.03	4.75			8.32			
R22-13	1.50	3.00		1.50	4.83	2.69					
R27-45	1.30	1.00		2.50	3.31	0.75		4.00			
R28-52	0.92	0.75		1.50	5.02			7.24			
R24-20	1.30		1.50		4.69		2.07				
R24-24*	1.30		1.50		4.82		2.14				
R28-50	0.92	0.66	1.46		3.83	0.69	1.95				
R23-16	1.30		2.00	2.00	4.25		1.90	3.69			
R23-17	1.30		1.00	2.00	2.82						

^{*} Base Composition: 2.55-2.59 T.C., 2.16-2.21 Si, 0.98 Mn, 0.09 S.

ing range. All heats were inoculated with 0.50 per cent of the total silicon to promote graphitization.

The chemical analyses of the beads are summarized in Table 5, together with the base phosphorus and alloy contents of the irons from which they were obtained. The beads from the plain irons revealed phosphorus contents considerably below the equilibrium value of 6.9-7.0 per cent for the unalloyed phosphide eutectic, indicating that they still contained some iron-carbon eutectic "impurity." This may be seen in Fig. 16, which clearly illustrates the steps in the solidification of compositions below the eutectic phosphorus content.

The separation of the dark primary austenite dendrites was followed by the solidification of the iron-carbon eutectic, which graphitized completely to form the homogeneous gray regions of ferrite. The remaining liquid then cooled to the phosphide eutectic temperature, separating its carbon on existing graph-

ite flakes, and solidified as the pseudo-binary iron-Fe₃P-graphite eutectic which appears as a mottled background. Close inspection will reveal that some of the graphite flakes have grown into and through the phosphide eutectic liquid during the carbon separation. The white needles visible in the eutectic are Fe₃P and not Fe₃C, since they could not be darkened by an acid FeCl₃ etch which readily attacks the latter phase.

The beads from the alloy irons, exemplified by Figs. 17, 18, and 19, revealed no phases which could arise from iron-carbon eutectic solidification, indicating that they are very close to the pure phosphide eutectic compositions in equilibrium with alloy gray irons. This conclusion is indirectly confirmed by the findings of Rote and Wood, 12 who demonstrated that molybdenum does not begin to diffuse into the phosphide until the completion of iron-carbon eutectic solidification. The high molybdenum content of the

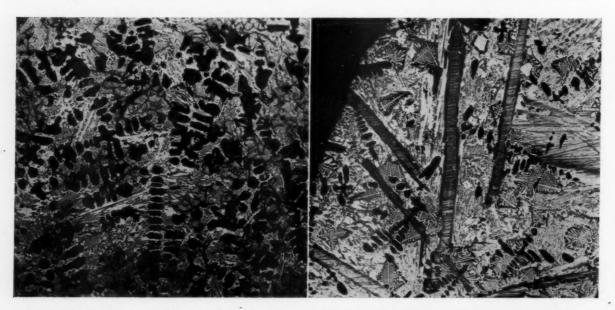


Fig. 16—Microstructure of exuded bead from high phosphorus plain iron R27-47. Bead analysis: 2.10 T.C., 1.12 Si, 4.84 P, 0.72 Mn. Etch—Acid FeCl₃. Mag.—100x.

Fig. 17—Microstructure of exuded bead from high phosphorus chromium-nickel iron R27-44. Bead analysis: 4.75 P, 8.32 Mo. Etch—Acid FeCl₃. Mag.—100x.

^{**} Strongly inoculated with both FeSi and graphite before pouring.

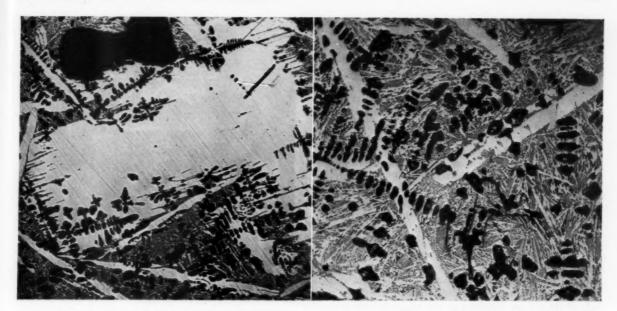


Fig. 18—Microstructure of exuded bead from high phosphorus chromium-nickel iron R28-50. Bead analysis: 3.83 P, 1.95 Cr, 0.69 Ni. Etch—Acid FeCl₃. Mag.—50x.

Fig. 19—Microstructure of exuded bead from high phosphorus chromium iron R24-24, showing effect of strong inoculation treatment. Bead analysis: 4.82 P, 2.14 Cr. Etch—Acid FeCl₃. Mag.—100x.

alloy beads indicates that the remaining liquid must have cooled considerably below the iron-carbon eutectic to provide sufficient time for the diffusion before exuding.

The bead analyses reveal that molybdenum and chromium both sharply reduce the phosphorus content of the eutectic and to about the same extent. In every case, the very low phosphorus contents were found to be associated with the formation of abnormally large massive carbides, similar to that shown in Fig. 18, and the amount of the phosphorus reduction showed a close relationship to the size of the carbide. The analyses of beads containing little or no excess carbide, exemplified in Fig. 19, indicate that the pure eutectics contain approximately 5.0 per cent P over a wide range of alloy compositions containing molybdenum or chromium. Incidentally, this tendency for molybdenum and chromium to carry additional carbon into the eutectic and thereby further reduce its phosphorus content would greatly increase the quantity of phosphide in a gray iron casting. At a given phosphorus level, a phosphide eutectic containing 3.5 per cent P would occupy twice as much volume as the phosphide eutectic in unalloyed iron containing 7.0 per cent P.

The segregation of molybdenum from straight molybdenum irons and those containing small additional percentages of nickel was remarkable. A 3.03 per cent Mo iron segregated 8.32 per cent Mo into the phosphide eutectic, while a 1.50 Mo, 0.75 Ni iron produced beads containing 7.24 per cent Mo. Additions of nickel or chromium appear to reduce the segregation to some extent. This tendency may be semi-quantitatively evaluated by computing the ratio of molybdenum to phosphorus for the four analyses in Table 5 containing molybdenum:

% Ni in Beads % Cr in Beads No. % Mo/% P in Beads R27-44 1.75 0 0 (0.50 - 0.60)R 28-52 1.44 0 R27-45 1.21 0.75 R23-16 1.90

Thus, a steady trend for reduced molybdenum segregation with increasing amounts of other alloys is indicated. Chromium probably exerts the greatest influence because of its own tendency for segregation.

The segregation of chromium to the phosphide eutectic appears to reach a maximum at about 2.15 per cent. Other alloys may reduce this segregation slightly, although the chromium to phosphorus ratios appeared to remain quite constant throughout. Nickel again exhibited no tendency toward segregation, distributing slightly less than its base content to the phosphide, and it might be expected that this alloy would not reduce the eutectic phosphorus content below 7.0 per cent.

The beads from R27-44, containing 4.75 P and 8.32 Mo, revealed some very interesting structures, which may be seen in Fig. 17. The acid ferric chloride etch darkened the carbides to differentiate them from the white phosphides. Due to rapid freezing and concentration gradients, four constituents appear in coexistence. A few austenite dendrites form the black scattered patterns, while white angular crystals of primary  $Fe_3P$  are also visible. The appearance of these two phases simultaneously is quite unusual, since they lie on opposite sides of the ternary eutectic.

As the bead cooled to near the eutectic temperature, the strong crystallization tendency of the carbide apparently depleted the melt in carbon at several points, resulting in herringbone binary iron-Fe₃P structures adjacent to the needles. Finally, the re-

mainder of the melt solidified as the fine ternary eutectic in the background. Although the carbide needles are actually part of the ternary iron-Fe₃P-Fe₃C eutectic, previous investigators have found that their strong crystallization tendency makes them appear to solidify separately from the remainder of the eutectic. It is interesting to note that the needles crystallized completely through austenite dendrites and over corners of primary phosphides.

Another interesting phenomenon was the influence of very strong inoculation on the phosphide structures. A bead obtained from a chromium iron inocu lated with ferro-silicon, illustrated in Fig. 18, shows a typical massive carbide which may be clearly distinguished by the relief effect, although the chromium content prevented darkening by etching. A similar iron, strongly inoculated with both ferro-silicon and graphite, produced beads of the type shown in Fig. 19. It can be seen that the intensive inoculation treatment persisted into the phosphide eutectic, preventing the formation of massive carbides and causing a small amount of graphitization. A similar result could not be obtained in the molybdenum irons, indicating that chromium carbide might be slightly less stable than molybdenum carbide in the presence of high phosphorus.

## Phosphide Volume Changes

Based on these findings, the liquid and solidification volume changes of representative plain and alloy phosphide eutectic compositions could then be determined. A review of the literature revealed that the pyknometer and sand cast bar techniques employed by Ash and Saeger^{10,11} in an investigation of the volume changes of white and gray irons should provide the best method for this study.

The pyknometer technique for the determination of the densities of liquid metals consists essentially of removing samples of known volume from the melt at various temperatures and weighing the resulting

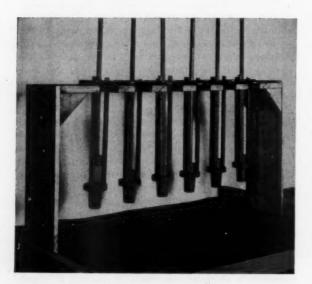


Fig. 20—Graphite immersion crucibles for liquid density determinations.

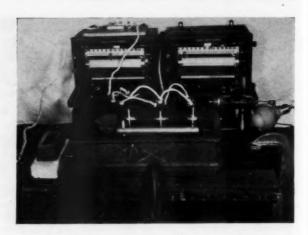


Fig. 21-Solid contraction equipment.

ingots. Since the volume of the metal is determined at the temperature in question and the mass remains constant, the densities are independent of the mode of solidification or the soundness of the ingots.

For this purpose, a set of six immersion crucibles shown in Fig. 20, were constructed entirely of graphite. A small lid, loosely threaded to the center plunger rod and machined to a close fit with the crucible lips, insured the entrapment of a known volume. Before each use, the crucibles were coated with a slurry of fine zirconia and fire clay and fired in a brazing torch. This coating prevented any reaction with the melt for the duration of the immersion. After coating, the crucible volumes at room temperature were determined by filling them with mercury and weighing the quantity required. The volumes at the sampling temperatures could then be determined from the expansion coefficient of graphite between these temperatures.

In sampling, the furnace was set to maintain a constant temperature, the melt carefully skimmed, and the crucible immersed with a quartz-tipped platinum-platinum rhodium thermocouple inserted in the cavity. When the crucible and lid reached the constant temperature of the melt, the thermocouple was removed, the lid tightly pressed against the lip of the crucible, and the entire assembly lifted out and placed on the rack for cooling in air. After cooling to room temperature, the ingots were removed, cleaned of any adhering coating, and carefully weighed.

The sand cast bar technique for measuring solid contractions consists of securely fixing one end of the bar and observing the movements of the free end on cooling after solidification. Thermocouples placed along its length measure the average temperature to establish the linear contraction-temperature relationship, and the volume contraction of the alloy is obtained by multiplying the linear contraction by a factor of three. The density of the bar at room temperature determines the lower end of the solid contraction curve, which may be recomputed in terms of specific volumes and plotted with the liquid specific volumes determined by the pyknometer method.

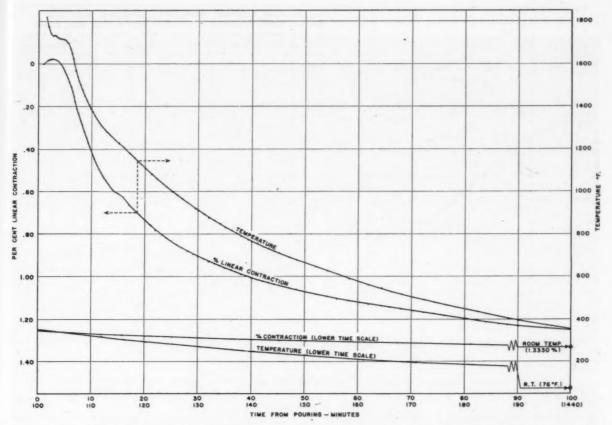


Fig. 22-Experimental linear contraction and temperature curves from nickel phosphide bar H8.

The solidification contraction is then given by the change in specific volume over the freezing range of the alloy.

A cut-away view of the solid contraction equipment used in the present investigation is illustrated in Fig. 21. The bar was cast in a 1.2-in. core sand test bar mold rammed into a flask of green sand. The fixed point was provided by a stainless steel studbolt inserted through the center of the mold cavity at one end of the bar and securely bolted to the bottom board. A core sand plug, placed in the free end of the mold with its face exactly 12 in. from the center of the fixed pin, provided a standard contraction length.

A fused quartz rod was then inserted through holes in the flask and plug and extending ½ in. into the mold cavity to translate the movements of the free end of the bar to the micrometer microscope. Three small alundum-tipped thermocouples inserted into the centerline of the mold cavity along its length were connected in parallel to measure the average bar temperature. The thermocouples were placed in oversize holes and completely freed from the mold as soon as the bar solidified to avoid any hindrance to contraction. Due to the low pouring temperatures of the phosphide eutectics, the bars had a very smooth, mirror finish, indicating that little or no mold resistance could have been encountered.

The freezing range of the alloy was determined from a separate specimen shown at the left side of Fig. 21. A hollow insulation brick provided very slow cooling for a close approach to equilibrium solidification conditions. Two thermocouples inserted into the center of the cavity were alternately read every 0.1 min from pouring.

The compositions chosen for this study were those representing the saturated phosphide eutectics in plain, straight nickel, straight molybdenum, and straight chromium irons, as determined by the exuded bead studies. Following preliminary melts to check alloy recoveries, the heats were melted in a 30-lb induction furnace from charges of wash metal, ferro-24-phosphorus, coke, electrolytic nickel, ferro-60-molybdenum, and ferro-70-chromium.

The melts were held under a coke blanket to obtain maximum solution of carbon and skimmed at 2600 F for the first crucible sampling. Successive samples were then removed on cooling at 125-150 F intervals until the last sample was obtained at approximately 1900 F. The remaining metal was reheated and poured into the solid contraction mold and insulation brick at 2400 F. Successive readings were taken on the micrometer microscope and bar temperature every 0.1 min after pouring until the contraction rates had slowed considerably. The readings were then continued at intervals until the bar reached room temperature. A specimen was then cut from the center of the bar, carefully ground, and weighed in air and water to determine its density.

A typical plot of the linear contraction and bar

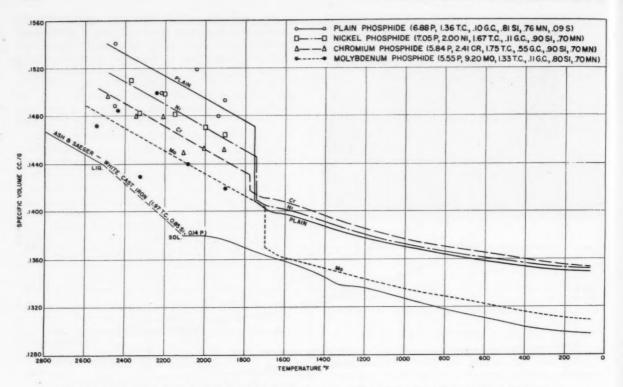


Fig. 23-Volume changes of experimental phosphide com- positions on cooling from the liquid state.

temperature data is shown in Fig. 22. The very slight expansion during solidification was obtained with all of the alloys and is probably due to the heat effect produced by rapid solidification after undercooling. Ash and Saeger¹¹ observed a similar small expansion in both white and gray irons which they attributed to a release of dissolved gasses. A well defined decrease in the rate of contraction may also be seen at approximately 1220 F (15.5 min after pouring), corresponding with the alpha-gamma transformation of the small amount of austenite in the phosphide eutectic.

The cooling curves obtained from the samples in the insulating bricks showed well defined eutectic temperatures with no indication of undercooling. The plain eutectic solidified at 1753 F, corresponding with the results of the German investigators, and the molybdenum-phosphide eutectic solidified at 1700 F, which lies about midway between the values obtained by Wood and Rote¹² on cooling and reheating phosphorus-bearing nickel-molybdenum irons. Chromium raised the eutectic temperature to 1777 F, while nickel reduced it slightly to 1742 F.

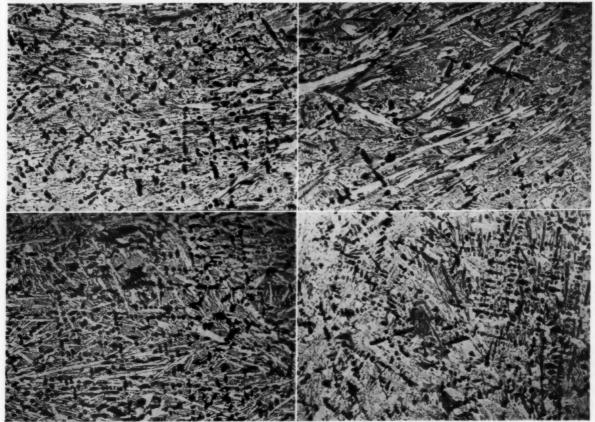
The specific volume-temperature relationships computed from the liquid and solid contraction data for the four phosphide compositions are shown in Fig. 23, together with the curve reported by Ash and Saeger¹¹ for white cast iron of approximately the same carbon and silicon content. The large quantity of low density phosphorus in the eutectics resulted in liquid specific volumes considerably greater than those shown by the white iron at corresponding temperatures. This effect was partially offset by replac-

ing some of the iron with high density molybdenum and nickel in the alloy eutectics. The average lines drawn through the liquid specific volume points for each alloy gave a liquid contraction rate of 1.1 per cent per 100 C, corresponding with the results of previous investigators¹¹ for a wide range of iron-base alloys.

The low graphite carbon contents in the chemical analyses of the solid contraction bars indicated that all four eutectics solidified almost entirely in the metastable iron-Fe₃P-Fe₃C form. This was confirmed by the microstructures of the bars, illustrated in Fig. 24, which show the typical metastable eutectic structures observed in the exuded bead studies and the carbides darkened by the acid ferric chloride etch. The slightly higher graphitic carbon in the chromium-phosphide appeared in concentrated spots similar to temper carbon, and was probably associated with the inoculation of the melt by a graphite crucible, causing the peculiar instability of the chromium carbide noted previously. In the other alloys, the small amount of graphite appeared only in the primary dendrites.

The solidification contractions of the metastable eutectics could be quite precisely determined from these data, since the liquid density points fell verclose to the average lines as the eutectic temperature was approached and the solid contraction curves originated in the eutectic temperature. However, these curves do not present a completely accurate picture of the eutectics in gray iron, where the phosphide phase is surrounded by graphite flakes and the diffusion of various elements is involved.

Heat H5—Plain phosphide (6.88 P, 1.36 T.C., 0.10 G.C., 0.81 Si, 0.76 Mn) Heat H8—Nickel phosphide (7.05 P, 2.00 Ni, 1.67 T.C., 0.11 G.C., 0.90 Si)



Heat H7—Chromium phosphide (5.84 P, 2.41 Cr, 1.75 T.C., 0.55 G.C., 0.90 Si)

Heat H6—Molybdenum phosphide (5.55 P, 9.20 Mo, 1.33 T.C., 0.11 G.C., 0.80 Si)

Fig. 24—Microstructures of phosphide compositions used in volume change determinations. Etch—Acid FeC1₃.

Mag.—100x.

It will be recalled from the metallographic studies of the cubes and exuded beads that the plain and nickel phosphides in gray irons consistently solidified as the graphite eutectic, while the saturated molybdenum and chromium phosphides solidified in the metastable carbide form. In addition, it has been quite conclusively shown that the phosphide in molybdenum irons originally forms as a low allov phosphide eutectic and the molybdenum diffuses in rapidly after the completion of iron-carbon solidification. Due to the similar segregation of chromium to the phosphide, it might be expected that this alloy behaves in a similar manner. Consequently, a few simple transformations of the original data are necessary to obtain a true picture of the phosphide volume changes in gray iron.

## Volume Changes in Gray Iron

First of all, the origins of the liquid specific volume curves must conform with the mechanism for phosphide eutectic formation. During iron-carbon solidification, the remaining liquid surrounding the eutectic cells is being continuously enriched in phosphorus and does not achieve its full phosphorus content until iron-carbon solidification has been completed. Consequently, the liquid specific volume curves for the saturated phosphides shown in Fig. 23 apply only at temperatures below the iron-carbon eutectic (about 2050 F). At higher temperatures, these curves would slope downward to the left of the plot and eventually intersect the curve for gray or white iron of the same carbon and silicon content.

Secondly, the liquid specific volume curves for the molybdenum and chromium phosphides should be adjusted to correspond with their aforementioned low alloy content on formation at 2050 F. Assuming that these eutectics form with only 1.0 per cent alloy content and the same phosphorus content, their specific volumes at 2050 F could be readily computed from the original data on the thesis of the proportionality between liquid specific volumes and percentages of added elements shown by Benedicts, Erickson, and Erickson.²⁵

The resulting values are plotted in Fig. 25. As would be expected, the initial specific volumes of these alloys fall very close to that of the plain eutectic and it is apparent that a slight error in the assumptions would have a minor effect on the overall

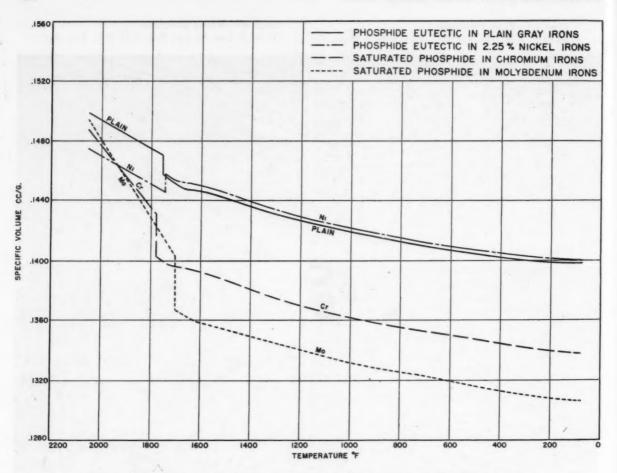


Fig. 25-Volume changes of saturated phosphide eutectics in plain and alloy gray irons.

plot. Since the molybdenum and chromium phosphides are saturated on solidification, the liquid specific volumes at their solidification temperatures could be transferred directly from the original data. The liquid specific volume curves were then approximated as straight lines between 2050 F and solidification.

Since nickel exhibits no tendency for segregation of reduction of the eutectic phosphorus content, the nickel phosphide would be expected to form at 2050 F with full alloy content. Consequently, its liquid specific volume curve was retained in its original form, as was that of the plain eutectic.

The final adjustment to the original data involves the consistent graphitization of the plain and nickel phosphide eutectics in gray iron. Since the experimental phosphide contraction bars contained only 0.11 per cent graphitic carbon, the room temperature densities which determined the lower end of the solid contraction curves were much greater than would be obtained with the fully graphitized eutectics. Thus, the entire solid specific volume curves fell at abnormally low values and produced solidification contractions for these alloys which do not represent their true behavior in gray iron. The only correction required in this case is simply the change in solidification contraction accompanying a given increase in graphitic carbon at the same total carbon content.

Ash and Saeger¹¹, Honda and Endo,²⁶ and Honda, Kase, and Matsuyama²⁷ have measured the relationship between solidification volume change and graphitic carbon for a wide range of iron alloys containing carbon, silicon, and phosphorus. Their results are all in very close agreement on the rate of change in solidification contraction with change in graphitic carbon, which may be expressed by the following average formula:

Increase in % Solidification Contraction = -2.0 x (Increase in % G. C.)

From measurements on the ternary Fe-C-P diagram. the iron-Fe₃P-graphite eutectic consists of approximately 49 per cent austenite, 47 per cent Fe₃P, and 4 per cent graphite by weight. Assuming that the austenite in the eutectic transforms to pearlite with 0.65 per cent combined carbon, the entire eutectic would then contain  $49\% \times 0.65 = 0.32\%$  combined carbon. With 2.10 per cent total carbon, as indicated by the exuded beads, the phosphide-graphite eutectics would then contain 1.78 per cent graphitic carbon, an increase of 1.67 per cent over that in the experimental phosphide bars. Thus, applying the above formula, -3.34 per cent should be added to the measured solidification contractions for both alloys to obtain the correct values for their contractions in gray irons. Computing from the specific volumes in Fig. 23, the

measured contractions are 4.21 per cent and 2.52 per cent, resulting in values of 0.87 per cent and —0.82 per cent for the solidification contractions of the plain and nickel phosphides in gray iron, respectively.

Since the liquid specific volumes are not affected by a graphitization correction, the solid specific volumes at solidification could be computed back from these contractions and plotted as origins for the new plain and nickel phosphide solid contraction curves shown in Fig. 25. In view of the high alloy content of the saturated molybdenum and chromium phosphides and their strong tendency for massive carbide for ation, it is extremely unlikely that they would produce any graphite on normal solidification in gray iron. Consequently, the molybdenum and chromium phosphide solid contraction curves were similarly translated downward slightly to correct the small amount of graphitization obtained in the experimental bars.

The resulting plot, Fig. 25, represents the volume changes of the saturated phosphide eutectics in plain, nickel, molybdenum, and chromium gray irons. In considering the molybdenum and chromium phosphides, it must be remembered that these volume changes apply only when the alloys have diffused into the phosphide eutectic in sufficient quantities to approach saturation contents. This diffusion depends, of course, on the cooling rate and the alloy and phosphorus contents of the base iron. In the 4-in. cubes, saturation contents were probably closely approached at 0.13 per cent P when the molybdenum or chromium contents reached 1.0 per cent. At lower alloy levels, the volume changes of the phosphides would be intermediate between those of the plain and the saturated alloy eutectics. Similarly, the volume changes shown for the nickel phosphide apply to a 2.25 per cent Ni iron, and lower percentages of this alloy would produce curves intermediate between that shown and the plain phosphide eutectic.

#### Mechanism for Void Formation

On the basis of the liquid and solidification contractions exhibited by the plain, molybdenum and chromium phosphide eutectic compositions, a satisfactory mechanism for internal shrinkage porosity formation could now be developed and quantitatively checked. The 4-in. cube will be chosen as a model for this discussion, since its solidification has been quite thoroughly studied in the present investigation.

Immediately after pouring, a rapidly solidified shell approximately 1/4 in. thick forms on the side and bottom surfaces. On further cooling, the surface of the shell is heated by the interior liquid to approximately 1850 F and the inner layers are built up somewhat by primary solidification. A major portion of the iron in the cube remains liquid for about 8 min, during which time it undergoes liquid contraction and recedes from the top surface of the mold. At the end of this period, iron-carbon solidification proceeds almost simultaneously from nuclei throughout the melt at various degrees of undercooling. The top surface completes solidification during this interval but does not acquire sufficient strength to resist the pressure of the atmosphere, and pipes down to relieve any void formation by iron-carbon eutectic contraction. Thus,

the cube is completely sound at the completion of iron-carbon solidification, regardless of the composition of the iron.

During or shortly after iron-carbon eutectic solidification, the rapid solid contraction of the cooler outer shell begins to subject the interior to high pressure. This pressure is augmented by the expansion of the center due to precipitation of graphite from saturated austenite after iron-carbon solidification, and both factors would tend to insure maximum internal soundness.

The data of Ash and Saeger¹¹ reveal that the graphite expansion reaches a maximum at 1935 F, and then drops off rapidly to a normal solid contraction rate. Although the pressure exerted by the outer shell undoubtedly persists to low temperatures, the inner layers at some point between the surface and center eventually gain sufficient strength to resist further deformation of the center. In another investigation, tensile tests on direct cooling from the liquid state revealed that gray iron exhibits a marked increase in strength at 1950 F, indicating that this phenomenon occurs at a center temperature somewhat above this point. Thus, internal voids cannot begin to form until the center has cooled below 1935 F, at which temperature the center has ceased expanding and the inner layers have gained sufficient strength to resist further deformation by residual pressure gradients.

During iron-carbon eutectic solidification, austenite separates out with the phosphorus content indicated by its solubility at these temperatures, and the liquid surrounding the eutectic cells, which are growing in a radial direction from the original nuclei, is being continually enriched by any phosphorus in excess of this solubility. Due to the slowness of the eutectic solidification and the fluidity of the high phosphorus liquid, which is 300-350 F above its freezing point, the concentration of the liquid would be expected to maintain an excellent equilibrium and form a continuous network around the cells as they grow out to meet each other.

The pressure conditions in the interior of the casting would further improve the continuity, since liquid transmits pressure in all directions and would tend to force its way between any cells which did not have a liquid layer between them. This interlocking network has been frequently illustrated by metallographic specimens etched with Stead's reagent, in which the high phosphorus iron produced by re-solution of Fe₃P during the peritectoid gamma-alpha transformation may be seen completely outlining the eutectic cells.

In the cooling range between the iron-carbon eutectic temperature at about 2050 F and the completion of phosphide eutectic solidification at 1700-1750 F, the liquid eutectic in this network experiences the liquid and solidification contractions shown in Fig. 25. Above a center temperature of 1935 F, these contractions contribute to the relief of internal pressures, but below this temperature, they can result in internal void formation.

Due to the thermal gradients existing in the cube, solidification of the phosphide eutectic proceeds progressively from the cooler outer layers to the interior.

In plain irons and those containing molybdenum and/or chromium, the solidification is accompanied by a marked contraction which would tend to produce a local vacuum at that point. As solidification proceeds, the vacuum increases in the outer layers until capillary action is set up in the network and the temporary voids are filled by liquid phosphide from the interior. This movement continues until sufficient "back vacuum" is built up in the center to resist further capillary action. At this point, further solidification results in voids in the freezing layer and the outer limits of the spongy region have been determined. Thus, the contraction of the phosphide eutectic in a major part of a heavy section can contribute to the formation of voids in the central regions without a detectable segregation of phosphorus to the

A similar analysis may be applied to void formation in lighter section commercial castings. In this case, large temperature gradients and non-equilibrium solidification conditions are produced by the rapid cooling of a smaller mass of iron in a green sand mold possessing high heat absorption properties. Under these conditions, iron-carbon eutectic solidification proceeds directionally from the mold walls, and the remaining liquid, which is being continually enriched in phosphorus by the solidification of low phosphorus austenite, segregates strongly to a point midway between the cooling surfaces. The hot spot of the casting then contains most of the phosphide eutectic produced by the given mass of iron, and its subsequent liquid and solidification contractions produce concentrated voids.

## Correlation Between Computed and Measured Void Volumes

In order to check the proposed shrinkage mechanism, theoretical 4-in. cube void volumes were calculated from the phosphide contractions shown in Fig. 25 and compared with the actual values obtained in the precision density measurements.

A sample calculation of the theoretical void volume in a 1.00 Mo, 0.13 P cube is presented in Table 6, together with a tabulation of data used in deriving the void volume formulas for the four grades of iron. In these computations, it was assumed that the phosphide network is interlocking and continuous throughout 6000 grams of the 7200-gram cube, extending to the rapidly solidified ½-in. layer which formed on the side and bottom surfaces. A network of this extent is indicated by the slow and nearly simultaneous solidification of the large central mass and also by the fact that a drastic reduction in thermal gradients, which would tend to extend the original network, did not produce a detectable change in porosity.

Due to the thermal gradients existing in the cube, the various layers extending out from the center experience varying amounts of liquid contraction after the center reaches 1935 F and void formation may begin. In the 4-in, cube cooling curves shown in Fig. 5, a layer approximately 11/4 in, from the center represents the average cooling conditions of the entire 6000-gram mass. Therefore, its average temperature of 1900

F when the center reaches 1935 F was chosen as the starting temperature for the liquid phosphide contractions.

The percentage of phosphorus in the various eutectics and the solubility of phosphorus in chromium and molybdenum irons were obtained from the exuded bead studies. The solubility of phosphorus in plain and nickel irons was roughly estimated from the relative amounts of eutectic visible at various phosphorus levels. It may be seen from the data to follow that, except for determining the level at which microscopic voids begin to form, slight varations in phosphorus solubility have relatively little influence on the larger void volumes associated with visible porosity.

Based on these data and assumptions, the void volume formulas shown at the bottom of Table 6 could be computed directly from the specific volumes in Fig. 25 and simple material balances. For comparison with the observed void volumes, it will be assumed that the saturated molybdenum and chromium phosphides apply to the eutectics in 1.0 per cent molybdenum and 1.0 per cent chromium irons, while the nickel phosphide applies to a 2.25 per cent nickel iron.

Theoretical void volumes computed from these formulas at representative phosphorus levels are assembled in Table 7, together with the measured values obtained in the present investigation and those reported by Timmons.⁵ In considering the measured void volumes, it will be recalled that these values are

Table 6—Derivation of Formulas for Computing 4-in. Cube Void Volumes Produced by Phosphide Eutectic Contraction

Sample Calculation of Void Volume From Phosphide Eutectic Contraction in a 1.00 Mo, 0.13 P Gray Iron

From Figure 25:	a 1.00 M	o, 0.13 P	Gray Iron	
Spec. Vol. of Liquid Mo				/gram
Spec. Vol. of Solid Phos.	Eut. on So	lidification	0.1367	
Liquid and Solidification 7% P in Molybdenum-Ph				gram Eut.
grams Phosphide Eutect	ic/grams l	Free Phosp	horus=1= .05	=20 gram
grams Free Phosphorus	grams Iro	n = (%P)	in Iron -	_ 0.03)
		-	100	
grams Phos. Eut./6000	grams Iron	$= 20 \times 60$	000× (%P	0.03)
			100	
		= 1200 ×	(% P -	0.03)
Contraction/6000 grams	Iron = 0		0× (% P	-0.03)
Therefore, for a 1.00 M Void Volume =	fo, 0.13P	gray iron:		
Tabulation of D	ata for Ve	oid Volume	Formula	S
	Plain	2.25 Ni	1.00 Cr	1.00 Mo
Sp. Vol. @ 1900 F	0.1485	0.1461	0.1457	0.1455
Sp. Vol. @ Solidif.	0.1458	0.1458	0.1403	0.1367
Contraction/gram Eut.	0.0027	0.0003	0.0054	0.0088
% P in Eutectic	7.0	7.0	5.0	5.0
grams Eut./% Free P				
in 6000 grams Iron	857	857	1200	1200
% P Soluble in Iron	(0.08)	(0.08)	0.03	0.03

2.25 Ni: Void Volume = 0.26 × (% P - 0.08)

1.00 Cr: Void Volume = 6.48× (% P - 0.03)

1.00 Mo: Void Volume =10.56× (% P — 0.03)

all slightly high at an additive factor due to the variation in density between the corner and center of the cube. The 0.04 per cent P cube in the Ni-Cr-Mo series showed no microporosity, indicating a correction factor of —0.35 cc which has been used to obtain the "corrected void volumes" for the remaining cubes poured from this base composition. The void volumes of the plain and straight molybdenum irons would require a correction of similar magnitude. A much larger correction is indicated for the straight chromium irons due to their section sensitivity, while little if any correction would be required for the straight nickel irons.

e

Taking these factors into consideration, it may be seen that an excellent correlation between the computed and measured void volumes is indicated. Since combined chromium-molybdenum additions tend to produce additive shrinkage tendencies, the corrected values from the Ni-Cr-Mo series may be approximately compared with the sum of the computed void volumes for the 1.0 per cent Cr and 1.0 per cent Mo irons at

1 ABLE 7—COMPARISON OF THEORETICAL VOID VOL-UMES PRODUCED BY PHOSPHIDE EUTECTIC CONTRACTION WITH AVAILABLE INTERNAL SHRINKAGE DATA

Theor		ube Void Volumention Formulas		
% P	Plain Iron			1.00 Mo Iron
0.04	0.00 cc	0.00 cc	0.06 cc	0.11 cc
0.06	0.00	0.00	0.19	0.32
0.08	0.00	0.00	0.32	0.53
0.11	0.07	0.01	0.52	0.84
0.13	0.12	0.01	0.65	1.06
0.14	0.14	0.02	0.71	1.16
0.18	0.23	0.03	0.97	1.58
0.25	0.39	0.04	1.42	2.32
0.28	0.46	0.05	1.62	2.64
0.50	0.97	0.11	3.04	4.96
0.75	1.55	0.17	4.66	7.60
1.00	2.12	0.24	5.96	9.71

4-In. Cube Void Volumes Obtained by Density Measurements in Present Investigation:

%P	%Ni	%Cr	%Mo	Measured Void Volume	Corrected Void Volume
0.04	0.66	0.84	0.83	0.35 cc	0.00 cc
0.06	9.9	2.0	9.0	1.15	0.80
0.08	99	99	9.0	1.31	0.96
0.11	22	**	20	1.53	1.18
0.14	**	20	19	1.72	1.37
0.18	20		99	2.29	1.84
0.28	22	81	2.5	2.13	1.78
0.13		(plain)		0.30	
0.13	1.00	,	/	0.32	
0.13	2.15			0.13	
0.13		0.50		0.65	
0.13		1.00		1.57	
0.13			0.46	0.49	
0.13			1.00	1.51	

Void Volumes Reported by Timmons From Density Measurements of 4-In. Cubes Containing 0.70 Mo, 0.30 Cr, 0.16 Ni, 0.12 Cu:

% P	Void Volume
0.015	0.35 сс
0.036	0.43
0.073	2.12
0.104	2.62
0.123	2.96
0.74	12.75

the same phosphorus level. It is interesting to note that the rate of increase in void volume with increasing phosphorus between 0.06 per cent to 0.18 per cent in this series is about 9.5 cc/% P, while the values given by the formulas are 10.56 cc/% P and 6.48 cc/% P for the 1.0 per cent Mo and 1.0 per cent Cr irons, respectively.

The void volume reported by Timmons at the 0.74 per cent P level also corresponds closely with the theoretical data, while the more balanced irons studied by Rote9 and in the present investigation showed a slightly reduced tendency for void formation as the phosphorus was increased above 0.20 per cent. It is highly probable that this varation is related to the decreased carbide stability of the phosphide eutectic observed by Bardenheuer and Kunkele21,22 at higher base phosphorus contents. The phosphide eutectics in more balanced irons containing higher percentages of graphitizing nickel would be expected to exhibit this change from metastable carbide to stable graphite solidification at much lower phosphorus levels than the strongly carbide-stabilized eutectics in unbalanced irons. This graphitization of the phosphide eutectic would, of course, result in much smaller phosphide solidification contractions than those obtained in the contraction studies and included in the theoretical formulas.

Conversely, it might be expected that molybdenum and chromium would have a greater tendency to carry excess carbide into the phosphide eutectic at very low phosphorus levels, thereby reducing its phosphorus content and increasing the quantity of eutectic considerably. This would account for the excessive amount of steadite observed in the low phosphorus cubes and the relatively large void volumes at phosphorus levels as low as 0.06 per cent in the Ni-Cr-Mo series.

The theoretical void volumes are also quite consistent when computed for the phosphorus levels at which visible voids first appear in plain and alloy gray irons. Visible porosity appears in plain gray irons at 0.25 per cent P, corresponding to a computed void volume of 0.39 cc, while larger voids were obtained in the Ni-Cr-Mo iron at 0.06 per cent P, corresponding to computed void volumes of 0.32 cc plus 0.19 cc for the 1.0 per cent Mo and 1.0 per cent Cr irons.

The indicated reduction in the shrinkage tendencies of plain gray irons at higher phosphorus levels by straight nickel additions is noteworthy. This phenomenon is produced by an increase in the liquid density of the phosphide, causing the normal solidification contraction of the plain eutectic to change to an expansion, as shown in Fig. 25. Incidentally, this expansion produced by nickel may explain the expansions observed by Turner⁶ in his very high silicon pig irons, since silicon and nickel frequently behave quite similarly in gray irons.

It is apparent from the preceding discussion that internal porosity formation can be related directly to the liquid and solidification contraction of the phosphide eutectics in gray iron. In addition, many of the anomalies observed in internal shrinkage studies can be logically explained by this mechanism. Higher

pouring temperatures, which reduce thermal gradients in the cube, would be excepted to increase porosity by providing more time for the diffusion of alloys into the phosphide and by extending the phosphide network. Risers would increase the total mass of metal for a similar effect.

Since the phosphide contractions occur after the completion of iron-carbon solidification and after the casting has assumed its final physical form, voids arising from this source would not be closely related to external shrinkage, carbon and silicon content, inoculation, gating and risering, or any of the other variables associated with iron-carbon eutectic solidification. The tendency for molybdenum to produce concentrated voids may be explained by the very low freezing point of the molybdenum-phosphide eutectic, causing it to maintain its fluidity longer for more complete capillary action in the network. Conversely, the reduced tendencies of very low carbon irons for visible porosity formation may be related to the large quantities of interlocking austenite dendrites breaking up the phosphide network and hindering segregation in light sections.

In general, many of the inconsistencies exhibited by this type of shrinkage may be explained by the sensitivity of the phosphide eutectic solidification to carbide stabilizing influences. Thus, a slight increase in residual alloy content, which may be balanced from the standpoint of iron-carbon eutectic graphitization, might produce enough segregation of molybdenum and/or chromium to the phosphide to change its solidification from the graphite to the carbide form, with a drastic increase in its contraction. In such a case, internal porosity would tend to appear in every hot spot or center of mass and potentially cause leak-

#### Conclusions

ers or failures in critical castings.

1. Visible internal porosity tends to appear in unalloyed gray iron castings whenever the phosphorus content exceeds about 0.25 per cent. Higher phosphorus increases the magnitude of the spongy regions, while reductions to 0.13 per cent P produce decreasing quantities of microporosity. Quantitative data indicates that this porosity is produced by the liquid and solidification contractions of the pseudo-binary phosphide-graphite eutectic which forms a liquid network in the iron when the base phosphorus exceeds approximately 0.08 per cent and solidifies at 1753 F.

2. Molybdenum and chromium additions to gray iron sharply reduce the critical phosphorus content for visible porosity formation and subsequently increase the magnitude of voids at a given phosphorus level. Molybdenum tends to concentrate porosity in visible spongy regions, while chromium produces smaller visible voids and more widespread microporosity. Combined chromium-molybdenum additions may produce visible voids at phosphorus contents as low as 0.06 per cent. This may be explained by the rapid diffusion of these alloys to the liquid phosphide, approaching saturation contents of 8.3 per cent Mo and 2.15 per cent Cr in the eutectic. This segregation changes the solidification of the phosphide eutectic to the metastable carbide form with a marked increase in

its contraction, and sharply increases the quantity of phosphide in the network by reducing the eutectic phosphorus content from 7.0 per cent to 5.0 per cent and the solubility of phosphorus in austenite to 0.03 per cent. In addition, both alloys tend to carry excess carbide into the eutectic, thereby increasing its volume still further.

3. Straight nickel additions reduce the porosity in plain gray irons to some extent, but balancing additions to molybdenum and/or chromium irons have little or no effect on their shrinkage tendencies. Nickel exhibits no tendency for segregation, distributing about equally between the phosphide eutectic and the base iron. In the presence of 2.0 per cent Ni, the solidification contraction of the plain phosphide eutectic is changed into an expansion with a consequent reduction in its tendency for void formation. However, in the presence of molybdenum or chromium, sufficient nickel is not present in the eutectic to balance the large quantities of these segregated alloys, and it is ineffective for preventing the large contraction produced by the carbide form of solidification.

4. Internal porosity of this type persistently appears in the hot spots of a casting at given phosphorus and alloy levels despite wide variations in carbon and silicon content or inoculation treatments. It can be removed by risering only when every center of mass is moved completely out of the casting by large risers.

5. The change from the graphite to the carbide form of phosphide eutectic solidification occurs when the molybdenum and/or chromium content exceeds a definite level in the base iron. Carbide solidification first appeared in slowly cooled 4-in. cubes at about 0.50 per cent Mo or Cr, and this change might be expected to occur at lower alloy levels in lighter commercial castings due to their rapid cooling rates. Thus, a slight increase in balanced alloy content which would be undetected in standard chill tests may cause internal porosity to suddenly appear in critical castings. If porosity of this type is encountered, check for high phosphorus or alloys in charge materials, or a build-up of alloy residuals in return scrap.

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## Bibliography

1. J. T. MacKenzie, Pig Iron Rough Notes, Winter and Spring, 1936

2. A. A. Timmins, "Shrinkage and Contraction of Cast Iron," Foundry Trade Journal, vol. 59, Sept. 15, 1938, pp. 191-193.

3. W. B. McFerrin, "Casting Defects Common to Automotive Gray Iron," Transactions, American Foundrymen's Association, vol. 48, pp. 1-30 (1940-1941).
4. W. West, "Phosphorus in Cast Iron," Metallurgia, vol. 7,

pp. 83-88 (1933). 5. G. A. Timmons, "Theory of Shrinkage in Gray Cast Iron," The Foundry, vol. 70, no. 12, pp. 82-83 (1942); no. 1, p. 80

6. T. Turner, "Volume and Temperature Changes During the Cooling of Cast Iron," Journal of the Iron and Steel Institute,

vol. 69, p. 48 (1906).
7. W. F. Bohm, "Mold Materials Are Factors in Gray Iron Shrinkage," American Foundryman, vol. 19, no. 1, January, 1951, pp. 26-29.

A. Sanders and C. C. Sigerfoos, "Gray Iron Shrinkage Related to Molding Sand Conditions," AMERICAN FOUNDRYMAN, vol. 19, no. 2, February, 1951, pp. 49-55.

9. F. B. Rote, Phosphorus in Nickel-Molybdenum Gray Cast Iron, Ph. D. Thesis, University of Michigan, January, 1944.

10. E. J. Ash and C. M. Saeger, Jr., "Methods for Determining the Volume Changes Undergone by Metals and Alloys During Casting," Transactions, American Foundrymen's Association, vol. 38, p. 107 (1930).

11. E. J. Ash and C. M. Saeger, Jr., "Volume Changes of Cast Irons During Casting," Transactions, American Foundrymen's

Association, vol. 40, pp. 172-200 (1932). 12. F. B. Rote and W. P. Wood, "Segregation of Molybdenum in Phosphorus-Bearing Alloyed Gray Cast Iron," Transactions, ASM, vol. 35, pp. 402-434 (1945).

13. B. Saklatwalla, "Eisen und Phosphor, die Konstitution

ihrer Verbindungen," *Metallurgie*, vol. 5, pp. 331-336 (1908). 14. B. Saklatwalla, "Phosphoreisen, seine Konstitution," *Metal-*

lurgie, vol. 5, pp. 711-713 (1908). 15. E. Gercke, "Experimentelle Thermische und Metallographische Untersuchung uber das System Eisen-Phosphor," Metallurgie, vol. 5, pp. 604-609 (1908).

16. R. Konstantinow, "Ueber die Phosphide des Eisens," Stahl

und Eisen, vol. 30, pp. 1120-1121 (1910).

17. R. Vogel, "Ueber das System Eisen-Phosphor-Kohlenstoff," Archiv fur das Eisenhuttenwesen, vol. 3, pp. 369-381 (1929-1930). 18. F. Wust, "Beitrag zum Einfluss des Phosphors auf das System Eisen-Kolenstoff," Metallurgie, vol. 5, pp. 73-87 (1908).

19. P. Goerens and W. Dobbelstein, "Weitere Untersuchungen uber das Ternare System Eisen-Phosphor-Kolenstoff," Metallur-

gie, vol. 5, pp. 561-566 (1908).

20. O. Von Keil and R. Mitsche, "Der Einfluss des Siliziums auf das System Eisen-Kohlenstoff-Phosphor," Archiv fur das

Eisenhuttenwesen, vol. 3, pp. 149-156 (1929-1930). 21. P. Bardenheuer and M. Kunkele, "Influence of Silicon

Content and Cooling Rate on the Constitution of the Phosphide

Eutectic in Cast Iron," Mitteilungen, Kaiser-Wilhelm Institut fur Eisenforschung, vol. 12, no. 4, pp. 33-38 (1930). 22. M. Kunkele, "Formation and Composition of the Phosphide Eutectic in Cast Iron," Mitteilungen, Kaiser-Wilhelm In-

stitut fur Eisenforschung, vol. 12, pp. 23-31 (1930). 23. J. E. Stead, "Iron and Phosphorus," Journal of the Iron

and Steel Institute, vol. 58, part 11, pp. 60-155 (1900).

24. R. Schneidewind and C. Harmon, "Microstructure of Silvery Pig Iron," Transactions, American Foundrymen's Society, vol. 55, pp. 602-606 (1947).

25. C. Benedicks, N. Ericsson and G. Ericson, "Bestimmung des Spezifischen Volumes von Eisen, Nickel und Eisenlegierungen um Geschmolzenen Zustand," Archiv fur das Eisenhuttenwesen, vol. 3, p. 473 (1930).

26. K. Honda and H. Endo, "On the Volume Change in Cast Iron During Solidification," Science Reports, Tohoku Imperial

University, vol. 16, p. 19 (1927)

27. K. Honda, T. Kase and Y. Matsuyama, "On the Change of Volume of Cast Iron During Solidification," Science Reports, Tohoku Imperial University, vol. 18, p. 699 (1929).

## DISCUSSION

Chairman: F. T. McGuire, Deere & Co., Moline, Ill.

Co-Chairman: W. A. HAMBLEY, Chas. A. Krause Milling Co., Birmingham, Mich.

Recorder: C. T. MAREK, Purdue University, Lafayette, Ind.

F. G. Sefing (Written Discussion): 1 The authors of this paper have presented excellent evidence of the mechanism of gas-free porosity in gray iron castings, which have been known for some time to be almost impossible to feed.

Experience in the foundry has shown that the carbide-forming elements tend to emphasize voids and/or microporosity. With the evidence presented in the paper we can now understand the precise magnitude of these effects and that they are closely tied

up with the phosphorus content. The authors' work indicates that any influence toward a hot

spot in the casting (bosses, etc.) cannot be fed in a practical manner and, therefore, porosity of the type discussed must be approached by some method other than by feeding.

It has been the writer's experience that when the porosity described in this paper exists, there is no evidence of discoloration toward dark brown, dark gray or black colors in the porous areas. The writer would like to emphasize, therefore, that when such discoloration occurs, the foundryman should be alerted for gases from either cores or molds having been trapped in the metal. These gases afford an opportunity for shrinkage and/or porosity which can be eliminated by preventing the gases from entering the metal during the pouring operation. It is to be emphasized that these discoloration effects are not to be misconstrued as having any relation whatever to the porosity or microshrinkage discussed in this paper.

It has been known for some time that carbides and phosphides tend to unite in complex eutectics and the work of the authors corroborates these contentions. It is not surprising, therefore, that the present paper gives definite evidence that adding carbide-forming alloys in the presence of phosphides markedly increases the tendency toward microporosity. There seems to be no alternative but to maintain a proper composition of gray iron with regard to phosphorus and alloys to prevent microporosity beyond that which can be taken care of by proper practical

feeding.

There seems to be two alternatives, then, from the evidence of

Fig. 25 of the paper:

(1) That with appreciable phosphorus content (0.10-0.14 per cent) nickel should be added to counteract the contraction, and

That if carbide-forming alloys are to be added, the phosphorus should be reduced to the order of 0.04 per cent.

Dr. C. R. Austin (Written Discussion): We have been very

much interested in this paper by Mr. Hamaker and his associates. It demonstrates an excellent balance between theory and practice and substantiates many of the views which we have held for many years.

In discussing the unsoundness in gray iron castings, the authors suggest that with the advent of high test irons a distinctly different type of defect from ordinary shrinkage began to appear. We would suggest that dendritic shrinks are all basically simply dependent on liquid-solid contractions. Obviously, phosphide-eutectic shrinks cannot be fed by risering since the liquid-solid change is always delayed to a time period or temperature level, well beyond the time or temperature defining "complete solidification" of the casting. Accordingly the significance of these delayed solidification regions are directly proportional to the amount of phosphorus in the iron and the authors have most effectively portrayed and demonstrated the relative effects of various alloy additions on the magnification of the internal shrinks

It must be recognized, however, that the high-test irons are more prone to internal porosity with or without appreciable amounts of phosphorus. All metals and alloys solidify via dendritic crystal growth and the intertices can be adequately supplied with solidifying metal by a sort of self-feeding process, the success of which is dependent on assuring progressive solidification in the casting.

In pure metals and alloys, such as soft gray irons, solidifying within a very narrow temperature range, this is a relatively

The International Nickel Co., Inc., New York.
 Asst. to Pres., Meehanite Metal Corp., Larchmont, N. Y.

simple matter. The higher tensile irons, however, solidify over a fairly wide temperature range and assurance of what we have termed "self-feeding" of these potential dendritic shrinks is a

much more difficult problem.

In general, the authors have experimentally examined the effects of various elements or combination of alloying elements on the tendency or proneness of the irons to promote internal shrinks. They have correctlyy considered the magnitude of these effects, a function of the affinity of the alloys to enter into combination with the iron-phosphide eutectic, thus increasing the volume of this low temperature solidification complex with concomitant increase in shrink voids. We cannot avoid raising the speculation as to how these various alloys modify the viscosity or fluidity of the liquid complex at temperatures approximating solidification. Any element tending to increase fluidity would aid in promoting the necessary self-feeding of the dendritic structures. Elements promoting increased viscosity would evidently increase susceptibility to internal shrinks.

The experimental analysis of this phenomenon presents many problems but such determination would appear to have a clearly defined place in the theoretical discussion of the subject matter.

The facts developed in this paper have long been recognized in Meehanite practice where gating and risering technique has been scientifically developed to promote progressive solidification. Further, two factors in this process are highly advantageous in the production of castings free from internal shrinks. Meehanite castings are low phosphorus irons and their method of manufacture generally renders the addition of any alloying element to develop high tensile properties, quite unnecessary.

The importance of avoiding alloy additions wherever practical has been well demonstrated by the authors, although they have emphasized the vital role of phosphorus when attempts are made to manufacture high tensile irons by doping up with alloy

The paper warrants the most careful study by all concerned with the founding of gray iron castings and the authors are to be highly commended on a paper of unusual merit. It is a happy combination of scholarly research directed to a most practical and vital problem in the gray iron casting industry.

A. D. BARCZAK (Written Discussion): 

First, this paper cer-

tainly constitutes an excellent job of research performed in a very thorough manner and has produced some astounding facts that have baffled foundrymen for many years.

We have been able to confirm in practice many of the con-

clusions drawn by the authors.

We have not been able, however, to substantiate the statement that variations in carbon or silicon of 0.25 per cent above or below the base content had a negligible effect on the magnitude

of shrinkage voids.

The carbon equivalent of the base iron discussed in this paper is 4.00. It has been our experience that with a gray iron of 4.15 carbon equivalent when we were able to reduce the phosphorus percentage below 0.12 per cent internal shrinkage porosity was eliminated on jobs whose design and varying section size had previously given trouble. If, however, we lowered our carbon equivalent below 3.90 by reducing carbon and still maintaining a 0.10 to 0.12 phosphorus level we again experienced shrinkage voids, visible porosity and microporosity in these

The results of the work as described in this paper presents a perplexing problem for the foundry operator. In the production of high test gray iron castings as well as pressure castings, it becomes very apparent that it becomes imperative that the raw materials used must be of very low phosphorus content. With automotive cast scrap containing 0.20 per cent phosphorus and malleable pig iron having 0.15 to 0.20 per cent phosphorus it becomes impossible to produce gray iron with a 3.25 per cent carbon, 2.25 per cent silicon and an 0.08 to 0.10 phosphorus

How can the operator produce these desired irons without the necessary raw materials? How soon could the scrap and metal suppliers as well as pig iron producers present to the gray iron industry satisfactory and sufficient amounts of the desired raw materials at an economic level that would be supported by today's buyers' market?

This paper opens up another phase of the foundry industry that has had many of us baffled for many years. It has been established here as well as in Europe that there are certain inherent characteristics in pig irons which produce castings with visible internal porosity, shrinkage voids and microporosity. It appears now that a great deal of this phenomenon can be explained by phosphorus content of the pig iron used when we consider its tremendous influence on gray irons at levels below 0.10 points. Until recent times, automotive gray iron castings specifications of 0.18 to 0.20 appeared to be highly satisfactory. Today this is no longer true.

Although this paper has given us a vast amount of information on internal porosity, there are still many questions to be answered. Several months ago we obtained some foreign pig iron in which the phosphorus content was 0.06 to 0.08 per cent. The rest of the chemistry of the pig iron was within good commercial range and yet after several days use of this pig iron, we encountered many shrinks and microporosity in the castings. Changing back to a domestic pig iron of similar chemistry, eliminated the trouble within two days. What was the cause? The phosphorus percentage was the same in both pig irons as

well as the gray iron castings produced.

J. E. SRAWLEY (Written Discussion): 4 The authors of this paper deserve to be congratulated on having carried out a most useful investigation in a very able manner. It is interesting to see that their figure of 0.25 per cent for the phosphorus content at which visible internal porosity first occurs in 4-in, cube castings of plain gray iron agrees very well with the figure of 0.3 per cent given by A. A. Timmins for 3-in. diam spheres.* The demonstration that this critical phosphorus content is drastically reduced by the presence of molybdenum and/or chromium is a valuable addition to our knowledge and may explain outbreaks of this type of porosity in low phosphorus irons which have

hitherto been puzzling.

The proposed mechanism for the formation of this type of porosity is a reasonable one and the very considerable effort that has been made to verify it indirectly is highly commendable. The difficulty of carrying out accurate quantitative work in this field of research needs to be experienced to be appreciated. Unfortunately, there seems to be no possibility of demonstrating directly that porosity only forms after solidification of the ironcarbide eutectic is complete. It is pertinent to recall that A. A. Timmins found that porosity was always absent in his 3-in. diam spherical castings of 1 per cent phosphorus iron, when the metal had been treated to produce an iron with a very fine graphite structure (A.S.T.M. type D).* The treatment consisted of adding 0.15 per cent titanium and passing carbon dioxide through the melt. This process is not of practical interest for the production of sound castings, but the effect must be explained by an adequate theory of porosity formation. The mere presence of titanium is not effective unless the fine graphite structure is obtained-with a coarse graphite structure a considerable amount of porosity of the type under discussion is present. The possibility that porosity did exist in a widely distributed form in the apparently sound sphere occurred to the present writer. This was dispelled by density measurements carried out on specimens of iron of exactly the same composition and matrix structure with different graphite structures. The fine graphite iron had a slightly but consistently higher bulk density than the coarse flake graphite iron. No visible porosity was present in either. There is no doubt that the effect is connected with the mechanism of solidification of the irons, but it is not easy to see exactly how to reconcile it with the theory proposed except by introducing the further hypothesis that graphitization expansion continues right down to the phosphide eutectic temperature in the case of the fine graphite iron. Evidently there is room for further work on this point.

The difficulty, in practice, of attempting to remove this type of porosity by feeding methods is demonstrated in the paper under discussion. Confirmatory evidence of this is to be found in some unpublished work carried out by R. Jolly at the British Cast Iron Research Association a few years ago. The test piece used was a block 4 in. diam and 5 in. high, which is comparable with a 4-in. cube. The metal composition was approximately 3.4 per cent C, 2 per cent Si, and 0.5 per cent P. Altogether some 34 different methods of running and feeding were tried

³ Superior Foundry, Inc., Cleveland,

^{*} A. A. Timmins, B.C.I.R.A. Research Report 260, Journal of Research and Development, vol. 3, no. 7, August 1950, p. 458 and Fig. 4, p. 450.

4 British Cast Iron Research Association, Bordesley Hall, Alvechurch, Nr. Birmingham, England.

and in no case was it possible to eliminate the porosity. The position of the porosity varied with the size and position of the feeder so that it was always located near the heat center of the

combined casting and feeder.

W. T. Sheffield (Written Discussion): 5 To those who have had 25 or more years experience in production of gray iron castings it must seem somewhat astounding to read that this "new type of defect has come into recognition." The internal porosity or shrinkage discussed in this very provocative paper has been a prolific source of foundry headaches for as long a time as the writer has had any responsibility for production of sound castings. Perhaps the novelty, to the authors at least, is disclosed in the sentence which states that this trouble "frequently occurs throughout a foundry in a number of castings which have been satisfactorily made for years and with no marked change in the routine analysis of the iron." In other words, for such jobs internal shrinkage has been a recent development.

Despite the carefully compiled bibliographic references notes as 1, 2, 3, 4, and 5, none of which is dated earlier than 1933, the writer believes that those whose experience goes back to the early 1920's or before will corroborate his statement that internal shrinkage or porosity gave plenty of trouble in those early years. . The writer's experience as chemist, metallurgist and general Foundry Superintendent during that period covered machine tool castings, heavy machinery castings (including gear blanks weighing up to 20 tons each) and automotive work. The noxious disease, if it may be called such, commonly known as internal shrinkage or porosity was disturbingly prevalent long before the development of modern high strength irons and also long before the general use of the alloys nickel, chromium and

On May 21, 1934 the writer presented a discussion of this subject before the Pittsburgh Foundrymen's Association. This discussion was a resumé of his experiences in overcoming such troubles and covered the period from 1920 to the date of the

discussion.

All ideas are or should be subject to changes with the seasoning of added experience. After reading the preprint of this paper, the writer dug up notes for that discussion of 18 years ago. He was rather surprised to discover that added experience had corroborated the views held at that time rather than altered

The writer can give a hearty "amen" to some of the conclusions drawn by the authors from their painstaking work. He feels also that a few of their conclusions should be modified somewhat

The conclusions in which the writer concurs with the authors are as follows:

1. That the internal shrinkage phenomenon occurs after complete solidification of the Iron-Carbon Eutectic.

2. As a corollary, it is utterly impracticable to remove internal shrinkage troubles by use of risers. Risers are effective only up to the point of solidification.

3. Additions of chromium and molybdenum, either singly or in combination, tend to increase any tendency toward development of internal shrinkage which the iron may have had origi-

nally

4. Nickel additions have very little effect upon the tendency to develop internal shrinkage. If the iron intrinsically has such a tendency, addition of nickel alone, without other changes, cannot be relied upon to remove this tendency. Conversely, if the iron be free from such trouble same tendencies, ordinarily the addition of nickel will not cause such a tendency of develop. The writer has known of very special circumstances where additions of nickel or copper over a certain amount has caused internal shrinkage to develop, but these have been very special instances.

5. The tendency to develop internal shrinkage seems to be independent of final carbon and silicon contents over a wide range; irons containing less than 3.0 per cent total carbon seem to have much less tendency toward internal shrinkage than when the total carbon content is higher than 3.0 per cent.

The writer feels, however, that the conclusion that "in unalloyed gray iron castings visible internal porosity tends to appear whenever the phosphorus content exceeds 0.25 per cent" is not necessarily valid for all types of unalloyed gray irons.

1. In two different shops we were able to eliminate internal shrinkage trouble without giving any thought at all to the phosphorus content. In both shops the phosphorus content ran from 0.300 per cent to 0.450 per cent.

2. In a third shop, an automotive shop, particular attention was given to the phosphorus content, keeping such content below 0.18 per cent. Yet we had spasmodic occurrence of this trouble. We would run for weeks with no trouble and thenout of a clear sky apparently-this trouble would appear all

over the shop.

3. We are currently producing castings 30 in. in diameter, with a heavy section, 6½ in. x 3½ in. running around the perimeter, casting weight approximately 300 lb without visible internal shrinkage. The iron contains 0.80 per cent chromium, 0.60 per cent molybdenum, 3.10-3.15 per cent total carbon. The phosphorus content is 0.16-0.17 per cent, a content appreciably higher than that at which the authors expect internal shrinkage should develop with that much chromium and molybdenum

The writer realizes fully that he is flying directly in the face of much scientific research that has claimed to establish beyond a peradventure or a doubt the role that phosphorus plays in causing internal shrinkage. He does not question the pains-taking accuracy of such work. But he does feel that the validity of the deductions can be questioned when it is assumed that their particular findings must perforce apply to all gray irons whether of the same nature or not. There is much more to evaluating the nature of gray iron than to merely state its resultant chemical analysis.

To reduce the argument to a syllogistic form by which the fallacy is most readily seen, the general type of reasoning could

be stated thus:

1st premise—Cats have fur 2nd premise-Cats are animals Therefore, all animals have fur

In this particular instance, the premises and conclusions could be stated thus:

1st premise-Alloyed automotive irons and a named metal showed internal shrinkage when the phosphorus content exceeded certain levels.

2nd premise (Implied)-Alloyed automotive irons are gray irons. Therefore, all unalloyed gray irons and gray irons containing alloys as specified, tend to show internal shrinkage when the phosphorus content exceeds certain well-defined levels.

In all seriousness, the writer feels that the real fundamental causes of internal shrinkage have not been established as yet. Some of us think we know something about controlling it, but an over-all adequate theory to account for its development and control has not been established as yet.

Theories must be deduced from facts. If factual experience occur which the theories do not explain, the theory cannot be considered as universally applicable. The theory must be broadened until it covers all known facts of experience.

To be logically valid, the true explanation of internal shrinkage should cover all known discrepancies, such as

1. Why internal shrinkage does not occur in some irons with a phosphorus content as high as 0.50 per cent.

2. Why internal shrinkage occurs very often in irons contain-

ing as little as 0.14 per cent phosphorus.

3. Why castings that have been free from this trouble for years with no attention to phosphorus content (even though higher than currently considered advisable) suddenly show this trouble with no change in the resultant analysis.

A true explanation should provide for an explanation of the data presented in this paper as well as accounting for the appar-

ent inconsistencies listed above.

In the writer's opinion it is unfortunate that the authors have mentioned a metal that has been commercialized under a

He wishes to make it crystal clear that he is not criticizing the quality of such metals. On the contrary, he believes these metals to be of dependably high quality; but when the implication is made that such metals are anything other than gray iron, it becomes a distortion of fact.

There have been too many instances in actual experience where such has not been the case. As examples:

⁵ Foundry Supt., Hershey Machine & Foundry Co., Manheim, Pa.

Name them what you will, these so-called processed irons are good gray iron and nothing else. There are many foundries in this country that are producing gray irons of equal merit.

MESSRS. HAMAKER, WOOD AND ROTE (Authors' Written Reply and Closure): The authors wish to express their appreciation to Messrs. Sefing, Austin, Barczak, Srawley, and Sheffield for their

very interesting and stimulating discussions.

Several of the comments have indicated that a few points might have been more clearly emphasized in the paper. First, it was not intended to imply that phosphide eutectic contraction is the only source of internal shrinkage or porosity in gray iron castings. In addition to the discolored gas porosity from molds or cores pointed out by Mr. Sefing, several other sources of den-

dritic internal voids should be recognized.

Perhaps the most common, and frequently misinterpreted, is ordinary shrinkage due to iron-carbon eutectic contraction. Internal voids from this source have exactly the same appearance as those from phosphide eutectic contraction and the difficulties encountered in attempting to control them have frequently caused phosphide shrinkage to be blamed for their inconsistency and persistence. Many castings are of such design or are risered in such a manner that they represent borderline cases from the standpoint of directional solidification. In such cases, a very slight change in pouring temperature, carbon equivalent, graphitic carbon content, or depth of chill may alter the temperature gradients, the interference of dendrites to the flow of feed metal, or the solidification contraction sufficiently to produce an ironcarbon shrink where none had appeared previously. A particularly puzzling example of this shrinkage occurs when a pocket of sand or a core retains the heat to such an extent that it retards solidification in a section which is seemingly well fed from adjacent heavier sections. Such defects frequently disappear and reappear with small variations in pouring temperature.

Another source of internal voids which should be recognized appears to arise from "oxidation" or "gasification" in the cupola. Although the voids may have a distinctly dendritic appearance, they tend to be scattered randomly throughout the casting rather than concentrated in localized hot spots like iron-carbon or phosphide shrinkage. A few ladles of "oxidized" iron can, of course, produce defective castings throughout a foundry similar to phosphide shrinkage. The positive identification and control of this "oxidation" in the cupola would certainly be a worthy

subject for investigation.

A second point the authors would like to emphasize is the metastability of the phosphide eutectic. A content of 2 per cent (or more) carbon and approximately 1 per cent silicon makes it behave quite similarly to cast iron with regard to the forms it may assume on solidification. The carbon in the phosphide eutectic may solidify entirely as carbide with a high contraction similar to white iron, it may partially graphitize with a moderate contraction similar to pearlitic gray iron, or it may graphitize completely with little or no net contraction similar to a ferritic gray iron. Like cast iron, the form of phosphide solidification depends on section size or cooling rate, inoculation treatments or the graphitizing tendency of the surrounding iron, and the percentage of carbide stabilizing alloys it contains. Thus, a sound casting may be obtained at phosphorus and alloy levels above those suggested in the paper if its cooling rate is sufficiently retarded and the graphitizing tendency of the surrounding iron is great enough to prevent the phosphide eutectic from solidifying in the carbide form. Similarly, high phosphorus plain irons which fall into the ferritic class may produce a completely graphitized phosphide eutectic with much smaller contractions than those obtained in the paper for the plain phosphide eutectic containing 0.65 per cent combined carbon as pearlite.

Another point to be kept in mind is that the total volume of voids resulting from a given phosphide contraction is proportional to the mass of iron in the hot spot of the casting. In other words, if a hot spot has only one-half the mass of a 4-in. cube, its void volume would be about one-half that computed

for a 4-in, cube of the given composition.

It is hoped that the few examples given above will help to explain how porosity similar in appearance and behavior to phosphide shrinkage may arise from other sources, and conversely, how an iron having phosphide shrinkage tendencies may produce a sound casting under certain conditions. However, it is the authors' contention that phosphorus and alloys must be held below the levels set forth in the paper to consistently produce sound engineering castings from pearlitic gray irons. Complex modern castings frequently have potential hot spots with cooling rates equivalent to or greater than the 4-in. cube test specimen from which the conclusions were obtained.

Dr. Austin presents an interesting point in his speculation as to the possible effect of phosphide eutectic fluidity on internal shrinkage. Although increased fluidity promotes soundness at the iron-carbon eutectic temperature, there is some question in the authors' minds as to its influence at the phosphide eutectic temperature. If we assume that the phosphide network in a hot spot has been completely cut off from a riser or other parts of the casting by phosphide solidification at its extremities, the fluidity of the remaining liquid would have no effect on the total volume of the resulting voids. The chromium phosphide, having a relatively high freezing point and probably low fluidity, does not travel through the network to any extent and produces scattered microporosity. The molybdenum phosphide, having a low freezing point and probably high fluidity, moves readily through the network to produce concentrated voids at the point of final solidification. It appears that high fluidity in the phosphide could improve soundness only if the network through the riser neck could be kept liquid until most of the phosphide eutectic in the casting had solidified. This would probably require very large risers similar to those found necessary to move the hot spot out of the casting.

It is hoped that the opening remarks of this closure will clarify Mr. Barczak's point concerning a relationship between internal shrinkage and carbon equivalent. Both iron-carbon and phosphide shrinkage may frequently occur in the same casting and have identical appearance, but arise from different sources. It appears that the problem of low phosphorus raw materials can best be met by segregation of the plain and alloy iron castings in a foundry. Plain gray irons, which can safely contain up to 0.25 per cent phosphorus, may be developed to meet most of the strength specifications, and the premium castings requiring alloy additions for special properties may be poured from low phosphorus cupola charges, consisting of Bessemer pig iron, steel, and selected scrap. It is imperative, however, that the alloyed return scrap be kept segregated from the higher phosphorus plain iron charges. With regard to the shrinkage resulting from a lot of foreign pig iron, there is always the possibility of small percentages of tramp elements whose effect on gray iron are not known. Increased use of spectrographic analysis may eventually identify these elements and prevent their introduc-

tion into the foundry.

The corroborative British data reported by Mr. Srawley is a valuable addition to the paper. Concerning the absence of visible porosity in 3-in. spheres containing I per cent phosphorus, the authors wish to submit the following explanation

consistent with the proposed theory.

At 1 per cent phosphorus, the iron-carbon eutectic temperature is reduced 65 to 70 F (Ref. 12 in the paper) below the average value of 2050 F used in the calculations for the 0.04-0.30 per cent phosphorus range under investigation. It should be fairly safe to assume that the maximum graphite expansion from saturated austenite and the internal pressure conditions which form exuded beads are similarly moved to lower temperatures. Thus, the temperature at which liquid phosphide contraction begins to produce voids might be reduced from the 1900 F value used in the calculations to about 1835 F.

Type D graphite is usually associated with a ferritic matrix and is indicative of a strong graphitizing tendency. This tendency could persist into the phosphide eutectic solidification, as evidenced by the influence of strong inoculation on the structure of the chromium-phosphide beads. In addition, higher phosphorus iron shows a definite tendency for increased graphitization of the phosphide eutectic (Refs. 21 and 22 in the paper). Thus, there is a strong possibility that the phosphide eutectic graphitized completely rather than retaining 0.65 per cent pearlitic combined carbon as in the pearlitic low phosphorus irons studied. Using Ash and Saeger's formula, this would reduce the solidification contraction of the plain phosphide eutectic from 0.87 per cent to only 0.23 per cent by volume and give a value of 0.1468 cc/g for the specific volume of the solid phosphide at solidification.

Assuming an 1835 F starting temperature and complete graphitization of the phosphide, the total liquid and solidification contraction of the plain eutectic would then be (0.1482 - 0.1468) = 0.0014 cc per gram of eutectic. The entire 3-in. sphere weighs about 1650 grams and a rapidly solidified 1/4-in, layer on the side and bottom surfaces would leave a mass of about 1400 grams contributing to voids through a phosphide network. Computing as before, the void volume in the 3-in. sphere would be  $1400 \times 0.0014 \times (1.00 - 0.08)$ 

= 0.26 cc. This void volume is  $0.07 \times 100$ 

below the computed value of about 0.39 cc at which visible voids first appeared in the 4-in. cube.

The greater density observed with Type D graphite than with flake graphite may be explained by the marked undercooling of the iron-carbon eutectic necessary for the formation of pseudoeutectic graphite structure. This undercooling, ranging from 50 to 75 F,* might reduce the starting temperature for void formation from liquid phosphide contraction to 1785 F in a 1 per cent P iron, with a subsequent reduction in computed void volume

The preceding remarks should help to clarify Mr. Sheffield's questions regarding exceptions to the phosphorus and alloy limits for avoiding internal shrinkage. In suggesting these limits, the authors should have stated that they had in mind modern engineering castings in high pressure and stress applications which are being designed for medium to high strength pearlitic irons and require nearly perfect density and soundness for satisfactory service. The straight-forward mechanism by which a given volume of voids (either visible or micro) and the confirmatory results obtained in numerous industrial studies provides a logical justification for expecting internal shrinkage tendencies in any pearlitic gray iron at the levels indicated. There are undoubtedly many high phosphorus castings in service today in which the graphitizing tendency of a low strength iron was sufficient to graphitize the phosphide eutectic completely and prevent void formation as described above, or in which the voids were not detected simply because they did not occur at a critical point in the casting.

* Schneidewind, R., and D'Amico, C. D., "The Influence of Undercooling on the Graphite Pattern of Gray Cast Iron," Transactions, A.F.A., vol. 47, pp. 831-853 (1939).
Schneidewind, R., and D'Amico, C. D., "The Solidification and Graphitization of Gray Iron," Transactions, A.F.A., vol. 48, pp. 775-803 (1940).

# MELT DEPARTMENT INCENTIVE PLAN

By

#### Erwin G. Tetzlaff*

While it has been long conceded that the best incentive plans are individual plans based on standard data resulting from adequate and valid time studies, it is often found that there are many foundry operations in which the time study approach is difficult or unfeasible. The melt department is typical of that type of situation.

At the plant with which the author is associated, where the policy to to get every employee possible under some sort of a workable incentive plan, a statistical approach is often used where the time study approach is impractical. Using the statistical approach, past performance is measured and incentive plans which offer premium pay for improved performances are then established. The plans are so established that the company benefits as well as the employees. Such tailor-made plans have been used in the plant for many years with very happy results.

The melt department incentive plan was set up using the statistical approach. However, it should be pointed out that while the standard values may not be applicable to other foundries, it is believed that the pattern of the plan can be tailored to suit most foundries.

## The Melt Department

To understand the plan it is important to have some knowledge of the department. The  $1\frac{1}{2}$  ton top charge acid electric furnace is always charged with 8000 lb of steel scrap plus, of course, the necessary furnace additions. Because of this the melt cycle times are and should be fairly constant.

The furnace is operated around the clock with three 8-hr shifts. Each shift has one melter and one helper. The foundry demands are such that the melt department must operate at top efficiency at all times. As proof of this it can be pointed out that with a 5-day-week operation well over 600 tons of good casting per month are produced in the shop, which is strictly a jobbing foundry.

While quite a variety of steels are produced, no high-alloy castings are made. The chemical laboratory makes a chemical analysis of each heat, and also runs preliminary carbon analyses as requested by the melters.

#### Three-Part Plan

With this picture of the department and operations in mind, it can now be explained that the plan consists of three parts:

1. Quality Portion in which each melter-helper team is on a separate basis. This part carries the most weight and offers the greatest opportunity for high incentive earnings.

2. Quantity Portion, or the man hour per heat portion, in which all melters and helpers share as a group.

3. Power Savings Portion in which all melters and helpers share as a group.

The incentive pay is calculated and paid weekly. The three parts are added up to give the total percentage of bonus for each melter and helper, as will be illustrated later in the paper.

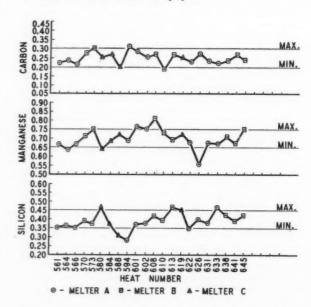


Fig. 1—Melters' quality control chart —.20-.30 carbon steel.

[•] Foundry Engineer, The Pelton Steel Casting Co., Milwaukee.

TABLE 1—Sample Point Value Charts—Melter's Ouality Bonus

		Bonus I	Points	No Penalty No Bonus		ty Po	ints
Type Steel	Component	Range	Pts.	Point Ranges	Rang	e	Pts
20-30C	Carbon	.2030	+4	-	Over Under	.30	_9 9
Common	Silicon	.3545	+1	.4655	Under .2529 .3034 Over		
	Manganese	.6575	+3	.6064 .7685	Under .5054 .5559 Over		9677
Mo-Mn	Carbon	.3040	+5	-	Under .2829 .4142 Over		-5 -3 -3 -5
	Silicon	.3545	+1	.4655	Under .2529 .3034 Over		_9 _6 _3
	Manganese	1.15-1.35	+3	1.05-1.14 1.36-1.45	Under Over	1.05 1.45	_3

Quality Portion—Development

For several years prior to the plan statistical quality control charts were employed in the melt department. On these charts are plotted the analysis results for each chemical determination for each heat (Fig. 1). Thus at a glance one can quickly tell whether or not the analysis fell within the specified range and the deviation, if any.

The seriousness of "off analysis" heats is well known to every foundryman. While often slight deviations of certain components do not make the castings unsalable, on the other hand entire heats must often be scrapped for failure to produce the required chemical and physical specifications. This is somethimes due to the refusal of the customer to accept the castings, and sometimes the castings show defects such as extreme porosity which makes them unsalable.

"Off analysis" heats can seriously affect that very important customer's good will. Nothing will irritate a casting purchaser more than to run into a batch of castings which are hard to machine. Proper heat treatment is often difficult and sometimes impossible if the melt heat does not meet the required chemical specifications.

With this picture in mind, the foundry is now ready to set up point value charts for each chemical and for each type of steel produced. Typical charts as used at the author's foundry are illustrated in Table 1.

The assignment of point values and penalties should not be handled lightly. Rather, it should be based on a careful evaluation of the importance of each metal component and the importance of each deviation from the required specification. This evaluation should not be done by one person. It should instead be done by a group of qualified people such as the plant metallurgist, melt superintendent, inspection department head, works manager, sales manager, etc. It should be kept as simple as possible for ease of calculation.

#### Quality Portion-Mechanics

Quality bonus calculations are now simple. After the final chemical analysis of each heat has been run, bonus points or penalty points can be assigned according to the tables as set up. The results of each heat can then be tabulated for each melter-helper team as illustrated in Table 2. At the end of a week, totals can be found for the *points earned* and of the *possible points* that could have been earned.

With these figures the per cent of efficiency can then

be found, using the formula:

At this plant, after finding the per cent efficiency the bonus caluculations are completed by using the formula:

% Quality Bonus = 0.8 (% efficiency—34%) This formula is taken from the curve in Fig. 2, which was developed in the following manner:

For a 6-month period prior to the plan, normal efficiency was found to be 65 per cent. It was also decided after a plant-wide study of the wage structure and incentive programs that a maintenance of past normal efficiency should pay a bonus of 25 per cent with about a 50 per cent top bonus for 100 per cent efficiency. The curve was established although the top was increased to 53 per cent in order to avoid carrying the variance into too many decimal places. The development of the curve, in the author's opinion, must be made to suit the situation of each foundry that may wish to adopt this sort of a plan.

#### Quantity Portion-Development

The quantity or man-hour per heat portion of the melt department incentive plan is made possible by the fact, as previously stated, that each heat melted is of the same size, and the melt times are or should

TABLE 2-MELTER'S QUALITY BONUS CALCULATIONS

Specifi-	Heat	Carb	on	Manga	nese	Silic	on	Points	Total Pos- sible
cations	No.	Anal.	Pts.	Anal.	Pts.	Anal.	Pts.	Earned	Points
27-37 C	2044	.32	+4	.74	+3	.46	0	7	8
27-37 C	2045	.35	+4	.68	+3	.45	+1	8	8
40-50 C	2046	.47	+6	.68	+3		0	9	10
10-20 C	2047	.16	+3	.76	0	.28	6	-3	7
Mo-Mn	2048	.33	+5		+3	.38	+1	9	9
20-30 C	2059	.26	+4		+3		+1	8	8
20-30 C	2060	.22	+4		+3	.46	0	7	8
Mo-Mn	2061	.38	+5	1.37	0	.37	+1	6	8
27-37 C	2062	.36	+4	.72	+3	.35	+1	8	8
10-20 C	2063	.21	0		0		+1	1	
20-30 C	2074	.31	-3	.70	+3	.44	+1	4	7 7
40-50 C	2075	.42	+6	.71	+3	.45	+1	10	10
20-30 C	2114	.21	+4	.63	- 0	.37	+1	5	7
27-37 C	2115	.31	+4	.73	+3	.43	+1	8	8
TOTALS					,			165	220
Points Ea	rned	= 10	55	= 759					

be almost identical. The importance of this portion should be obvious. In normal times it tends to control the working hours. In times like the present it helps to bring about a maximum of production in each 24-hr work day to supply a highly geared foundry.

In order to get a maximum of production for each day, the melt crews exercise every precaution to prevent delays and breakdowns. Very little time is consumed between the tapping of one heat and the charging of the next. Excellent teamwork prevails so that there is very little furnace heat loss.

All preventive maintenance measures are religiously observed, and when breakdowns or furnace damages occur, remedial measures are quickly taken. Necessary patching materials and spare parts are al-

ways kept in readiness.

The development of the man-hour per heat portion was again made by the statistical method. Again an average was taken for a 6-month period prior to the bonus plan and was found to be 3.66 man-hours per heat. This figure thus became the productivity standard. For any given week the "allowed hours" equal the number of heats melted during the week times 3.66.

#### Quantity Portion-Mechanics of Calculation

In all group bonus plans based on man-hours or time saved, the company has long employed what is commonly known as a 75-25 split, i.e., the employees are paid a per cent bonus found by dividing 75 per cent of the saved hours by the standard or allowed hours. The formula is as follows:

Using the figures for a typical week during which 80 heats were melted with a total of 250 man-hours, the calculations are as follows:

$$\frac{0.75 \text{ (80 Heats} \times 3.66 - 250)}{80 \text{ Heats} \times 3.66}$$
 ==11.0% Quantity Bonus

#### Power Savings Portion—Development

A study of the per heat power consumption during the bonus base period showed a considerable variance in the amounts of power required. It was soon learned that alert operation could result in power savings. Consequently, it was decided to set up a power savings portion in the bonus plan in which the melters with their helpers and the company shared the savings on an approximate 50-50 basis.

To find the per cent of bonus to be paid for each kilowatt per heat saved the following formula, allowing a 50-50 split, was used:

Ave. base pay per heat times X.

Calculations revealed that each kilowatt per heat saved should pay 0.075 per cent. The average power

consumption was 2130 kilowatts per heat, and the formula becomes:

Again using typical weekly figures in which 164,000 kilowatts were consumed to make 80 heats for an average of 2050 kilowatts per heat, the bonus calculations are as follows:

$$(2130-2050) 0.075\% = 6\%$$
 Bonus

#### Calculating Total Bonus

After making bonus calculations for each portion of the three-part plan as illustrated, the problem of getting a total earned bonus for each melter and each

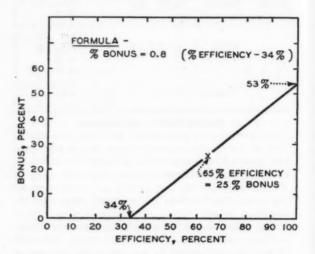


Fig. 2—Melters' quality bonus plan—per cent bonus to per cent efficiency relationship.

melter helper is simply a matter of addition. This can be illustrated by the bonus earnings for a typical week as shown below:

		Bonus Earned, %							
Name	Shift	Quality	Quantity	Power	Total				
Melter A	lst	32.1	11.0	6.0	49.1				
Helper A	1st	32.1	11.0	6.0	49.1				
Melter B	2nd	28.6	11.0	6.0	45.6				
Helper B	2nd	28.6	11.0	6.0	45.6				
Melter C	3rd	29.2	11.0	6.0	46.2				
Helper C	3rd	29.2	11.0	6.0	46.2				

The bonuses are calculated and paid weekly, and are included on the regular pay check. While the per cent of bonus may be the same for a melter and his helper, there is still a bonus pay differential in that their base rates differ.

### Summary

In conclusion, it is again pointed out that the statistical approach was used in setting up this three-part melters' bonus plan because a time-study approach was unfeasable.

The plan is tailor-made for this foundry. However, it is believed that the pattern can be used by other foundries with variations based on the statistical findings. Perhaps other factors can and should be added.

The plan has worked well in this plant. Since its installation there have been few serious "off analysis" heats. Even slight specification deviations occur less frequently.

The melters and helpers are now more alert—always seeking better controls and always confering with the melt superintendent on ways to improve performance. While there is a spirit of competition, there is also effective cooperation.

It all adds up to more earnings for the participants and better castings as well as savings for the company.

#### Acknowledgment

The author wishes to acknowledge the great help rendered by Walter E. Brandt, metallurgist of the Pelton Steel Casting Co.

#### DISCUSSION

Chairman: J. E. HYLAND, John Deere Spreader Works, E. Moline, Ill.

Co-Chairman: C. J. PRUETT, McWane Cast Iron Pipe Co., Birmingham, Ala.

Recorder: C. J. PRUETT.

A. K. Fenn (Written Discussion): 1 I quite agree with Mr. Tetzlaft's concession that the best Incentive Plans are of an individual basis resulting from standard data established by adequate and valid Time Studies. I will further agree that there are many Foundry operations which are difficult, but not impossible, to place on an individual plan. A company policy of placing every employee possible on incentive is indeed commendable, and it is our policy in this plant, where we are doing our utmost to have every productive employee on incentive work 100 per cent of the time.

Mr. Tetzlaff and his aides have apparently put a great deal of time and thought into the development of this incentive plan and in breaking it up into a three-part plan, it shows a very

well balanced set-up.

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Quality Portion-It is pointed out by Mr. Tetzlaff that past performance averages for both the quality and quantity portions were accepted as the allowed standards and in the case of the quality portion, the meeting of this standard would pay a 25 per cent bonus. I fail to see the reasoning behind the choosing of these averages as standards. It is a commonly accepted fact that day work, and it would appear from Mr. Tetzlaff's paper that his Melt men were on day work previous to the installation of this incentive plan, runs approximately 60 to 65 per cent efficiency. This ties in with Mr. Tetzlaff's statement that for the 6-month period prior to the installation of the plan, normal efficiency was found to be 65 per cent. Why then should any company be willing to pay a 25 per cent bonus for only a 65 per cent efficiency? Would it not be more reasonable to choose a higher efficiency point as your starting point at which to pay bonus, thus creating an incentive to your Melt men to raise the quality of their product? At the present time you are in effect paying a 25 per cent dividend to your operators in order to achieve a 65 per cent quality product. It is undoubtedly the policy of the company to try and achieve a product that is always 100 per cent in quality. This may seem improbable but it is not necessarily impossible. It should be the desire of your Melt men to try and achieve this target at all times and not to expect a 25 per cent bonus for achieving only 65 per cent efficiency. If your basic wage structure is sound and in line with comparative industries, then I see no reason why it is necessary to pay such a high bonus.

Quantity Portion—Under the "Mechanics of Calculations for the Quantity Portion Bonus" you state that it is the company policy to use a 75-25 split. I presume from your statement that you mean on any time saved by the employees in the operation of their duties then the company is only willing to pay them a 75 per cent portion of the savings and that the company takes the other 25 per cent as their share. This is something with which I cannot agree. Any employee, who through increased efficiency, can realize a saving for the company should be entitled to that saving in whole and the company should be satisfied with the saving they realize through a possible reduction in man hours and an assured reduction of melting costs. A worker, whether justified in his thinking or not does not like to share the direct result of his greater output.

Power Savings Portion—The power saving portion of the plan is certainly a very fine development but I still cannot agree with the idea of splitting the savings between the worker and the company, although in this portion of the plan there certainly

are arguments in favor of such a system.

Summary-In conclusion I will again say that I believe this to be a very fine and well thought out incentive plan but as Mr. Tetzlaff points out the standard values must be patterned in order to suit the individual foundry. It has not been my intention to criticize unjustly the mechanics of this incentive plan but rather to offer some constructive criticism which will bring to light debatable points that may be thoroughly discussed and answered at this meeting. Undoubtedly Mr. Tetzlaff and his associates have very strong arguments in favor of the standards that they chose and the bonuses that their company is willing to pay, and it will be interesting and informative to hear these arguments presented. Every company has its own wage structure and policies and with them they must be prepared to meet com-petitive prices in comparative industries. Should they be able to do so, and show a good percentage of profit at the year end, then the company is on a sound footing and will enjoy a long term of prosperity, but only if a proper ratio is maintained between base rates and earned bonuses. Never should loose standards and high bonuses be used as an excuse for low base rates or vice versa. Sooner or later the incentive system will run away with itself and create a major labor problem. Workers will compare base rates only with other industry and if they are low will soon begin to agitate for increases and still insist on enjoying their easily earned bonuses. When that happens the company will soon find itself unable to meet competitive prices and unable to profitably continue operations.

MEMBER: This Tentative Induction Furnace Incentive System includes the operation, and maintenance of refractory linings, of four Ajax induction furnaces and the making up of charges

for these furnaces.

The crew consists of: One Furnace Operator One Furnace 1st Helper

Two Furnace 2nd Helpers

The duties and responsibilities of the members of this crew are here briefly discussed to impress the fact that the efficiency with which each crew member performs his duties does affect the earnings of the entire crew.

The furnaces should be operated in close coordination in an effort to have one furnace ready to tap at a time thus reducing

"hold-back" to a minimum.

Relining and tightening of furnaces is done by members of the crew when other duties do not require their full attention. Yet, it is recognized that there are times when this work must be done on non-productive time. In such cases one or more members are directed to come in earlier than the regular starting time. Pay for such work performed outside of productive time is computed individually at their non-incentive pay rate.

The 2nd Helpers work in the "weigh-up" and furnace areas as necessary to properly perform duties required of them in

those areas.

DS-D12 only

In making up charges, care is exercised to place the part of the charge to go into the furnace first, on top. Where much of the charge is made up of scrap, the small, compact pieces, i.e., those which reduce the air space to a minimum, are placed on top in the charge tub.

#### INDUCTION FURNACE MELT STANDARD

Ajax Furnaces No. 1 (800-lb capacity) and No. 2 (1200-lb capacity)

All alloys other than DS-D12

333 KVA

Man Minutes per 100 lb

16.5

19.4

¹ Timestudy Sup., Crouse-Hinds Co. of Canada Ltd., Toronto.

Ajax Furnaces No. 3 (1200-lb capacity) and 666 KVA No. 4 (2500-lb capacity) All alloys other than DS-D12 11.4 DS-D12 only 13.5

#### Examples:

800-lb heat of "D20," furnace No. - = 132.0 standard minutes or time of 66 minutes from tap to tap.

1200-lb heat of "D20," furnace No. 2 = 198.0 standard minutes or time of 99 minutes from tap to tap.

1200-lb heat of "D20," furnace No. 3 = 136.8 standard minutes or time of 68.4 minutes from tap to tap.

2500-lb heat of "D20," furnace No. 4 = 285.0 standard minutes or time or 142.5 minutes from tap to tap.

The standards are computed as though all heats were poured into 200-lb ladles; therefore, when poured into bull-ladles, the crew gains extra time. At the present approximately 10 per cent of the heats are poured into bull-ladles. Should this quantity be increased, the company reserves the right to adjust the standard accordingly.

Wash and pill heats are computed the same as production heats, except, when one operator alone produces such heats outside of scheduled crew time, he then will be paid the nonincentive rate for that time.

Time for unavoidable delays, such as when directed to wait for molds, shall be computed as down time and where one furnace only is affected this time will be multiplied by two. Where both furnaces are delayed, this time will be multiplied by four, thus changing the time to man minutes to be divided equally among the entire crew in the computation of earnings.

This incentive system is also applicable for the operation of

one furnace at a time with a two-man crew.

In case of "Off-Analysis" heats, to the extent that they must be scrapped, and it is definitely shown that through carelessness or neglect a member, or members, of this crew is at fault, pay is computed at the non-incentive rate for such heats. Determination of fault will rest with the foundry metallurgist. As several days may elapse before final analysis is determined, deduction will, in most cases, be delayed.

#### PENALTY FORMULA

$$Y = SX$$
  $X = \frac{(a-b)}{a}$ 

Y = Crew penalty, in Standard Minutes, for "Off-Analysis"

S = Standard Minutes paid for "Off Analysis" heat.

- X = Ratio of man minutes gained in per cent of actual man minutes spent to produce heats on day of "Off
- a = Incentive earning rate of crew, in per cent, on day of "Off Analysis" heat. Should this be 100% or less, no penalty can be applied.

b = Standard Time quoted as 100%.

#### Example:

Y = ? S = 198 Standard Minutes 
$$X = ?$$
  
a = 145% b = 100%  $X = ?$   

$$X = ?$$

$$X = ?$$

$$X = 198 \times X$$

$$X = \frac{(145 - 100)}{145} = .310$$

 $Y = 198 \times .310 = 61.4$  Standard Minutes

#### Proof:

198 - 61.4 = 136.6 minutes (actual man minutes to produce heat) 136.6 × 145% = 198 Standard Minutes

It is urged that at all times this crew work as a unit towards

the goal of tapping heats. The normal production starting time for all members of this crew will be 6:30 A.M., but this may be occasionally changed to coincide with the production schedule.

Penalty for "Off Analysis" heat is the difference between Standard Minutes allowed and actual time taken.

$$Penalty = Standard Minutes - \frac{Standard Minutes}{incentive earning rate in per cent}$$

#### Example:

Standard Minutes = 198 Incentive earning rate = 145% 198 198 — — = 198 — 136.6 = 61.4 Standard Minutes Penalty Per cent incentive earning rate — 100 Standard Minutes X Per cent incentive earning rate = 198  $\times$  .31 = 61.4 Standard Minutes Penalty

When incentive rate is 100% or less no penalty is applied.

- J. R. VOGAN (Written Discussion):2 There are three basic ideas presented in Mr. Tetzlaff's paper which show alert and progressive thinking on the part of the author. They are:
- The incorporation of "quality" into the incentive plan. 2. Incorporation of what is generally considered overhead (power). This is, at least to some extent, a variable governable by the melters and thus is rightfully a part of an incentive plan. 3. The use of analysis of data to make easier the work of

evolving an acceptable incentive program.

On these three points, the paper merits praise. However, I would like to touch off the discussion by suggesting improvements and refinements to the author's methods of obtaining some of his goals, and by directing several questions to the author for specific answers to clarify points not immediately

apparent from the preprint. The author's assignment of point values and penalties which are the basis for the quality bonus were "based on a careful evaluation of the importance of each metal component and the importance of each deviation from the required specification." While there are bound to be differences of opinion among various manufacturers as to the relative importance of different components, it is of interest to point out the theory of the beneficial effects of higher silicon and manganese contents as expressed in the CAST METALS HANDBOOK, page 255, the factor of inter-relationships when considering the influence of individual elements, and the importance of sulphur and phosphorus, which were apparently not included in the author's calculations. Pertinent excerpts from this source are: "-Manganese and silicon have important deoxidizing influences on the molten metal. In making steel for castings it is desirable to degasify the metal as thoroughly as possible. For this purpose it is distinctly helpful to introduce liberal amounts of silicon and manganese, in the form of alloys containing these elements. Steel casting producers have viewed with satisfaction the dissipation of the former erroneous idea held by some customers of steel castings, that a manganese content appreciably higher than 0.75 per cent was likely to result in brittleness. The effects of manganese in steel as high as 1.75 per cent, -have brought a merited appreciation of the beneficial effects of the element. To some extent, relatively high silicon content has been recognized as helpful in the same general two-fold way; that is, for deoxidation and for affecting physical properties. Those steel founders who have been taken into consultation when specifications for steel castings have been in the process of development, have emphasized the important factor of relationship when considering the influences of any element in the chemical composition. A high proportion of silicon, carbon or manganese, to a large extent is helpful or harmful according to the proportions of the other elements present. It is fundamentally

wrong to evaluate them on any other basis. It is a logical pro-

² American Brake Shoe Co., New York.

cedure to specify physical properties and maximum limits for phosphorus and sulphur, while prescribing minimum limits for silicon and manganese; and generally to allow the producer to follow his own judgment regarding the other constituents included in material intended for carbon steel castings."

#### QUALITY PORTION

1. The author's statement that "melt cycle times are, and should be, fairly constant" brings up the questions:

(a) What is "fairly constant"?

(b) What has been the variation in the type, and in the size, of the scrap charged into the furnace over the past few years?

It is my impression that the scrap market has fluctuated so widely during the last two or three years, that this is one quantity which might be considered as a variable instead of a constant. For instance, the melters would be penalized if the manganese content of the scrap fluctuated, since they would apparently have no way of knowing to change the amount of the manganese addition.

2. This leads to the next suggestion. It would seem to be more fair to base the reward, or penalty, for adherence to chemical specifications upon how much the melter had control over the end result, rather than solely upon the importance of that factor upon the acceptance or rejection of the heat. Specifically, if the carbon of a heat were outside the specifications, the melter should receive a relatively more severe penalty, since not only is the carbon content easier to control, but also a preliminary carbon check is available upon request. Therefore, carbon should practically never be off-analysis.

3. There is more reason for increasing the penalty points approximately exponentially, rather than linearly, e.g., as the silicon regresses from the specifications in 5-point steps, increase the penalty points: —1, —3, —9; or —2, —6, —18; rather than —3, —6, —9. Specifically, these points should be increased inversely as the probability of getting the "off-analysis" due to random, uncontrollable, fluctuations inherent in the process itself. Regardless of any desired limits to fluctuations in the percentage of each element, the process itself has a standard deviation and the only way to decrease those "built-in" fluctuations is to eliminate the causes. It is true that one of the causes could be sloppy control by the melter, but this cause should be differentiated from all other causes.

Why did the author take the average efficiencies of 65 per cent (which he calls "normal") and give that rate a 25 per cent bonus? Why not just increase base rates 25 per cent?

(a) It is quite possible that an arithmetic mean (65 per cent) is the only feasible method of getting an index of operations in this plant. However, it would be much more accurate to make check time studies (take a sample) and, in this way, give the average value a base index, rather than just arbitrarily assigning an index of 125 per cent of base quality.

#### QUANTITY PORTION

1. In the quantity portion the arithmetic mean (average) in a six-month period was accepted as the standard.

(a) If there were an incentive plan of some sort during this six-month period, the melters are being penalized.

(b) If there were no incentive then, it would seem to be an optimistic viewpoint that the melters were working at the desired, base, day-work rate.

The solution would again seem to be a small number of time studies to check the melters pace rate, and to thereby establish a fair base for the comparison of future production.

2. Since there would be definitely different melt times for different analyses of steel (even if only slight), this variable might be eliminated by putting a time clock on the furnace, and paying an efficiency bonus on the decrease in down-time between heats.

What was the author's reason for splitting the quantity bonus between the company and the men? Since the company gains from the increased efficiency and reduces overhead, why not use a 100 per cent bonus plan for the quantity portion? (If the answer is "company policy," what is the reason behind that company policy?)

#### POWER SAVINGS PORTION

 (a) The average power consumption is 2130 kw per heat. What were the fluctuations in the weekly average during the six-month period which was used as a base? (What was the high weekly average, and the low, during that time?)

(b) What percentage of this total variance is due to the efficiency, or inefficiency, of the melters? What percentage was due to the melters? to different analyses? to variance

in scrap? etc.?

It would seem that the bonus should be based on only what the melter can control, and should not include factors over which he would have no influence. Again, a very small number of time studies (very much fewer than would be needed to develop the data) could be used to establish an estimate of the controllable deviations.

2. What provision has been made for the possibility of a negative power bonus?

#### CONCLUSION

Since the author admits that time studies are the best basis for incentive plans, it would seem better to sample the process by means of time studies, and use these studies to establish the base (normal, standard) for the past production averages, rather than arbitrarily decide what is normal.

It would also be helpful to apply mathematical statistics to the analysis of the data, in order to differentiate between the different types of deviations, so that the incentive plan could be based upon what the melter, himself, can actually control.

Let me repeat that, while I do not agree on several of the specific methods, or steps, used to develop the plan, the basic ideas are very sound. There is also much to be said for the development of incentive plans by the use of past production data as a first approximation, rather than using it as a check.

This method is faster, and therefore cheaper; although, as much care must be used in establishing the base, or normal, as in establishing the base pace used in rating timestudies. It was indeed interesting to see that at least one company has started an investigation into this precedure for developing incentive plans.

MR. TETZLAFF (Reply to Mr. Fenn): Mr. Fenn's criticism on the Quality Portion of the payment of a 25 per cent bonus for the maintenance of the past normal efficiency of 65 per cent is a very logical one. However, we wish to point to the sentence that says that the 25 per cent figure was arrived at after a "study" was made. The truth is that up to the inception of the three-part plan, we had a very crude production bonus in effect, so we had to maintain, to an extent, the earnings afforded by the old plan. Other considerations were the earnings of melters and helpers in the Milwaukee area and also the earnings of equally rated jobs in our own plant. Our base rate curve based on job evaluation did not allow us to change the base rates.

If we were starting from scratch with a day work situation, I am sure we would pay a bonus only after an improvement over past performance as Mr. Fenn suggests.

With reference to the criticism of the 75-25 split arrangement, we again agree with Mr. Fenn that employees are entitled to the full savings if the standard is based on 100 per cent efficiency and that efficiency is determined by soundly pace rated time studies. In fact that is our policy on such standards.

However, where past performance is the basis for the standard we have a different situation. It would be a rare group whose pre-standard performance was even near 100 per cent efficiency. Furthermore, performance can also usually be improved by better planning on the part of management and supervision. Thus, we and our employees feel that the 75-25 sharing arrangement is more than fair on all group plans not based on time study.

For further information we would like to make reference to the section on Wage Plans in the *Production Handbook* by Al-

ford, Bangs and Hageman.

Mr. Fenn's comments on the economics and psychology of incentives plans are well taken. We do not feel, however, that incentives must be tight to be sound. We are interested in high quality, high productivity and low manufacturing costs, and we are not afraid to pay employees generously to gain that end.

This has been our policy for many years with very happy results all around.

MR. TETZLAFF (Reply to Mr. Vogan): In reply to Mr. Vogan's comments on specification deviations and the desirability of higher silicon and manganese, we must say that we produce castings only for sale to several hundred customers. They and not we set up the specifications, although we try to fit their orders into standard specifications set up to satisfy the majority of orders.

It must be remembered, too, that our castings must often be repair welded or be welded into units with rolled steels and forgings. High silicon and high manganese makes welding difficult. High silicon content also makes the castings more susceptible to cracks and hot tears. Manganese is expensive and often scarce so why run up the amount when the added benefits are slight, if any? For further information on steel specifications, we suggest a reference to the Steel Founders' Society of America.

In answer to the question, "What is fairly constant?," we can say that the normal deviation in melt cycle times is less than 5 min.

As to the scrap variation we can say that we are located in an excellent market area, and that our sources have been giving us

consistently good scrap with only small specification deviations.

With reference to the assigned point values, we will not take issue with Mr. Vogan since each person or group of persons will probably evaluate the importance of specification deviations differently.

Mr. Vogan also seems to take issue with our allowing a 25 per cent quality bonus for a maintenance of post efficiency and also with our 75-25 split. For these answers we would like to refer back to our answers to Mr. Fenn.

In answer to the questions on the power savings portion, we can only say that our experience has proven that power is a controllable item. We have never experienced a negative power bonus, but we would not penalize the group if such should occur.

MR. TETZLAFF (Author's Closure): Our paper was not offered to present a perfect and highly polished incentive plan. We merely wished to tell about a plan that has worked well for us. If we were to start over again, we could add many improvements and refinements. With the same thought we hope that in telling of our experience, others can develop plans with necessary modifications which will do a job for them. If this should happen, we will be happy to hear about the plans.

# INFLUENCE OF SOME RESIDUAL ELEMENTS AND THEIR NEUTRALIZATION IN MAGNESIUM-TREATED NODULAR CAST IRON

By H. Morrogh*

#### Introduction

The conditions for the production of nodular cast irons by the use of magnesium have been described in a comprehensive manner in the original patent by Millis, Gagnebin and Pilling.¹ Subsequent literature has amply confirmed the claims of these investigators and therefore no detailed description of the process and the product is necessary here. Nodular cast irons may be defined as gray cast irons having graphitic carbon entirely in the form of nodules in the as-cast state without the necessity for heat treatment in order to achieve that structure. The graphite nodules in nodular cast irons have a characteristic structure, referred to as "spherulitic," the genesis of which is the subject of much speculation.²

To obtain a nodular structure by the process under consideration the molten iron is treated with sufficient magnesium in appropriate form shortly before casting to give a residual magnesium content in excess of about 0.04 per cent in the solidified casting, the sulphur content being at the same time reduced to a value of the order of 0.015 per cent or lower. The molten iron must be inoculated simultaneously with, or subsequent to, the magnesium addition. Experience has shown that the process is applicable to a wide range of compositions with respect to carbon, silicon, manganese, sulphur, phosphorus and nickel, the only limitations being the ability of the iron to solidify gray, which in normal circumstances is controlled by the carbon and silicon contents and section size of the casting.

#### Trace Elements in Raw Materials

The influence of these principal constituent elements on the production and properties of nodular cast irons has been well described in the literature, but, unfortunately, the raw materials of the foundry industry (pig iron, cast iron scrap, steel scrap, etc.) are not always "simple" iron-carbon-silicon-manganese-phosphorus-sulphur alloys. Small amounts of a wide variety of other elements are frequently present. For the production of normal flake graphite gray cast irons these "trace" elements can usually be ignored.

In the case of nodular cast irons, however, these "trace" elements can have a profound effect on the fundamental object—to produce nodular graphite structures. Reference has been made by Millis, Gagnebin and Pilling¹ to certain elements which should be avoided or limited in amount, but subsequent literature has largely avoided this issue.

Experience with the process in Britain quickly indicated that certain pig irons and other raw materials should be avoided if nodular graphite structures were to be produced in all casting sections, or even produced at all. It has been found that the unsuitability of these raw materials is due to their content of "trace" elements which prevent magnesium from having the desired effect.

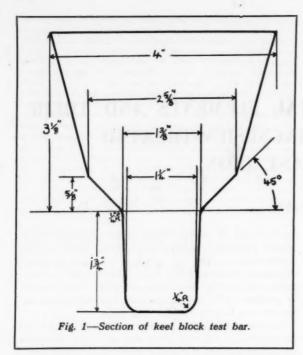
#### Limiting Subversive Elements

Millis, Gagnebin and Pilling¹ have pointed out that copper in large amounts inhibits the formation of nodular structures. They prefer to limit the copper content of nodular irons to not more than 2 per cent, but they point out that nickel will increase the tolerance for copper. They also indicate that certain other comparatively rare elements not usually found in cast iron should be avoided or limited to traces because they interfere with the production of nodular graphite. The elements tin, lead, antimony, bismuth, arsenic, selenium and tellurium were listed as subversive elements in this conection. Tin was found to be particularly detrimental, and it was recommended that it should not be allowed to exceed 0.05 per cent.

Holdemann and Stearns⁶ treated cast iron with various magnesium alloys containing subversive elements. An alloy of 20 per cent magnesium-80 per cent antimony gave a coarse flake graphite structure, but the magnesium content of the treated metal was only 0.012 per cent. An alloy of equal parts of magnesium and bismuth gave a flake graphite structure even though the residual magnesium content was 0.04 per cent. These investigators also obtained flake graphite in an iron treated with a magnesium alloy containing aluminum, copper, iron, lead, tin and zinc. Holdemann and Stearns did not give the subversive element content of any of their irons.

Morrogh⁵ illustrated a wide range of flake graph-

The British Cast Iron Research Association, Alvechurch, Birmingham, England.



ite structures in magnesium-treated irons. These structures we now know to have been due to the presence of titanium in the pig irons used, although this was not realized at the time. (This effect of titanium will be referred to later in this paper.) For a given magnesium content in these irons the amount of flake graphite was found to increase with increasing section size, while in a given section size the amount of flake graphite decreased with increasing magnesium. It is the present author's experience that if subversive elements are entirely absent,* and if the residual magnesium content is not less than 0.04 per cent, nodular graphite only will be obtained regardless of the section size.

Difficulties of this type have obviously been encountered by other investigators. For instance, Vennerholm, Bogart and Melmoth⁷ obtained flake forms of graphite particularly in heavy sections, and they were unable to minimize this occurence. No details were given of the trace subversive elements in their materials, but approximately 0.4 per cent copper was present. The structures obtained indicate that some subversive elements were present.

Similarly, Kraft and Flinn⁸ found appreciable amounts of flake graphite, increasing with increasing section size in irons of "marginal chemistry." These irons also contained about 1 per cent copper. It will be shown later that copper increases the sensitivity of magnesium-treated nodular irons to subversive ele-

nodular graphite and flake graphite in magnesiumtreated irons, implying that such irons have been treated with insufficient magnesium, although the one iron of which the analysis was given contained 0.04 per cent magnesium.

Again it would appear that we have flake structures produced by traces of subversive elements although no reference to these is made. Diagrams have been given in the literature showing the occurrence of flake graphite in heavy sections10 or indicating the desirability of higher residual magnesium contents in large sections. 10,11 This is another indication of the presence of harmful trace elements. In the absence of appreciable amounts of these elements completely nodular structures can be produced even in very heavy

Morrogh⁵ showed that aluminum can prevent the production of spherulitic graphite nodules by mag-

From this brief survey it would seem that the problem of subversive elements in cast iron is common to many countries where the production of nodular cast iron has been studied.

In Britain many pig irons and refined irons have been found to contain subversive elements and the difficulty has been overcome in the first instance by the use of melting stock low in these elements or which was proved by trial to be satisfactory. This procedure may be difficult to adhere to in view of the prevalent shortage of raw materials, and it is natural that an attempt should be made to find a method of neutralizing the harmful effects of these subversive elements to permit a wider choice of raw materials. It is the purpose of this paper to describe in more detail the effects of these elements and to indicate a method by which they can be successfully neutralized.

#### Need Improved Analytical Techniques

In the writer's experience few, if any, pig irons are entirely free from these subversive elements. As analytical procedures become more sensitive so does the widespread occurrence of these elements become more apparent. At this stage it is improbable that all the subversive elements have been detected. Furthermore, the amount of these elements which are relevant in this context are frequently so small that they are beyond the limits of the established analytical techniques. Improvements in techniques may well modify our views on the amounts which are harmful. This paper is only concerned with those elements which the present writer believes to be the most important. Much further work remains to be done, particularly on the quantitative aspect of this problem.

It has been found that small amounts of titanium, lead, bismuth and antimony have a pronounced subversive effect; tin in amounts up to 0.05 per cent, and arsenic in amounts up to 0.1 per cent, can be tolerated without harmful effect on graphite formation; aluminum can also have a harmful effect. The harmful effect of copper has not been entirely confirmed but has been found to be closely related to the subversive element content of the iron. A number of examples will be given to illustrate the effects of these elements.

In addition it has been found that a small amount of cerium added prior to, simultaneously with, or

ments. Wittmoser9 also shows mixed structures of

^{*} The expression "entirely absent" needs some qualification in this context. It is doubtful whether some subversive elements are ever entirely absent in the sense of not being present at all. The expression must be taken as meaning that the elements concerned are not detected by normal analytical procedures. It may be that some elements have a harmful influence even in amounts smaller than can be detected by these techniques.

subsequent to the magnesium addition is capable of neutralizing these subversive elements, either singly or in any combination. This effect will also be illustrated by a number of examples. This use of cerium enables pig irons to be used regardless of their subversive element content.

It should be pointed out that subversive elements may not only originate from the pig iron, steel scrap, etc., but also from furnace and ladle refractories, fuel, and ladle additions. Finally, this subject, while being of profound practical importance, is also of considerable theoretical interest—any theory to explain the effects of magnesium in producing nodular structures must also lead to an explanation of why some elements inhibit this effect and why cerium is capable of restoring it.

Except where otherwise stated, all the mechanical properties reported have been obtained on "keel" block test bars (Y blocks) cast in green sand molds. The keel block section is given in Fig. 1. Each block was 8½ in. long.

#### Influence of Lead

Table I gives the chemical analyses of a number of irons treated with various amounts of lead.

TABLE 1—CHEMICAL ANALYSES OF IRONS TREATED
WITH VARIOUS AMOUNTS OF LEAD

Iron	% Pb.	Composition, Per Cent								
No.	Added	T. C.	Si	Mn	S	P	Ni	Mg	Ce	Pb
1	0.00	3.44	1.96	< 0.02	0.018	0.026	0.74	0.084	_	0.004
2	0.01	do	2.03	20.02	0.012	do	0.70	0.059	_	0.005
3	0.03	do	2.04	≥0.02	0.014	do	0.72	0.059	-	0.009
4	0.05	do	1.98	≥0.02	0.013	do	0.81	0.054	-	0.013
5	0.075	3.46	2.11	≥0.02	0.014	do	0.66	0.055		0.011
6	0.10	do	2.11	≥0.02	0.012	do	0.68	0.052	-	0.011
7	0.05	3.33	2.09	< 0.02	0.014	do	0.81	0.061	0.021	0.014
8	0	3.27	2.35	0.50	0.011	0.059	1.67	0.083	_	_
9	0.05	do	2.44	0.47	0.010	do	1.58	0.113	-	0.012
10	0.05	do	2.35	0.50	0.010	do	1.66	0.079	0.017	0.013

Irons Nos. 1-6 were prepared from Swedish charcoal iron and illustrate the effect of increasing additions of lead on magnesium-treated irons. The Swedish charcoal iron was used because of its generally low content of residual elements. Iron No. 7 was also prepared from Swedish charcoal iron. Lead was added to this iron and it was subsequently treated with cerium misch metal as well as magnesium. Irons Nos. 8, 9 and 10 were prepared from a refined iron with an initial sulphur content of 0.130 per cent. Iron No. 8 was treated with magnesium, iron No. 9 with lead and magnesium, and iron No. 10 with lead, magnesium and cerium misch metal.

In each case the lead was added to the metal in the ladle prior to any other addition. The magnesium was added to the metal in the ladle in the form of a nickel-magnesium alloy when the temperature reached 1380 C as indicated by an immersion pyrometer. The cerium misch metal was added after the magnesium addition. The final addition was 0.5 per cent silicon as 80 per cent ferro-silicon in each case. Each addition was stirred into the molten metal by a refractory plunger.

The mechanical properties of this series are given in Table 2.

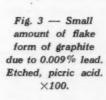
TABLE 2—EFFECT OF ADDITIONS (LEAD) ON MECHANICAL PROPERTIES

Iron No.	Màximum Tensile Stress, psi	Elongation,	Brinell Hardness No.
1	67,200	19	169
2	69,850	17	174
3	61,620	6	172
4	47,450	_	203
5	54,470	_	195
6	53,800	_	198
7	64,720	20	152
8	102,790	4	285
9	57,790	_	257
10	103,020	4	300

The microstructure of iron No. 1 is shown in Fig. 2 and is seen to consist of graphite nodules in a matrix of ferrite with a small amount of pearlite. Iron No. 2 had a structure very similar to No. 1, but iron No. 3 had a small amount of flake form of graphite as is shown in Fig. 3. The occurrence of this flake graphite is accompanied by a marked drop in elongation. Irons Nos. 4, 5 and 6 had similar microstructures, showing a form of flake graphite in a matrix of pearlite with a little ferrite and a few scattered nodules. These irons had no measurable elongation.

The predominantly ferritic structure of nodular irons prepared from this Swedish charcoal iron is attributable partly to the very low manganese content (less than 0.02 per cent) and partly to the very low content of pearlite stabilizing elements such as tin, arsenic, boron, bismuth, etc. The flake graphite structure of iron No. 4 is illustrated in Fig. 4. This structure is typical also of irons 5 and 6.

Fig. 2 — Nodular iron prepared from Swedish charcoal iron with no special addition other than magnesium and inoculant. Etched, picric acid. ×100.



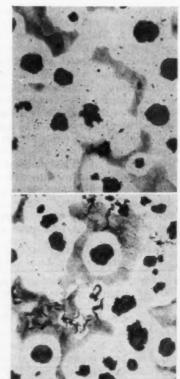




Fig. 4—Flake form of graphite in iron No. 4 with 0.013% lead. Etched, picric acid. ×100.

It should be noted that this flake graphite is not the normal variety, and this is reflected in the tensile strength values of irons 4, 5 and 6 which are somewhat higher than would be expected in irons having normal flake graphite structures with compositions as quoted above. The flake graphite shown in Fig. 4 is similar in many respects to the compacted flake graphite which occurs in normal irons having 0.02-0.04 per cent magnesium and which has been described in detail on previous occasions. 3.4.5 The increase in the amount of pearlite which accompanied the formation of the flake form of graphite in iron 4 is not easily explained.

It is not obvious, for instance, whether this is a direct effect of lead on the stability of pearlite carbide or whether it is a consequence of the change to flake graphite. It is perhaps significant that the compacted flake graphite formed in normal irons having 0.02-0.04 per cent magnesium may also be accompanied by pearlite, whereas ferrite occurs with higher magnesium contents when the structure is nodular.⁵

In Tables 1 and 2 iron No. 7 can be compared with iron No. 4. The lead additions are the same in each case and the residual lead contents are very similar. Iron No. 7 was treated with cerium misch metal after the magnesium addition, and it can be seen from the mechanical properties shown in Table 2 that the cerium addition has effectively neutralized the harmful effect of the lead. This is particularly noticeable in the elongation value. The microstructure of iron No. 7 was exactly similar to that of iron No. 1 illustrated in Fig. 2.

Iron No. 8 had a structure of nodular graphite in a matrix of pearlite together with a very small amount of an acicular flake form of graphite similar to that illustrated in Fig. 28. (This acicular graphite will be referred to later, but it occurred in this iron because it contained 0.04 per cent titanium.) The

TABLE 3—CHEMICAL ANALYSES OF IRONS TREATED WITH VARIOUS AMOUNTS OF BISMUTH

No.	m:	Composition, Per Cent								
	% Bi Added	T. C.	Si	Mn	S	P	Ni	Mg	Ce	Bi
11	0.00	3,44	2.07	< 0.02	0.023	0.024	0.65	0.049	-	0.0005
12	0.01		2.08	20.02	0.012		0.71	0.066	-	0.001
13	0.03		2.16	≥0.02	0.012		0.72	0.072	_	0.003
14	0.05		2.27	≥0.02	0.011		0.70	0.054	-	0.005
15	0.075	3.40	2.11	20.02	0.016		0.66	0.053	****	0.012
16	0.10		2.10	≥0.02	0.015		0.66	0.056	_	0.017
17	0.05	3.39	2.26	< 0.02	0.014		0.78	0.096	0.021	0.006

amount of this form of graphite was insufficient to have any influence on the tensile strength of the iron.

Iron No. 9, containing lead, had flake graphite, as shown in Fig. 5, in a matrix of pearlite. The occurrence of this flake graphite gave a marked drop in tensile strength. Iron No. 10, which also contained lead, but which had been treated with cerium as well as magnesium, had nodular graphite in a matrix of pearlite with no flake graphite, as shown in Fig. 6. The cerium addition has again neutralized the lead and completely restored the mechanical properties and microstructure of the material.

The few results presented so far clearly demonstrate the harmful effect of lead in nodular irons and the ability of cerium to neutralize the lead. In this series of experiments 0.009 per cent lead began to have a subversive effect and 0.011 per cent lead completely replaced nodular graphite with flake graphite.

It should be noted, however, that these figures are not necessarily applicable to all castings since the amount of lead which has a harmful influence appears to depend on the cooling rate of the section. With rapid cooling larger amounts of lead can be tolerated than with slow cooling. Thus a given lead content may only give a small amount of flake graphite in a relatively small section, whereas a completely flake graphite structure would be obtained in a relatively large section.

#### Influence of Bismuth

Table 3 gives the analysis of a series of irons prepared to illustrate the influence of bismuth.

These irons were again prepared from Swedish charcoal iron, the general procedure being the same

Fig. 5—Flake form of graphite in iron No. 9 containing 0.012% lead. Etched, picric acid. ×100.

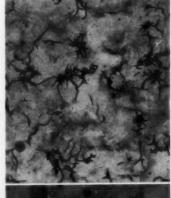
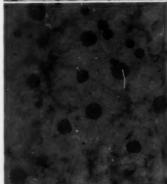


Fig. 6—Iron No. 10
with 0.013% lead
and 0.017% cerium.
Graphite nodules in
a matrix of pearlite.
Etched, picric acid.
×100.



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as that for the lead series. The bismuth was added before the magnesium addition. The mechanical properties for this series are given in Table 4.

Table 4—Effect of Additions (Bismuth) on Mechanical Properties

	MECHANICAL	PROPERTIES	
Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.
11	69,850	17	174
12	61,840	20	162
13	63,600	10	155
14	43,720		187
15	37,140	-	191
16_	34,050	-	196
17	73,020	16	173

Irons Nos. 11 and 12 had structures very similar to that already illustrated in Fig. 2. Iron No. 13 had traces of a flake form of graphite as shown in Fig. 7, and irons Nos. 14, 15 and 16 had a flake form of graphite in a matrix of pearlite with a little ferrite and occasional nodules as shown in Fig. 8. Iron No. 17, treated with cerium and containing 0.006 per cent bismuth, had a good nodular structure in a matrix of ferrite with a small amount of pearlite, i.e., a structure very similar to that illustrated in Fig. 2.

These results indicate that bismuth in amounts of the order of 0.003 per cent begin to have a harmful effect, and 0.006 per cent bismuth can completely inhibit the nodular structure. The mechanical properties and microstructure of iron No. 17 clearly illustrate the ability of cerium to neutralize the influence of bismuth.

As is the case with lead, experience suggests that the amount of bismuth which can be tolerated depends upon the cooling rate of the casting—larger amounts of bismuth can be tolerated in rapidly cooled sections than in slowly cooled sections.

#### Influence of Antimony

Antimony has a harmful effect similar to that of lead and bismuth, as is shown by the results given in Tables 5 and 6 giving chemical analyses and mechanical properties, respectively.

Irons Nos. 18-21 are a series with increasing antimony additions. The first addition of antimony, giving a residual antimony content of only 0.004 per cent, had a profound effect on the elongation value. The structure of iron No. 18 with no antimony addition was almost identical with that illustrated in Fig. 2, whereas iron No. 19 had a small amount of

TABLE 5—CHEMICAL ANALYSES OF IRONS TREATED WITH VARIOUS AMOUNTS OF ANTIMONY

		Composition, Per Cent										
	% Sb Added	T. C.	Si	Mn	S	P	Ni	Mg	Ce	Sb		
18	0.00	3,42	2.06	< 0.02	0.014	0.024	0.72	0.046				
19	0.01		2.20	≥0.02	0.013		0,79	0.055		0.004		
20	0.03		2.16	≥0.02	0.013		0.76	0.047		0.012		
21	0.05		2.19	≥0.02	0.015		0.74	0.056		0.026		
22	0.05	3.45	2.32	< 0.02	0.012	0.023	0.84	0.075	0.015	0.024		
23	0.05		2.27	≥0.02	0.011		0.79	0.081	0.028	0.022		
24	0.05		2.27	≥0.02	0.013		0.82	0.077	0.049	0.015		

Fig. 7—Iron No. 13 with 0.003% bismuth. Etched, picric acid. ×100.

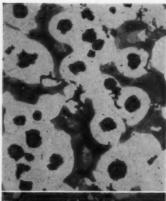


Fig. 8—Iron No. 14 with 0.005% bismuth. Etched, picric acid. ×100.

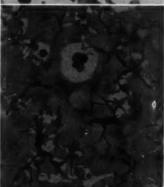


TABLE 6—EFFECT OF ADDITIONS (ANTIMONY) ON MECHANICAL PROPERTIES

Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.
18	70,560	17	178
19	66,340	2	212
20	81,720	_	232
21	81,496	_	231
22	85,324	. 7	225
23	76,200	12	207
24	73,040	15	167

flake graphite together with spherulitic nodules in a matrix containing considerably more pearlite than iron No. 18, as is shown in Fig. 9.

Irons Nos. 20 and 21 had similar structures with rather more flake graphite than iron No. 19 but with an essentially pearlitic matrix with only a small amount of ferrite. The structure of iron No. 20 is shown in Fig. 10.

The first addition of cerium almost completely restored the nodular structure, only a trace of flake graphite being present, but it only restored a little of the original ferrite, as is shown in Fig. 11. With 0.028 per cent cerium in iron No. 23 the nodular structure was completely restored, although there was still slightly more pearlite than in the original structure. This is reflected in the hardness of this iron. The structure of this iron is illustrated in Fig. 12. Iron No. 24 with 0.049 per cent cerium had a structure exactly similar to that of iron No. 18 (Fig. 2).

These results clearly demonstrate that antimony contents as low as 0.004 per cent can have a harm-

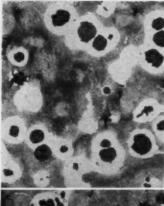


Fig. 9—Iron No. 19 with 0.004% antimony. Etched, picric acid. ×100.

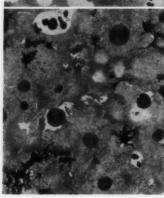


Fig. 10—Iron No. 20 with 0.012% antimony. Etched, picric acid. ×100.

ful effect on the production of spherulitic nodular structures by magnesium. With increasing amounts of antimony the tendency for pearlite formation is increased. Cerium is clearly capable of neutralizing the harmful effect of antimony, but the extent of neutralization depends upon the amount of residual cerium.

Irons Nos. 1-7 (with lead additions), irons 11-17 (with bismuth additions), and irons Nos. 18-24 (with antimony additions) were all prepared from the Swedish charcoal iron referred to previously. This iron was chosen partly because of its known low content of residual elements and partly because of its low manganese content which facilitates the formation of ferritic structures. This makes the material particularly useful for observing the effects of the various elements being studied upon the tendency to pearlite formation.

It has been found that the elements tin and arsenic have a profound effect in encouraging pearlite formation, but in order to place the results obtained with these elements in the correct perspective the

TABLE 7—CHEMICAL ANALYSES OF IRONS WITH VARIOUS AMOUNTS OF MANGANESE

Iron	Composition, Per Cent										
No.	T.C.	Si	Mn	S	P	Ni	Mg				
25	3.34	2.13	>0.02	0.012	0.029	0.70	0.048				
26		2.07	0.12	0.011		0.60	0.049				
27		2.11	0.25	0.010		0.65	0.045				
28		2.15	0.44	0.011		0.66	0.044				
29	3.48	2.18	0.72	0.010	0.021 .	0.84	0.052				
30		2.19	0.95	0.009		0.86	0.068				

influence of manganese on the matrix structure of nodular irons prepared from the Swedish iron will be demonstrated first.

Table 7 gives the analyses of a series of nodular irons prepared from the Swedish iron with increasing manganese additions, and Table 8 gives the mechanical properties and microstructures of this series.

It will be seen from these results that even with a manganese content of 0.95 per cent an appreciable amount of ferrite was present.

#### Influence of Tin

Tables 9 and 10 give the chemical analyses and mechanical properties, respectively, of a series of nodular irons with various tin additions. These irons were all prepared from Swedish charcoal iron.

In this series there is a progressive increase in the amount of pearlite with increasing additions of tin. This is accompanied by a progressive increase in tensile strength, elongation and hardness. In all cases the nodular graphite structure was perfect with no

TABLE 8—EFFECT OF MANGANESE ON MECHANICAL PROPERTIES

Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.	Microstructure
25	62,700	18	164	Nodular+90% ferrite
26	65,230	15	166	Nodular +90% ferrite
27	72,600	13	173	Nodular + 85% ferrite
28	74,470	5	199	Nodular + 40% ferrite
29	84,920	2	235	Nodular +25% ferrite
30	94,100	2	239	Nodular+15% ferrite

Fig. 11 — Iron No.
22 with 0.022%
antimony and
0.015% cerium.
Etched, picric acid.
×100.

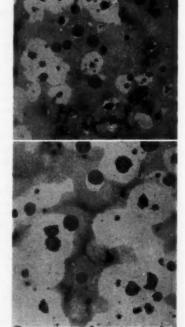


Fig. 12—Iron No. 23 with 0.022% antimony and 0.028% cerium. Etched, picric acid. ×100.

TABLE 9—CHEMICAL ANALYSES OF IRONS WITH VARIOUS AMOUNTS OF TIN

	Composition, Per Cent								
Iron No.	% Sn Added	T.C.	Si	Mn	5	P	Ni	Mg	Sn
31	0.00	3.49	2.08	< 0.02	0.010	0.023	0.66	0.047	
32	0.01		2.09		0.011		0.73	0.048	0.012
33	0.03		2.05		0.010		0.69	0.049	0.014
34	0.05		2.02		0.009	0.022	0.66	0.052	0.016
35	0.075	3.34	2.11		0.011		0.66	0.044	0.022
36	0.125		2.19		0.008		0.71	0.046	0.041

TABLE 10—Effect of Tin on Mechanical Properties

Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.	Microstructure
31	62,520	13	156	Nodular+90% ferrite
32	65,000	15	161	Nodular +90% ferrite
33	86,520	10	208	Nodular+30% ferrite
34	91,800	4 .	239	Nodular + 5% ferrite
35	99,496	4	249	Nodular+ 2% ferrite
36	101,000	3	253	Nodular. All pearlite

flake graphite. Irons Nos. 31 and 32 had structures very similar to that shown in Fig. 2, but there was a sharp increase in the amount of pearlite in iron No. 33 with 0.016 per cent tin. The structure of this iron is shown in Fig. 13. With 0.041 per cent tin in iron No. 36 the metallic matrix became completely

pearlitic, as is shown in Fig. 14.

It would seem from these results that, in the absence of appreciable amounts of subversive elements, up to 0.04 per cent of tin can be tolerated without harmful effect on the formation of nodular graphite. Tin is shown, however, to have a powerful influence on the formation of pearlite and comparing the results given in Tables 9 and 10 with those in Tables 7 and 8, it would appear that 0.04 per cent tin has a more powerful effect than 0.9 per cent of manganese. Experiments have been carried out with cerium additions to tin-bearing irons, but these have shown that cerium is incapable of neutralizing the pearlite stabilizing effect of tin.

#### Influence of Arsenic

The influence of arsenic is illustrated by the results given in Tables 11 and 12.

Arsenic is seen to have an effect very similar to that of tin, but at least twice as much arsenic is required to achieve the same degree of pearlite stabilization as that of a given amount of tin. It is also

TABLE 11—CHEMICAL ANALYSES OF IRONS WITH VARIOUS AMOUNTS OF ARSENIC

		Composition, Per Cent									
	% As Added	T. C.	Si	Mn	5	P	Ni	Mg	Ce	As	
37	0.00	3.43	2.10	< 0.02	0.011	0.023	0.74	0.047			
38	0.01		1.99	20.02	0.010		0.79	0.054		0.010	
39	0.03		1.96	≥0.02	0.009		0.74	0.052		0.029	
40	0.05	3.51	2.01	>0.02	0.012	0.023	0.62	0.045		0.025	
41	0.075		2.08	20.02	0.013		0.60	0.045		0.046	
42	0.10		2.11	20.02	0.010		0.63	0.044		0.080	
43	0.125		2.05	≥0.02	0.011		0.62	0.049		0.094	
44	0.125	3.41	2.10	< 0.02	0.016	0.025	0.74	0.073	0.017	0.078	
45	0.125		2.03	≥0.02	0.011		0.68	0.056	0.028	0.086	
46	0.125		1.95	≥0.02	0.009		0.69	0.055	0.037	0.085	

Fig. 13—Iron No. 33 with 0.016% tin. Etched, picric acid. ×100,

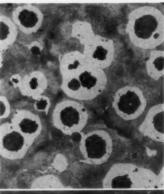


Fig. 14—Iron No. 36 with 0.041% tin. Etched, picric acid. ×100.

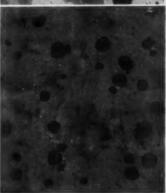


TABLE 12—EFFECT OF ARSENIC ON MECHANICAL PROPERTIES

Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.	Microstructure
37	68,100	13	177	Nodular+90% ferrite
38	68,100	14	179	Nodular+90% ferrite
39	66,770	4	183	Nodular+40% ferrite
40	79,970	12	185	Nodular+50% ferrite
41	93,370	3	266	Nodular+ 5% ferrite
42	89,600	2	262	Nodular+ 1% ferrite
43	95,000	2	274	Nodular+1/2% ferrite
44	93,590	9	239	Nodular+20% ferrite
45	100,390	9	242	Nodular + 20% ferrite
46	101,000	9	229	Nodular+15% ferrite

clear that at least 0.09 per cent arsenic can be tolerated without harmful effect on the formation of nodular graphite.

Whether arsenic or tin should be considered as subversive elements depends on the type of nodular iron required. If the aim is to produce a high strength iron with a pearlitic matrix these elements may be ignored, but if the aim is to produce a relatively soft iron of good ductility, either in the ascast condition or after heat treatment, then the amounts of tin and arsenic must be kept to a minimum.

The results given in Tables 11 and 12 for irons 44-46, which should be compared with irons 42 and 43, indicate that cerium may slightly offset the pearlite stabilizing effect of arsenic. The influence of cerium is not pronounced, however, and is of doubt-

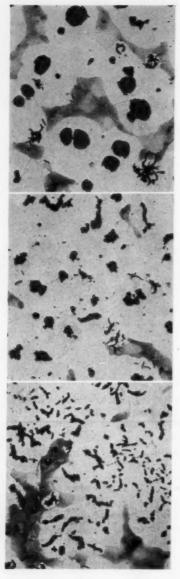


Fig. 15—Iron No. 52 with 0.13% aluminum. Etched, picric acid. ×100.

Fig. 16-Iron No.

53 with 0.34% alu-

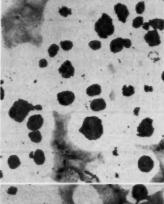
minum. Etched, pic-

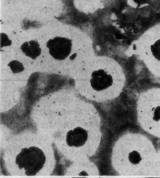
ric acid. ×100.

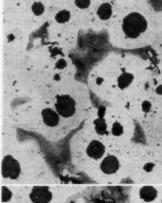


Fig. 19—Iron No. 59 with 0.02% titanium. Etched, pic-

ric acid. ×100.







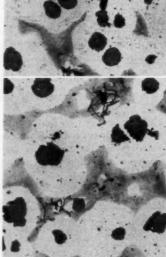


Fig. 17—Iron No. 54 with 0.44% alu-

minum. Etched, pic-

ric acid. ×100.

Fig. 20—Iron No. 60 with 0.03% titanium. Etched, picric acid. ×100.

ful practical value if large amounts of arsenic are encountered.

#### Influence of Aluminum

Aluminum could inhibit the formation of spherulitic nodules and cause the formation of flake graphite structures in magnesium-treated irons. It was, however, suggested in this earlier work that aluminum caused the retention of sulphur, which in turn caused the formation of flake graphite. It will be shown, however, that aluminum can have a harmful effect even when the sulphur contents are normal for nodular irons.

The results given in Tables 13 and 14 illustrate the influence of aluminum and the ability of cerium to neutralize its harmful influence.

Irons Nos. 47 and 48 had structures very similar to that already illustrated in Fig. 2. Minute traces of a flake form of graphite were present in irons Nos. 49, 50 and 51 containing 0.07, 0.08 and 0.10 per cent

Fig. 21—Iron No. 61 with 0.04% titanium. Etched, picric acid. ×100.

aluminum, respectively, but this did not appear to influence the mechanical properties. More of this flake graphite occurred in iron No. 52 containing 0.13 per cent aluminum, and this was accompanied by a sharp drop in the elongation value. Figure 15 shows

TABLE 13—CHEMICAL ANALYSES OF IRONS WITH VARIOUS AMOUNTS OF ALUMINUM

	~ 41	Composition, Per Cent										
Iron No.	% Al Added	T. C.	Si	Mn	S	P	Ni	Mg	Ce	Al		
47	0.00	3.48	2.07	< 0.02	0.008	0.019	0.74	0.068				
48	0.052		2.10	20.02	0.011		0.73	0.070		0.04		
49	0.103	3.36	2.00	≥0.02	0.015	0.029	0.89	0.076		0.07		
50	0.130		2.05	≥0.02	0.010		0.86	0.076		0.08		
51	0.160		2.01	≥0.02	0.011		0.89	0.079		0.10		
52	0.22		2.01	20.02	0.013		0.90	0.078		0.13		
53	0.33	3.26	2.09	20.02	0.009	0.024	0.72	0.060		0.34		
54	0.44		2.15	20.02	0.007		0.71	0.064		0.44		
55	0.55		2.09	≥0.02	0.008		0.73	0.057		0.49		
56	0.55	3,34	2.00	< 0.02	0.009	0.023	0.61	0.072	0.008	0.54		
57	0.55		1.94	< 0.02	0.009	*	0.66	0.050	0.024	0.55		

TABLE 14—EFFECT OF ALUMINUM ON MECHANICAL PROPERTIES

Iron No.	Maximum Tensile Stress, psi	Elongation,	Brinell Hardness No.	Microstructure
47	66,790	20	162	Nodular+10% pearlite
48	68,540	18	174	Nodular+10% pearlite
49	69,848	17	170	Nodular+trace flake +10% pearlite
50	67,424	20	158	Nodular+trace flake +10% pearlite
51	67,648	18	160	Nodular+trace flake +10% pearlite
52	62,290	6	164	Nodular+little flake +10% pearlite
53	57,570	9	148	Flake+nodules+5% pearlite
54	56,000	8	148	Flake +5% pearlite
55	55,370	6	152	Flake+nodules+ 5% pearlite
56	61,840	19	151	Nodular+100% ferrite
57	63,148	19	152	Nodular+5% pearlite

the structure of this iron and the small amount of flake graphite can be clearly seen.

Iron No. 53 had a compacted flake form of graphite together with somewhat imperfectly formed nodules, as illustrated in Fig. 16. Iron No. 54 with 0.44 per cent aluminum had no spherulitic nodules and all the graphite was in the form of compacted flakes, as shown in Fig. 17. Iron No. 55 had a structure very similar to that of iron No. 53.

Irons Nos. 56 and 57 treated with cerium had graphite structures consisting entirely of spherulitic nodules in a completely ferritic matrix. It should be noted that iron No. 56 contained only 0.008 per cent cerium, and yet this was sufficient to neutralize the harmful effect arising from more than 0.5 per cent aluminum. More will be said later on the amount of cerium required to neutralize the subversive elements, but in passing it might be inferred

that of the total concentration of subversive element only a small fraction appears to contribute to the subversive effect.

#### Influence of Titanium

Titanium probably occurs more frequently in foundry pig irons in amounts capable of easy detection than any other subversive element. Its influence on microstructure is illustrated in Figs. 18, 19, 20 and 21, showing the general structures of irons Nos. 58, 59, 60 and 61, respectively, the chemical analyses and mechanical properties of which are given in Table 15.

Traces of a flake form of graphite can be seen in Fig. 19. Figure 20 shows still more flake graphite, and Fig. 21 a further increase in this constituent. In this series the amount of flake graphite was insufficient to have any marked effect on the mechanical properties examined.

The influence of titanium depends very considerably upon the magnesium content and the section size into which the metal is cast. This can be illustrated by considering the structures of the samples, the analyses of which are given in Table 16.

TABLE 16—CHEMICAL ANALYSES—TITANIUM AND MAGNESIUM

Iron No.	Composition, Per Cent												
	T.C.	Si	Mn	S	P	Ni	Mg	Ti					
62	3.91	2.16	0.27	0.022	0.05	0.31	0.043	0.08					
62	3.84	2.18				0.43	0.050	0.08					
64	3.95	2.18				0.59	0.075	0.08					
65	3.87	2.30				1.07	0.117	0.08					

Standard 1.2 in., 0.875 in. and 0.6 in. test bars were cast from each composition. Such test bars are unsuitable for studying the mechanical properties of nodular irons, but serve admirably to demonstrate the effect of section size. Figures 22, 23, 24 and 25 show the structures of the 1.2-in. bars of irons Nos. 62, 63, 64 and 65, respectively. It can be seen that the amount of flake graphite decreases in a uniform manner with increasing magnesium content.

Figure 26 shows the structure of the 0.6-in. bar cast from composition No. 64, and should be compared with Fig. 24. The decrease in amount of flake graphite with a decrease in section size is clearly seen. Figure 27 shows the structure of 1.6-in, bar from composition No. 65 with 0.117 per cent magnesium. Figure 27 should be compared with Fig. 25. The increase in amount of flake graphite with increasing section is again apparent.

The present author5 has suggested this structure de-

TABLE 15—CHEMICAL ANALYSES AND INFLUENCES OF TITANIUM ON PHYSICAL PROPERTIES

Iron			(	Composition	, Per Cen	1			Maximum T. Stress,	Elonga-	Brinell Hardness
No.	T.C.	Si	Mn	S	P	Ni	Mg	Ti	psi	tion, %	No.
58	3.60	2.06	< 0.02	0.012	0.024	0.76	0.056		68,800	19	170
59		2.11	< 0.02	0.012		0.81	0.062	0.02	71,000	12	179
60		2.12	< 0.02	0.013		0.76	0.068	0.03	70,750	15	178
61		2.07	< 0.02	0.010		0.75	0.060	0.04	75,000	15	185

pendence on magnesium content and section size to be a property of all magnesium-treated hypereutectic nodular irons, but it is now known to be a function of the subversive element content of the base iron used. With very low subversive element contents, nodular

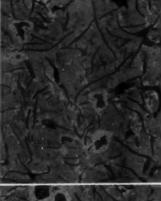


Fig. 22—Iron No. 62. 1.2 in. diameter bar with 0.043% magnesium. Etched, picric acid. ×100.

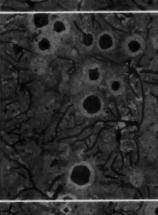


Fig. 23—Iron No. 63, 12 in. diameter bar with 0.050% magnesium. Etched, picric acid. ×100.

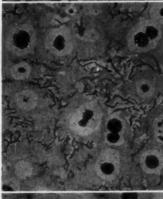


Fig. 24—Iron No. 64. 12 in. diameter bar with 0.075% magnesium. Etched, picric acid. ×100.

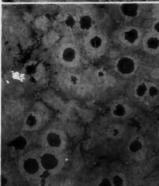
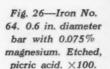


Fig. 25—Iron No. 65. 1.2 in. diameter bar with 0.117% magnesium. Etched, picric acid. ×100.



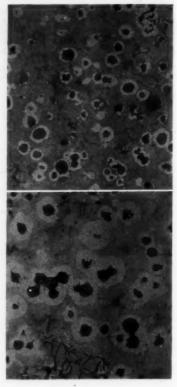


Fig. 27—Iron No. 65. 0.6 in. diameter bar with 0.117% magnesium. Etched, picric acid. ×100.

irons can be made in a wide range of section sizes provided the magnesium content is in excess of 0.04 per cent.

Cerium is capable of neutralizing the subversive effect of titanium, as is shown by the results given in

Irons 66-71 were prepared from British pig irons and irons 72 and 73 from an American pig iron. In each case the addition of a small amount of cerium has produced a marked improvement in the tensile strength due to the replacement of a mixed flake and nodular graphite structure by a completely nodular graphite structure.

#### Influence of Copper

The influence of copper in nodular cast irons depends very considerably upon the subversive element content of the materials. The effect of copper on the formation of nodular graphite is not completely understood and much further work remains to be done. Copper appears to make nodular cast irons more sensitive to the effect of subversive elements.

In the absence of substantial amounts of subversive elements, the present investigator has produced good nodular structures with copper contents of up to 3 per cent. With relatively high nickel contents, i.e., with 5-20 per cent nickel, the amount of copper which can be tolerated in the presence of subversive elements is increased. The harmful effects introduced by copper can be neutralized completely by the addition of small amounts of cerium.

Table 18 gives the analyses and mechanical properties of a series of nodular cast irons with increasing copper contents prepared from the Swedish iron.

TABLE 17-NEUTRALIZING EFFECT OF CERIUM ON TITANIUM CONTENT

Iron				Comp	oositon, Per	Cent				Maximum Tensile Stress,	Brinell Hardness
No.	T.C.	Si	Mn	S	P	Ni	Mg	Ce	Ti	psi	No.
66	3.71	2.45	0.69	0.010	0.053	0.86	0.055		0.07	64,000	239
67		2.40	0.70			1.15	0.095		0.07	78,900	260
68	3.83	2.34	0.66	0.021		0.78	0.051	0.007	0.08	98,600	263
69	3.58	2.80	0.70	0.011	0.054	1.15	0.056		0.09	68,500	262
70		2.75	0.68			1.46	0.090		0.09	75,600	260
71	3.71	2.94	0.66	0.010		0.85	0.056	0.016	0.08	90,400	241
72	2.88	2.60	0.25	0.009	0.040	0.75	0.065	_	0.15	58,000	168
73		2.54	0.25	0.013	0.040	0.75	0.050	0.039	0.16	80,000	205

TABLE 18-EFFECT OF INCREASING COPPER CONTENTS

			Comp	osition	n, Per	Cent			Max. Tensile Stress.	Brinell Hard-
Iron No.		Si	Мв	S	P	Ni	Mg	Cu	psi	No.
74	3.22	2.12	<b>~0.02</b>	0.012	0.025	0.80	0.10	-	63,800	152
75		2.01	≥0.02	0.015		0.79	0.065	0.27	82,900	213
76		1.95	20.02	0.016		0.76	0.062	0.60	110,480	262
77	3.33	1.96	20.02	0.014		0.83	0.078	2.35	107,500	278
78		2.06	≥0.02	0.013		0.83	0.071	3.13	116,500	306

Fig. 28—Iron No. 79 with 0.04% titanium. Etched, picric acid. ×100.



All of these irons had good nodular structures. Iron No. 74 without copper had a matrix of 95 per cent ferrite and 5 per cent pearlite. Iron No. 75 with 0.27 per cent copper had about 50 per cent ferrite and 50 per cent pearlite, and all the other irons of this series had completely pearlitic structures. In this series, 3 per cent copper was without harmful effect on nodular graphite formation.

Table 19 gives the analyses and mechanical properties of a series of irons with increasing copper contents, all prepared from a base iron containing titanium.

Iron No. 79 had an almost completely nodular graphite structure with traces of a flake form of graphite in a matrix of pearlite. This structure is shown in Fig. 28. The small amount of the flake form of graphite is caused by the titanium present in this iron. Iron No. 80, with 1.15 per cent copper, had a flake form of graphite in a matrix of pearlite with a few spherulitic

Fig. 29—Iron No. 80 with 0.04% titanium and 1.15% copper. Etched, picric acid. 100×.



TABLE 19-EFFECT OF COPPER WITH TITANIUM PRESENT

Iron				Compo	sition, Per	Cent				Maximum Tensile Stress, psi	Hardness
No.	T. C.	Si	Mn	S	P	Ni	Mg	Cu	Ti		
79	3.32	2.19	0.45	0.011	0.058	2.09	0.076	_	0.04	103,000	290
80		2.20	0.44	0.010		2.29	0.094	1.15		79,000	299
81		2.22	0.42	0.003		2.37	0.131	1.96		57,400	278
82		2.28	0.44	0.012		2.38	0.148	2.78		48,400	278

TABLE 20-EFFECT OF CERIUM WITH COPPER AND TITANIUM PRESENT

Iron												Brinell Hardness
No.	T.C.	Si	Mn	S	P	Ni	Mg	Cu	Се	Ti	Tensile Stress, psi	No.
83	3.26	2.07	0.43	0.026	0.060	2.03	0.061	_	0.025	0.03	102,800	295
84		2.24	0.43	0.023		2.20	0.057	1.20	0.024		97,800	321
85		2.18	0.42	0.025		2.19	0.069	1.89	0.025		105,000	337
86		2.20	0.42	0.026		2.27	0.085	2.65	0.027		106,000	331

nodules, as is shown in Fig. 29. Iron No. 81, with 1.96 per cent copper, had still less nodular graphite, and iron No. 82, with 2.78 per cent copper, had somewhat coarser flake graphite with only a few nodules. This series shows that in the presence of 0.04 per cent titanium even approximately 1 per cent copper is sufficient to seriously interfere with the formation of nodular graphite.

Table 20 gives the analyses and mechanical properties of a similar series to that of Table 19, but a small

addition of cerium was made to each iron.

All of these irons had good nodular graphite structures in matrices of pearlite. The small amount of cerium present completely neutralized the combined harmful effects of copper and titanium.

The results given so far demonstrate the danger of using copper-magnesium alloys for the introduction of magnesium into nodular cast iron melts, particularly in the presence of subversive elements such as titanium. This difficulty can be overcome by the use of cerium additions and, in particular, by the use of copper-magnesium-cerium alloys. Table 21 gives the

TABLE 21—EFFECT OF CU-MG AND CU-MG-CE Additions

V				Max. Tensile Stress,	Brinell Hard- ness					
No.	T.C.	Si	Mn	S	P	Cu	Mg	Ce	psi .	No.
87	3.38	1.97	0.61	0.009	0.055	0.99	0.042	_	49,300	245
88		2.04	0.59	0.005		1.51	0.071	-	46,700	239
89		2.05	0.62	0.015		1.37	0.070	0.027	111,200	293
90		2.01	0.59	0.012		1.46	0.096	0.042	103,900	288

analyses and mechanical properties of four irons prepared from an iron containing titanium. The first two of these irons were prepared using a copper-magnesium alloy containing 15 per cent magnesium; the other two irons were prepared using a copper-magnesium-cerium alloy containing 14.2 per cent magnesium and 2.2 per cent cerium.

These irons contained 0.07 per cent titanium. Irons No. 87 and 88 had flake graphite structures in a matrix of pearlite, whereas irons Nos. 89 and 90 had good nodular graphite structures in pearlite.

#### Amount of Cerium for Neutralization

Some interest, both theoretical and practical, attaches to the amount of cerium required to neutralize subversive elements. A final statement on this point can only be prepared after considerable work, but for the present it will suffice to say that very small amounts of cerium are capable of neutralizing considerable quantities of subversive elements. This is illustrated by the results given in Table 22.

Both of the irons contained lead and titanium in substantial amounts. Iron No. 91 was treated with magnesium, but not with cerium, and had a structure of flake graphite in a matrix of pearlite. Iron No. 92 was treated with a very small amount of cerium in addition to magnesium, and the residual content of less than 0.005 per cent cerium was sufficient to neutralize the combined subversive effects from 0.019 per cent lead and 0.08 per cent titanium. This iron had a completely nodular graphite structure in a matrix of pearlite.

#### Summary and Conclusions

A very brief account has been given of some of the work carried out by the author in the laboratories of the British Cast Iron Research Association on the subject of the influence of subversive elements on the production of nodular graphite structures by the use of magnesium. It has been shown that very small amounts of lead, antimony, bismuth and titanium inhibit the formation of nodular graphite. Aluminum has a similar effect. The influence of copper is complex and depends upon whether the iron contains subversive elements such as titanium, in which case even as little as 1 per cent copper can cause the formation of substantial amounts of flake graphite.

Arsenic in amounts up to 0.09 per cent and tin in amounts up to 0.04 per cent have been studied and found to be without harmful effect on the formation of nodular graphite. However, these two elements have a powerful influence in causing the formation of pearlite in irons which would otherwise contain sub-

stantial amounts of ferrite.

It has been shown that cerium is capable of neutralizing the harmful effects of titanium, lead, bismuth, antimony, aluminum and copper. The amount of cerium required appears to be very low.

It is not expected that all of the possible subversive elements have been dealt with, but it is thought that those mentioned are the most important from the

practical point of view.

It has been found that the cerium used for neutralizing the subversive elements may be added to the molten metal prior to, simultaneously with, or subsequent to the addition of the magnesium. The cerium may be added as misch metal or as some alloy containing misch metal or cerium. The present writer has found it convenient to use nickel-magnesium-cerium alloys containing 10-18 per cent magnesium and 0.5-3 per cent cerium. The use of such alloys permits the production of nodular cast irons irrespective of the subversive element content of the melting stock, and thus increases considerably the range of raw materials available for the process.

Subversive elements do not, however, always arise from the melting stock, but may be introduced into the melt from contaminated refractories. Lead, antimony and bismuth are quickly absorbed by hot refractories such as furnace and ladle linings, and are gradually transferred from the refractories to molten metal

TABLE 22—CERIUM NEUTRALIZING EFFECT ON SUBVERSIVE ELEMENTS

Iron	Composition, Per Cent									Maximum	Brinell Hardness	
No.	T.C.	Si	Mn	S	P	Ni	Mg	Ce	Pb	Ti	Stress, psi	
91	3.65	2.38	0.70	0.007	0.047	0.70	0.051	_	0.020	0.08	22,400	189
92		2.23	0.70	0.010		0.71	0.056	< 0.005	0.019	0.08	103.000	260

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on each successive contact. Thus a ladle lining contaminated with lead may continue to introduce lead into several successive taps of molten metal. The addition of cerium can completely neutralize the harmful effects arising from such refractories.

There is some desire to avoid the use of nickel as a means of adding magnesium to nodular iron melts on account of the high cost and present short supply of this element. Nevertheless, in the author's experience, nickel-base alloys are technically superior to all other alloys which have been tried.

Copper-base alloys have been suggested and used for the production of nodular cast irons. The results presented in this paper have demonstrated the danger of introducing copper into nodular cast irons, but by using copper-magnesium-cerium alloys this danger would be avoided. Unfortunately, the yield of magnesium from copper-magnesium alloys having 10-18 per cent magnesium is not as good as that from the corresponding nickel-base alloys.

It might, however, be practicable to use copper-base alloys containing 25-40 per cent magnesium and 1-6 per cent cerium. With these alloys the yield calculated on the basis of the amount of magnesium added is very low, but it appears that the yield calculated on the total weight of addition used can be as good as that when using nickel-base alloys. On account of the relatively high cost of cerium one would endeavor to use as little of this element as possible, but the results available so far indicate that only very small amounts are necessary.

#### Acknowledgments

The author wishes to thank the Council and Director of the British Cast Iron Research Association for permission to publish this paper.

The use of cerium (misch metal) in cast irons to ensure the production of nodular (spheroidal) graphite structures is covered by a series of patents in Great Britain and a number of industrial countries granted to or applied for by the British Cast Iron Research Association, Alvechurch, Birmingham, England.

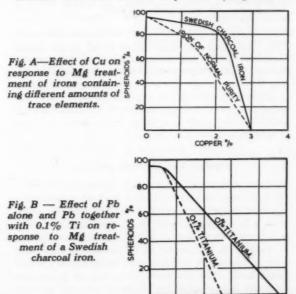
#### References

- 1. K. D. Millis, A. P. Gagnebin and N. B. Pilling, U. S.
- Patent No. 2,485,760, Oct. 25th, 1949.

  2. F. H. Buttner, H. F. Taylor and J. Wulff, American FOUNDRYMAN, Oct. 1951, pp. 49-50.
- 3. K. D. Millis, A. P. Gagnebin and N. B. Pilling, U. S. Patent No. 2,485,761, Oct. 25th, 1949.
- 4. W. W. Braidwood and A. D. Busby, Fonderie Belge, 1949, No. 20, May-August, pp. 21-30.
- 5. H. Morrogh, Journal of Research and Development, The
- British Cast Iron Research Association, Feb. 1950, pp. 251-298. 6. G. E. Holdemann and J. C. H. Stearns, AMERICAN FOUNDRY-
- MAN, Aug. 1949, pp. 36-41. 7. G. Vennerholm, H. Bogart and R. Melmoth, TRANSACTIONS,
- American Foundrymen's Society, vol. 58, pp. 174-184 (1950). 8. R. W. Kraft and R. A. Flinn, A.S.M. Preprint No. 33, 1951.
  - 9. A. Wittmoser, Giesserei, Nov. 1951, pp. 572-577.
- 10. E. Piwowarsky, K. Krämer, W. Patterson and M. Krichel, Die Neue Giesserei, vol. 3, June 1950, pp. 91-103.
- 11. J. E. Rehder, AMERICAN FOUNDRYMAN, Sept. 1949, pp. 33-37.

#### DISCUSSION

- Chairman: A. P. GAGNEBIN, International Nickel Co., New
- Co-Chairman: R. SCHNEIDEWIND, University of Michigan, Ann Arbor.
- Recorder: G. A. TIMMONS, Climax Molybdenum Co., Detroit.
- A. B. Everest: 1 This paper is of great interest particularly to industries in Europe where their pig irons and scrap contains greater quantities of subversive or interfering elements than are found in the raw materials in the United States.
- Since the cerium process and the magnesium-nickel process are both being used for the production of spheroidal graphite irons, it is a challenge to try combinations of both.
- W. STEVEN AND R. M. LAMB (Written discussion presented by Dr. A. B. Everest): 1 We were privileged to see a copy of Mr. Morrogh's excellent paper before this meeting, and welcome the opportunity thus provided, of contributing to its discussion.
- As indicated in the introduction to the paper, the effect of residual subversive elements on the development of spheroidal graphite was clearly indicated in the original Inco-Mond patents covering the magnesium process but since then, little has been published on this undoubtedly important subject. Considerable further work has however, been completed at the Bayonne Laboratories of 'The International Nickel Co. Ltd., and at the Birmingham Laboratories of The Mond Nickel Co. Ltd., and it is pleasing to see how closely our results agree with those now published by Mr. Morrogh.
- We agree with Mr. Morrogh's list of those elements which interfere with the development of spheroidal graphite, but would add indium and thallium. Both of these elements exert appreciable effects in small amounts, but fortunately neither occurs in commercial irons to a dangerous extent. They are therefore, only of academic interest.
- We also note with interest the evidence for the effects of trace elements on pearlite formation. Our experience suggests that copper, arsenic, antimony, bismuth, lead, indium, tin and thallium promote the formation of pearlite in as-cast structures, but that titanium has no significant effect.
- Experience has shown that the effects of interfering elements are additive. Thus, traces of two or more elements present in amounts which individually would have no significant effect, may together adversely affect the formation of spheroidal graphite. This is illustrated by Fig. A, which shows the effect of copper on the percentage of spheroidal graphite developed in two different base irons, one of relatively high purity (Swedish charcoal pig) and the other containing normal amounts of trace elements. More direct evidence is provided by Fig. B, which



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shows, in a similar manner, the effect of lead on the graphite of the same base iron with and without an addition of 0.1 per cent titanium. These diagrams clearly demonstrate that the effect of an individual element can be reliably assessed only from tests on irons of the highest purity and that the likely response of a commercial iron to magnesium treatment can be judged only from a consideration of its contents of each of the interfering elements.

The feature of greatest commercial interest in Mr. Morrogh's paper, however, is his report of the results he has obtained with combined additions of magnesium and mischmetal. We can confirm that mischmetal, whether added before, after or together with the magnesium alloy will neutralize the effects of the important known interfering elements, and can also offer some information on the amount of mischmetal required to exert an effective influence. Our tests have shown that an addition of only 0.005 per cent of mischmetal (1.8 oz/ton), added after the magnesium alloy, is sufficient to neutralize completely the trace elements present in a badly contaminated iron and smaller amounts are sufficient for most of the contaminated irons encountered commercially. Positive effects have been observed with additions of mischmetal as small as 0.001 per cent. If however, the mischmetal is added before the magnesium alloy, or together with the magnesium alloy, either as an alloy or as a mixture, a proportion of the rare earth metals is lost as sulphides and greater additions are required to exert the same improvement. This effect is of course, greater the higher the initial sulphur content of the iron treated.

Examples of the effect of mischmetal are provided by the test results presented in Table A, which refer to the treatment of an iron of low initial sulphur content (0.02 per cent), deliberately contaminated with interfering elements.

TABLE A

No.	Additions	Mg content of test bar**	% Spheroidal graphite in test bar
A	1% Ni-Mg*** alloy	0.067	less than 1%
В	0.01% mischmetal followed by 1% Ni-Mg alloy	0.094	90–100
C	1% Ni-Mg alloy, containing 0.5% mischmetal	g 0.076	90-100
D	1% Ni-Mg alloy followed by 0.005% mischmetal	y 0.067	90-100

* All melts were finally inoculated with 0.5% silicon added as ferro-silicon

• * 20-in, diam bar cast in green sand mold.

*** 15% Mg, 85% Ni.

The table shows that when treated in the normal manner (treatment A) the iron developed practically no spheroidal graphite. When 0.01 per cent of mischmetal was added before the magnesium alloy (treatment B) an almost fully spheroidal graphite structure was developed. Tests not included in the table showed that smaller additions of mischmetal, added at this stage of the treatment, effected an improvement, but were not sufficient to neutralize completely the interfering elements present. The iron of test C was treated with a Ni-Mg alloy which had previously been alloyed with 0.5 per cent mischmetal. The mischmetal added during the treatment was therefore equivalent to 0.005 per cent and this was sufficient to restore completely the spheroidal graphite structure. Treatments with alloys of lower mischmetal content only partially restored the graphite structure. In test D, the mischmetal was added after the Ni-Mg alloy and although 0.005 per cent developed a fully spheroidal graphite structure, smaller amounts proved insufficient.

It is of interest to record here that tests have shown that cerium is the most potent of the constituents of mischmetal in neutralizing interfering elements, but that lanthanum and the other rare earths also contribute an appreciable effect.

There is little doubt that the use of the rare earths to neutralize the effects of trace elements will prove of considerable commercial value-particularly in those countries less favored with raw materials of high purity than the United States.

С. К. Donoнo: 2 Is calcium beneficial in neutralizing subversive elements?

MR. MORROGH: Calcium is not effective in neutralizing subversive elements. Although calcium can be used to produce spheroidal graphite irons, it is subject to the effects of subversive

elements as is magnesium-nickel alloy.

C. F. Walton: I merely want to confirm Mr. Morrogh's findings concerning the effects of subversive elements.

MEMBER: What is the influence of zinc on the formation of

spheroidal graphite? MR. MORROGH: The original experiments involved the control of the graphite formation in nickel-carbon alloys. In this investigation we found that a number of different elements caused spheroidal graphite formations; zinc and cadmium were among these elements. We had experimented with zinc and cadmium to produce spheroidal graphite in gray iron. Most of the experiments were unsuccessful, but one or two of these experiments would indicate that zinc "encourages" formation of spheroidal

graphite. T. E. EAGAN: 4 Subversive élements have been found in pig irons of the United States. We attribute the inconsistent results in the production of spheroidal graphite irons to lack of consideration for the presence of subversive elements. Mr. Morrogh's paper is an unusually valuable and timely contribution to the subject of nodular gray iron. Many producers will benefit by a study of this paper.

A. P. GAGNEBIN: 5 I want to point out the importance of trace elements on pearlite stability in addition to their effect on the formation of spheroidal graphite.

Mond Nickel Co., Birmingham, England.
 American Cast Iron Pipe Co., Birmingham, Ala.
 Case Institute of Technology, Cleveland.
 Cooper-Bessemer Corp., Grove City, Pa.
 International Nickel Co., New York.

## HIGH STRENGTH NONHEAT-TREATED ALUMINUM CASTING ALLOYS

By

#### Walter Bonsack*

The desire of a Design Engineer is to design a casting which will adequately serve the purpose for which it is intended. Frequently high strength is a very important factor. In designing for aluminum alloy castings, the engineer may choose two courses: these are (a) make the section of the casting heavy and apply a high safety factor or (b) choose the strongest alloy.

Heavy sections, however, mean extra weight and usually added cost. Use of high strength aluminum alloys, therefore, seems to indicate an attainment of the ideal. We must utilize a standard for measurement and for the discussion of the newer alloys by drawing on the older alloys for comparison. This discussion by necessity includes data not only on the present high strength alloys, but on the older

alloys as well.

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Up to relatively few years ago, one thought automatically of heat treated casting alloys when the term "high strength" was mentioned. Today other alloys must be considered within the same term. These modern alloys generally do not require heat treatment, but if some of them are heat treated they will surpass the properties generally found in the older more common alloys by a large margin. The newer high strength alloys are mostly of the same general type. They are aluminum alloys containing magnesium and zinc as major alloying elements. Other minor alloying elements are added to enhance some special property of the alloy.

In obtaining a basis of comparison it is important to remember that alloy qualities are measured in different ways. The design engineer is intererested primarily in the mechanical and possibly some of the physical properties. He will consider alloy quality in terms of strength, ductility, toughness, machinability and other characteristics. The foundryman who produces the casting will think about fluidity or castability, shrinkage, melting difficulties and other factors which are his primary concern.

When the design engineer considers high strength, he refers to tensile strength, yield strength, compres-

sion, shear strength and usually ductility connected with this strength. Since fairly high strength can be obtained by straight alloying only, ductility also must be considered since the combination of the two provide a sounder yard-stick. An alloy can be rendered stronger by solely increasing alloying elements, but ductility is lost rapidly and the resultant strong and brittle alloy has but few applications. One finds by choosing the composition within a usable property area that most common aluminum alloys range in strength from 17,000 to 40,000 psi and in ductility from 12 to 0 per cent. This spread is caused not only by alloying but by the different methods of casting and heat treatment.

application is limited primarily by the amount of castings required in production to amortize the die, and also by the weight and dimensions of the casting. To simplify this discussion, die casting will not be discussed. By abstaining, however, the highest strength range of aluminum alloys is automatically eliminated from the discussion. For the remaining casting methods, sand and permanent mold, the tensile strength ranges from 17,000 to 35,000 psi in the as-cast state and the ductility or elongation varies between 12 and 0 per cent. Aluminum alloys can be heat treated to improve these properties. Such treatment will pro-

Casting methods include sand, permanent mold

and die casting. Die casting is highly specialized. Its

duce either higher strength or ductility depending on the alloy and the treatment applied. Heat treatment, however, means expense. It is an extra procedure and as such calls for equipment, fuel, and handling as

added costs in making a casting.

#### Alloying Elements in Aluminum

Not all aluminum alloys are heat treated. Some do not respond to heat treatment or the gain in properties is too small to warrant the expenditure involved. These facts cause the elimination of a number of well-known casting alloys from the list of high strength alloys. The remainder are shown in Table 1 which gives their nominal composition. It is interesting to note that there are only three major alloying elements-copper, magnesium and siliconwith all others classed either as minor elements or as

^{*} Director of Research, Christiansen Corporation, Chicago.

TABLE 1-NOMINAL COMPOSITION*

Permanent Mold Casting Alloys							Sar	dcasting	Alloys		
		Cu	Mg	Si	Others		Cu	Mg	Si	Others	
1	ccl	10	0.25	1.0	_	-	10	0.25	1.0	_	
2	CN21	4	1.5	_	2 Ni		4	1.5	-	2.0 Ni	
3	cs4	4	_	2.0	-	Cl	4		_	-	
4 5	sc8\	3.5	$\pm$	6.5	-		3.5	±	6.5		
6	sc21	1.5	0.5	5.5			1.5	0.5	5.5	_	
7	sc41	1.5	0.7	12	0.6Mn		_		_	_	
8	scl	_	0.3	7			_	0.3	7	-	
9	sn41	1.0	1.0	12	2. Ni		_	-	_	-	
10	sc42	1.5	0.4	8	0.4Mn		1.5	0.4	8	0.4 Mn	
11	<b>G3</b>	_	_	_	-		_	10	_	_	

^{*} See Table 5 in Appendix.

impurities. As indicated in Table 1, most of the alloys can be used for both sand and permanent mold casting. Each group nevertheless has a few which are used for but one type of casting method, as for example alloys C1 and G3 in sand casting, and such alloys as SC41 and SN41 in permanent mold. A few of the alloys such as CG1, CN21, SC41 and SN21 have specific uses. These four alloys are used primarily for pistons and cylinder heads. Remaining are about six alloys in each casting group which constitute the high strength alloys.

Heat treatment of common aluminum alloys in order to gain high strength and increased ductility usually consists of a solution treatment, a quench and an aging treatment. Depending on the method of casting and the wall thickness of the casting, the solution treating time may vary from 4 to 24 hr at a temperature below the solidus temperature of the alloy. Depending on the final result desired, the aging treatment may be at room temperature or it may be at slightly elevated temperatures between 250 and 500 F. Most of the alloys are sensitive to sudden temperature changes occurring in quenching. This results in quench stresses, which can and may have to be removed by an aging treatment before machining to avoid or minimize distortion. All of these alloys when subjected to these treatments suffer dimensional changes (growth) which must be projected in the design and which must be eliminated by aging, particularly when the casting is used at somewhat elevated temperatures. Aging treatments of this order are either of long duration at comparatively low temperatures or of shorter time at higher aging temperatures. The first condition usually affects the ductility adversely and in the second case the strength.

Either of the heat treatments are expensive since they require costly furnace installations and, of course, fuel and time consumption. Therefore, whenever a high strength aluminum alloy is required for the job it is necessary to consider the heat treating cost in addition to the alloy cost.

The alloys previously tabulated develop various properties under heat treatment. Table 2 illustrates some typical properties obtained under certain heat treatments. In this table only a few heat treatments of the more common alloys are cited. There is a considerably larger number, but most of these produce special properties not of immediate interest. However, these tables are interesting in a number of respects when you observe that:

- (a) Permanent mold castings give higher strength and ductility for the same alloy.
- (b) Most casting alloys are used in the artificial aged condition.
- (c) Most alloys with the exception of G3 produce elongations below 10 per cent.

The above tabulations represent the status of commercial aluminum alloys prior to and partially existent in World War II. It should be observed that the only alloy which produces high strength and ductility can be cast only in sand, and because it has a 10 per cent Mg content it is hard to handle consistently in the foundry. It is therefore understandable that a great demand exists for alloys which are stronger and more ductile without the added expense and time of heat treatment.

Alloys meeting these requirements were available long before World War II. As with most innovations, however, it requires not only time and patient research to develop them to their fullest advantage, but the process of introducing the alloys and educating users further slows their application. Frequently it depends on who develops the alloys, who fosters them and how much aggressive effort is spent acquainting users with the alloys.

In the early 1920's in Germany, Guertler and Sander¹ discovered the tremendous hardening and strength-

TABLE 2—HIGH STRENGTH ALUMINUM CASTING ALLOYS (HEAT TREATED)

T 4			Sand Cast						Permanent Mold Cast								
	T 4		T 6			T 7			T 4			T 6			T 7		
. Y.S.	EL	T.S.	Y.S.	EL	T.S.	Y.S.	EL	T.S.	Y.S.	EL	T.S.	Y.S.	EL	T.S.	Y.S.	EL	
								-									
_	_	40	30	_	_	-	-	-	_	-	47	43		_	_		
_	, -	_	-	_	32	28	0.5	_	_	_	47	42	0.5		_	_	
16	8.5	36	24	5.0	39	26	5.0	-	_	_	-	-	_	-	_	-	
	_	_	_	_	_	_	-	37	20	8.0	40	26	5.0	39	20	4.5	
28	3.0	35	24	3.5	39	30	1.5	_	_	_	41	23	4.5	40	24	4.0	
_		36	24	2.0	_	_	-		-	_	40	27	3.0	_	-	_	
14	3.5	35	25	2.5	38	36	0.5	40	19	8.0	42	27	3.5	40	30	2.0	
-	-	_	_	_			_	41	27	2.5	50	41	1.0	43	38	1.0	
_	-	33	24	4.0	34	30	2.0	_	-	-	40	27	5.0	33	24	5.0	
-	-		_	_	-	-	-			-	47	43	0.5	_	_	_	
20	2.5	37	25	2.5	36	28	3.0	40	21	5.0	44	27	4.0	41	28	3.0	
25	14.0		_		-	_	-	-	_		_	-		_	-	-	
	28 ————————————————————————————————————	16 8.5 	16 8.5 36 28 3.0 35 — 36 14 3.5 35 — — 33 — 33 — 36 14 3.5 35	16 8.5 36 24 28 3.0 35 24 14 3.5 35 25 	16 8.5 36 24 5.0 28 3.0 35 24 3.5 28 3.0 35 24 3.5 14 3.5 35 25 2.5 	16     8.5     36     24     5.0     39       28     3.0     35     24     3.5     39       —     —     36     24     2.0     —       14     3.5     35     25     2.5     38       —     —     —     33     24     4.0     34       —     —     37     25     2.5     36       25     14.0     —     —     —	16     8.5     36     24     5.0     39     26       28     3.0     35     24     3.5     39     30       14     3.5     35     24     2.0     38     36       14     3.5     35     25     2.5     38     36       10     33     24     4.0     34     30       20     2.5     37     25     2.5     36     28       25     14.0	16     8.5     36     24     5.0     39     26     5.0       28     3.0     35     24     3.5     39     30     1.5       2     -     36     24     2.0     -     -       14     3.5     35     25     2.5     38     36     0.5       -     -     33     24     4.0     34     30     2.0       20     2.5     37     25     2.5     36     28     3.0       25     14.0     -     -     -     -     -     -	16     8.5     36     24     5.0     39     26     5.0     37       28     3.0     35     24     3.5     39     30     1.5     —       14     3.5     35     25     2.5     38     36     0.5     40       —     —     —     —     41       —     —     —     —     41       20     2.5     37     25     2.5     36     28     3.0     40       25     14.0     —     —     —     —     —     —	16     8.5     36     24     5.0     39     26     5.0     37     20       28     3.0     35     24     3.5     39     30     1.5     —     —       14     3.5     35     25     2.5     38     36     0.5     40     19       —     —     —     —     —     41     27       —     —     33     24     4.0     34     30     2.0     —     —       20     2.5     37     25     2.5     36     28     3.0     40     21       25     14.0     —     —     —     —     —     —	16     8.5     36     24     5.0     39     26     5.0     37     20     8.0       28     3.0     35     24     3.5     39     30     1.5     —     —     —       14     3.5     35     25     2.5     38     36     0.5     40     19     8.0       -     -     -     -     -     -     -     -     -     -       -     -     33     24     4.0     34     30     2.0     —     -     -       20     2.5     37     25     2.5     36     28     3.0     40     21     5.0       25     14.0     -     -     -     -     -     -     -     -	-     -     -     -     -     -     -     -     47       16     8.5     36     24     5.0     39     26     5.0     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -	-     -     -     -     -     -     -     -     47     42       16     8.5     36     24     5.0     39     26     5.0     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -	$\begin{array}{cccccccccccccccccccccccccccccccccccc$	-     -     -     -     -     -     -     -     47     42     0.5     -       16     8.5     36     24     5.0     39     26     5.0     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     - <td>-     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -</td>	-     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -     -	

e

ening effect of the compound MgZn2 on aluminum. This compound not only raised the strength, yield strength and hardness of aluminum, but did not decrease the ductility in the usual rapid manner as, for example, compound Mg2Si did in duralumin. There is, however, a shadow on the bright picture painted here on their discovery. This development resulted during a period of development and introduction of the aluminum-silicon and aluminum-silicon-magnesium alloys. While these alloys did not provide the properties which the Al-Zn-Mg alloys indicated, they nevertheless gave foundrymen a series of alloys which were easier to cast and gave improved, sounder castings over any alloy previously used. One cannot, therefore, blame the foundryman for reaching primarily for an alloy group which considerably eased his work. Being enthusiastic about these Al-Si alloys, they were not interested in alloys which, while giving much superior properties, were somewhat more difficult to handle in the foundry. If the literature of the 20's and early 30's is followed closely, one finds the Al-Zn-Mg alloys buried among the work then popular on Al-Si and Al-Si-Mg alloys and heat treat-

A few years later in the United States, however, one finds that T. S. Fuller and David Basch² rediscovered the value of the Al-Zn-Mg alloys. One large firm has used this type of alloy for many years in a total amount exceeding 15 million pounds³ since its incipiency. Such large usage by one company resulted from the many advantages it recognized in these alloys to be discussed later. Compstock⁴ increased the properties of this type of alloy by adding Cr and Ti as alloying elements giving the alloys, as he claims, increased corrosion resistance due to Cr and fine grain due to Ti. This alloy is known to foundrymen as 40E and has been discussed numerous times in the trade journals.

In the middle of the 1930's the present author in studying the effect of Zn on existing aluminum cast-

ing alloys found an even more powerful hardening and strengthening element in the ternary compound Mg₃Zn₂Al₂. Out of this discovery a series of new alloys were developed.⁵ These alloys generally speaking have lower zinc content than the above-mentioned 40E but have a considerably higher Mg content.

Other investigators either dwelled on the Guertler-Sander type alloy or increased the Zn content considerably and lowered the Mg content. The latter alloy is known as Tenzaloy and others as alloys A612 and C612. These three types of alloys have been used to various extents and lately have been added to the list of aluminum casting alloys of ASTM. Tables 3 and 4 give the composition of these alloys. The main variables, as mentioned, are Zn and Mg. In the first three alloys Zn ranges from 5.2 to 8.0 per cent and Mg from 0.2 to 0.8 per cent. In the last two, however, Zn is lower and ranges from 2.6 to 4.5 per cent while Mg ranges from 1.4 to 2.4 per cent. This difference sets these two alloys apart from the other three within the same group.

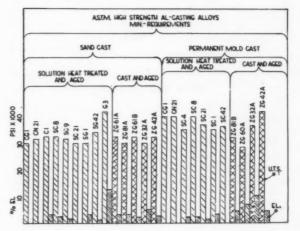


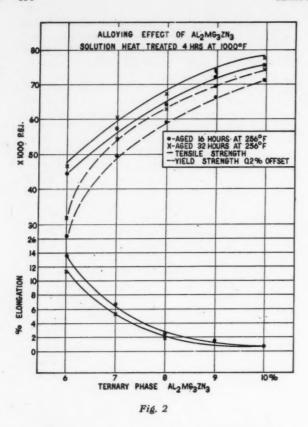
Fig. 1

TABLE 3—Composition of Sandcasting Alloys

Alloy	Cu	Fe	Si	Mn	Mn Mg Zn Cr	Zn	Cr	Ti	Ni	0	Others	
									Each	Total		
zc61A	0.3	1.0	0.25	0.3	0.5-0.65	5.2-6.0	0.4-0.6	0.10-0.25	_	0.05	0.2	
zc81A	0.4-1.0	1.0	0.3	0.6	0.2-0.5	7.0-8.0	0.3	0.2	0.1	0.1	0.2	
<b>z</b> 611B	0.35-0.65	0.5	0.15	0.05	0.60-0.80	6.0-7.0		0.20	-	0.05	0.15	
ZG32A	0.2	0.8	0.2	0.4-0.6	1.4-1.8	2.7-3.3	0.2-0.4	0.2	_	0.05	-	
zc42A	0.2	0.8	0.2	0.4-0.6	1.8-2.4	4.0-4.5	0.2-0.4	0.2	_	0.05	_	

TABLE 4—Composition of Permanent Mold Alloys

Alloy	Cu	Fe	Si	Mn	Mg	Zn	Cr	Ti	Ni	Others	
										Each	Total
zc81B	0.4-1.0	1.3	0.3	0.6	0.2-0.5	7.0-8.0	0.3	0.2	0.1	0.1	0.2
zc60A	0.35-0.65	1.4	0.3	0.05	0.25-0.45	6.0-7.0	_	0.20	-	0.05	0.15
ZG32A	0.2	0.8	0.2	0.4-0.6	1.4-1.8	2.6-3.3	0.2 - 0.4	0.2	_	0.05	_
ZG42A	0.2	0.8	0.2	0.4-0.6	1.8-2.4	4.0-4.5	0.2-0.4	0.2	_	0.05	_



For comparison, the properties of a few of the well known alloys should be shown. Figure 1 plainly indicates these alloys are as good or better than most of the heat-treatable casting alloys in this evaluation. In fact, considering these minimum figures, one finds in general the ductility is higher for the same strength figures. In some of these alloys even

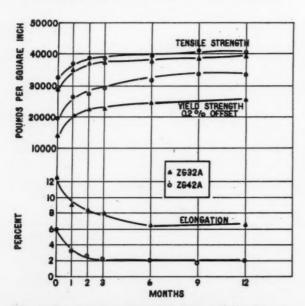


Fig. 3—Room temperature aging characteristics of sandcast alloys ZG32A and ZG424A.

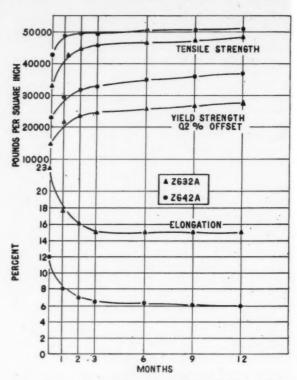


Fig. 4—Room temperature aging characteristics of chillcast alloys ZG32A and ZG42A.

higher properties can be obtained if one wants to utilize solution heat treatments. The discussion to this point has been centered generally on a comparison of the older alloys with these new ones. In Fig. 1 it was shown that these new alloys do not require solution heat treatments to be in the same class as the older heat-treatable alloys.

The two alloys ZG32A and ZG42A have been selected for a detailed discussion as they are two of the group of alloys developed by the author and on which full data are available. These alloys, as stated before, gain their properties by precipitating a ternary Al-Mg-Zn complex or compound. As shown in Fig. 2, the larger the addition of this phase, the higher the strength. It is remarkable, however, how slowly the ductility decreases compared to the older known al-

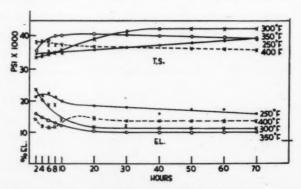


Fig. 5—Aging characteristics of ZG32A, chillcast at various temperatures.

loys. It should be observed here that Fig. 2 shows the alloys in the heat-treated state. This purposely was done to illustrate that the alloys are capable of being heat treated and that properties unheard of in the casting field can be obtained. Tensile strength of over 70,000 psi and yield strength of over 60,000 psi still with 1 to 2 per cent elongation certainly is remarkable. Alloys containing 10 per cent of this ternary compound are ductile enough to be rolled or forged. Sheet rolled of this alloy develops under certain heat treatment 88,000 psi tensile strength, 80,000 psi yield strength and 8 per cent elongation. Because these properties are so high in the heat-treated state, it is logical to find them also extremely high in the as-cast state.

Being susceptible to heat treatment, it is also logical that they age well at room temperatures. Figure 3 shows the aging curves of two of these alloys when sand-cast. These two alloys were chosen by the ASTM B-7 Committee because ZG32A represents an alloy of good high strength with very high ductility while alloy ZG42A has lower ductility but proportionally higher yield strength and tensile strength. It should be noted the tensile strength of these alloys as sandcast without aging are on a 28,000 to 35,000 psi level and the elongation depending on the alloy varies from 6 to 12 per cent (Fig. 3). Aging at room temperature changes these properties quite rapidly but after three months they become quite stable, the tensile strength level having increased to 36,000 to 41,000 psi and the elongation having decreased to a

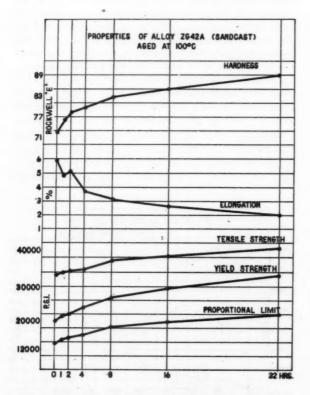


Fig. 6—Properties of alloy ZG42A (sandcast) aged at 100C.

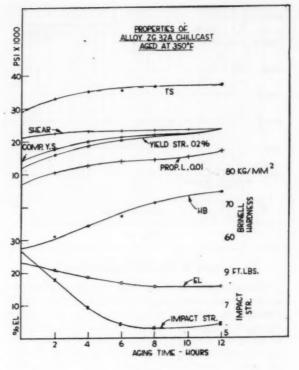
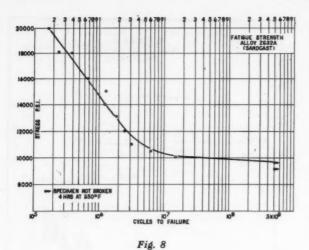


Fig. 7

2 to 8 per cent level depending on the alloy. The same general picture is obtained by chill-casting them, but the levels obtained are much higher both in strength and ductility. Figure 4 shows the alloys when cast in permanent mold. In the freshly cast state, the strength level is now 33,000 to 43,000 psi and the elongation 12 to 23 per cent depending on the alloy. Aging at room temperature increases strength to a level of 45,000 to 50,000 psi after three month aging and the elongation decreases to a level of 6 to 16 per cent. A relation of properties of this order should be of considerable interest to the designer since by simple alloying he can change the property relations to his needs and still not change at all or only slightly the desirable general characteristics.

The high ductility of these alloys right after casting also has great advantage because it allows a considerable amount of cold work to be done to the casting without harming it. Bending of the casting can be accomplished without cracking. Some foundrymen and designers have taken advantage of this high ductility characteristic. A striking example of this is in a gearshift lever for a large tractor which needed two 45-degree offset bends within 6 in. To cast this lever with these bends required a complicated pattern and parting line arrangement in the mold. The lever was cast straight and was then bent into final shape without subtracting from its properties. This lever had been cast in sand and since the ductility of chillcasting is nearly double that of sand casting, considerable more cold work is possible opening new avenues for clever designing.



## Aging at Elevated Temperatures

Although these alloys develop very good properties at room temperature, equally good properties can be obtained at elevated temperatures. In cases where testing or acceptance inspection cannot wait for natural aging at room temperature, elevated temperatures can be used to obtain similar results. Figure 5 shows the aging characteristics of alloy ZG32 chillcast when aged up to 70 hr at 250 to 400 F. At 250 F the aging process is relatively slow but 400 F overaged the alloy quite rapidly. In all cases, however, the strength is above 34,000 psi and the elongation never lower than 10 per cent. Even aging at 212 F in boiling water promotes faster aging than room temperature. Figure 6 illustrates the aging characteristics at this temperature for alloy ZG42A. Boiling castings for 8 hr in water produces almost the same properties that room temperature aging produces in three weeks. As it can be seen in the diagram, 35,000 psi tensile strength, 3 per cent elongation, and 26,000 psi yield strength can be obtained for sand castings.

Figure 7 presents the mechanical properties of alloy ZG32A when cast in permanent mold and aged for various times at 350 F. As mentioned before, 350 F is a fairly high aging temperature and tends to age the alloys quite rapidly. Still, it is interesting to note what properties can be obtained, particularly for comparison with common solution heat-treated and aged aluminum alloys. Most of these alloys are aged between 300 and 400 F. One of the most interesting curves is that of impact strength. Although falling off quite rapidly for the first 8 hr, it begins to recover beyond this time. Still, its lowest point reached is over 5-ft-lb which is higher than any other common commercial heat-treated alloy. The lowest elongation level is above 15 per cent and considerably higher than for any other alloy. As can be expected with this type of heat treatment and with this high elongation and impact strength the elastic properties are not too high, but compared to many a commercial alloy they are right in line. The tensile strength is competitively between 30,000 and 40,000 psi.

Figure 8 shows the fatigue strength of an alloy of this type when sandcast and aged 4 hr at 350 F. At

500 million revolutions it shows an endurance limit of about 9,500 psi which again is as good or better than in most common alloys.

Figure 9 gives the strength and elongation data of ZG42A alloy chillcast and tested at elevated temperatures. The test used is the standard ASTM short time test. The alloy was tested in two conditions: (1) In the solution heat-treated and aged state, and (2) in the cast and aged state. The aging treatment for both was 8 hr at 300 F. It is indicated plainly that the solution heat-treated alloy loses its strength proportionally faster than the non heat-treated alloy. This is true for all aluminum alloys. ZG alloys are not noticeably affected by heat till the temperatures reach 250 or even 300 F. Dropping off in strength occurs beyond 350 F; the elongation, however, begins to climb rapidly beyond 400 F. The relatively fast drop of strength at the beginning of the curve indicates these alloys are not as good as the Al-Cu alloys. Therefore, the Al-Mg-Zn alloys do not recommend themselves for high temperature applications.

#### Other Properties

The specific gravity of alloys ZG32A and ZG42A is around 2.75 and the weight per cubic inch 0.096 lb. The thermal conductivity is about 0.22 (c.g.s. units) and the electrical conductivity 26 per cent that of Cu.

Corrosion. The alloys exhibit a very high corrosion resistance and are to be classed with the best. The ASTM has these alloys under test in comparison with other casting alloys in locations all over the country including residential, industrial, seashore and subtropical districts. Results on these alloys will not be obtained for perhaps a year but short time salt solution immersion tests plainly indicated the superiority of these alloy types over the existing ones.

The color of these alloys is nearly a silvery white and their solid solution constitution permits an easy high luster polish. Anodic oxidation produces a casting of uniform pleasing appearance.

Brazing. The solidification temperature of these alloys is above 1100 F and therefore permits fabrication of assemblies by brazing. Brazing on aluminum alloys usually is successfully done at temperatures

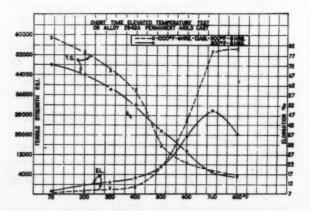


Fig. 9

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between 1000 and 1100 F. Most common aluminum casting alloys will not stand up at these temperatures; these alloys, however, will not only stand up but will not lose in properties. Brazing of castings to sheet or forgings and other types of material therefore is possible.

Machining. Contrary to the case with some aluminum alloys, these alloys because of their composition are machined quite freely and milling speeds of 400 fpm are reported without noticeable wear on the cutting tool.

#### Heat Treating Affects Dimensions

Dimensional Stability. Most known aluminum alloys exhibit considerable dimensional changes when heat treated or aged. They grow so noticeably in dimensions that this fact has to be considered seriously in design and application. The Al-Zn-Mg alloys do not grow, in fact to be extremely critical, they never indicate but minute shrinkage. Measurements have shown a shrinkage of 0.0002 in./in. after 70 hr aging at 350 F but only 0.00003 in./in. after 70 hr at 250 F aging. After aging 8 hr at 210 F no change in dimension could be ascertained. This property alone makes the alloys highly desirable for any application where dimensional changes, even in a small degree, interfere with the usefulness of the product. Projectors, calculating and business machines, instruments such as gyroscopes and others are being made from this type of alloy because of its dimensional stability.

Castability. As mentioned in the beginning of this paper, alloys of this type are not as easy to handle in the foundry as the Al-Si alloys; however, any good foundry will be able to use these alloys to definite advantage if they understand their basic nature. Being relatively high in Mg content, they must be melted with care so that no Mg, or very little is lost. Careful skimming will give free flowing clean metal. Sections as thin as 1/8 in. can be successfully cast. Feeding and shrinkage is equivalent to the well know heattreatable 4 per cent Cu alloys and adequate feeding and risering will take care of this alleged deficiency. It is said that the alloys are apt to crack in permanent molds due to hot-shortness. Several ways exist to eliminate this problem. One of the simplest ways is to decrease the solidification cycle to prevent any cracking. Practical experience has shown that castings made from this metal must be ejected from the mold 10 to 20 per cent faster than Al-Si alloys. This can be considered an advantage rather than a drawback since proportionally more castings will be produced within a given time. Another recommendation is to use greater fillet radii; more careful blending of sections is desireable for these alloys as for all casting manufacture. It will not only make it easier for the foundryman to make improved castings but will make all castings much stronger. Bean6 made this fact very clear in his publications and lectures on casting design. These alloys are not as sensitive in properties in regard to cross-sectional changes as are most com-

A word of caution should be given to the foundryman. The high properties of the alloys depend on the Mg-Zn or Al-Mg-Zn complex compounds. These are

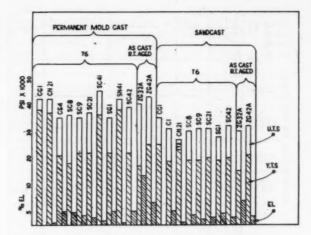


Fig. 10—Comparison of heat treated casting alloy properties non heat-treated Al-Mg-Zn alloys.

easily destroyed by the addition of silicon. In all alloys of this type, it is imperative that the silicon content be kept as low as indicated in the specifications and, preferably, lower. This means, in addition, that contamination with silicon either from silicon alloy scrap or from molding or core sand must be avoided. This, of course, is no problem in any well kept foundry.

#### Summary

Al-Mg-Zn type alloys are the modern high strength alloys. They obtain the same or higher properties without heat treatment as provided by the older heat-treatable casting alloys. Figure 10 gives a comparison of the typical properties of the older alloys with two alloys of the newer type. Typical properties of the other three alloys mentioned herein are not available, but experience has shown they fall nearly within the same range as the two described. Alloys of this type require exacting foundry practice and are somewhat more expensive than the common alloys due to the high purity metal required by their composition. Other important advantages of dimensional stability, excellent machinability, high luster, corrosion resistance and other characteristics make them an excellent choice for quality castings. Their ability to be brazed make them an exceptional engineering material.

#### References

- 1. Wm. Guertler and Wm. Sander U.S. Patent 1,629,699 (1927).
- 2. U.S. Patent 1,740,549 (1930).
- 3. L. W. Eastwood and L. W. Kempf, "Al-Mg-Zn-Cu Casting Alloys," Transactions, American Foundrymen's Association, vol. 56. p. 114 (1948).
- 4. U.S. Patent 2,146,330 and 2,146,331.
- 5. W. Bonsack U.S. Patent 2,240,489, 2,249,740, 2,274,657, 2,290,216 25.
- 6. Wm. T. Bean, "Simplification of Light Metal Casting Design and Its Effect upon Serviceability," Transactions, A.F.S., vol. 55, pp. 430-440 (1947).

#### DISCUSSION

Chairman: W. J. KLAYER, Aluminum Industries, Inc., Cincinnati.

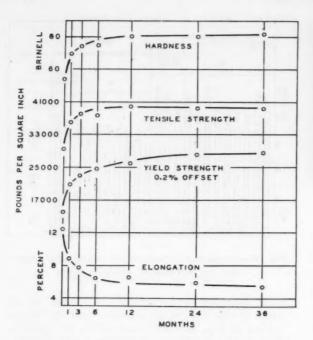


Fig. A—Room Temperature Aging Curve for Sand Cast ZG32A.

Co-Chairman: R. C. Военм, Wellman Bronze & Aluminum Co., Cleveland.

Recorder: W. J. KLAYER.

WM. N. BRAMMER (Written Discussion): We wish to congratulate Mr. Bonsack on his excellent paper and to express our complete agreement with the points which he brings out. We believe the following additional evidence on the characteristics of the ZG32A and ZG42A alloys will supplement his presentation.

The author mentions a milling speed of 400 fpm without noticeable wear on cutting tools. One user of ZG32A is milling at a surface speed of 2,500 fpm and at times goes up to 4,000 fpm without lubrication. He has taken cuts up to 1/8 in. at

Fig. C-Street Lighting Pole Plate Castings.

these speeds, although the metal removed in production is much less than this. He also drills a 5/16 in. hole at 3600 rpm drill speed and receives excellent finishes at these speeds.

The remarkable dimensional stability of these alloys is one of the most outstanding characteristics, thus it becomes possible to obtain optimum results with light weight castings in assemblies of optical systems, binoculars, prism assemblies, gauge equipment and many other miscellaneous applications wherein precise dimensional accuracy is a prerequisite for ideal operation.

The author, in Figs. 3 and 4 of his paper, offers charts showing changes in the mechanical properties with aging up to 12 months. We have considerable data for longer periods as shown in Fig. A. This shows changes in mechanical properties with three years' aging of sand cast ZG32A. The properties shown after the first few months are representative of the excellent ultimate properties developed after long term aging. We are confident that the whole story is told in the first few weeks, as shown in the author's charts.

The author points out that the two alloys ZG32A and ZG42A depend on the ternary compound  $Al_2Mg_0Zn_3$  for their properties and those differ in composition from the other high strength alloys which he has listed. The aluminum corner of the ternary aluminum-magnesium-zinc equilibrium diagram is given in Fig. B. The Al-Mg-Zn alloy ZG61A mentioned by the author is in the field designated  $\alpha+MgZn_3$ . Others which he mentions are far down almost to the binary Al-Zn base line, so that their properties are developed from another source rather than from the powerful hardening and strengthening compound  $Al_2Mg_9Zn_3$ .

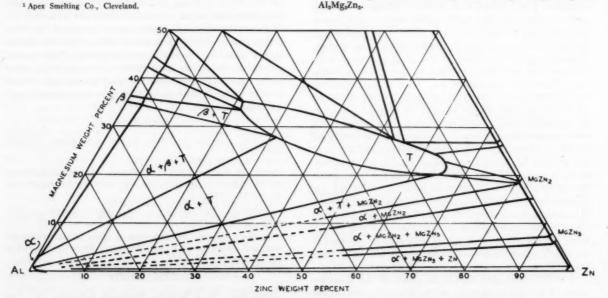


Fig. B-Al-Mg-Zn-Distribution of the Phase in the Solid. (After Koster and Dullenkopf)

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Fig. D-Pole plate casting.

The author mentioned the excellent brazing properties of these alloys and set their solidification temperatures at above

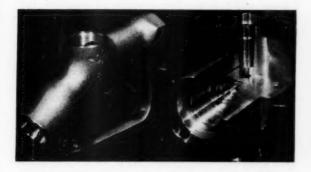


Fig. E (Left)—Gasoline pump valves (Right) Highpressure oxygen valve.

1100 F. However, our latest laboratory tests show the solidification range of ZG32A and ZG42A as being 1030 to 1170 F. We would suggest, therefore, some additional investigations on this subject inasmuch as present data indicate a solidification temperature somewhat lower than the 1100 mentioned by the author.

Figure C is an end view of a ZG32A street lighting pole plate casting which offers evidence of the high mechanical properties combined with exceptional ductility obtainable with this alloy in the chill cast technique. The casting is seen on the left which by vise action has been press deformed to the position on the right. This casting is used on top of conventional street lighting poles and holds the elliptical arm which in turn supports the lighting fixture.

The pole plate casting in Fig. D is made in a single-cavity die with thin, uniform sections spread over a varying contour with three complete holes as-cast and a 4-degree taper on the internal core which has a draw of over  $3\frac{1}{2}$  in.

Figure E (Left) shows one of a series of sand-cast mediumvelocity gasoline pump valves being manufactured. Anti-sparking, pressure-tight, easily machined and light weight, it is an excellent ZG32A application in a common, every day use.

Figure E (Right) shows a permanent-mold ZG42A highpressure oxygen valve with a machined quadrant showing the extensive counterboring, threading and drilling, as well as the mass cross-sectional area of this casting. There are over 20,000 of these units in operation today.

Figure F illustrates a sand-cast tractor hitch which we feel vividly depicts the ease of castability. Here we have a casting some 29½ in. in lateral diameter with radical cross-sectional area changes which was cast with little or no foundry difficulty.

We hope this additional information will prove enlightening and interesting to those who are considering this type of alloy.



Fig. F-Sand-cast tractor hitch.

#### **APPENDIX**

Since this paper has been written, the ASTM designations used throughout this paper, have been changed as shown in Table 5.

TABLE 5

Old Alloy No.	New Alloy No.	Old Alloy No.	New Alloy No.
CG1	CG100A	SC21	SC51A
CN21	CN42A	SC41	SC122A
CS4	CS42A	SG1	SG70A
Cl	C4A	SN41	SN122A
SC8	SC64A	SC42	SC82A
SC9	SC64B	G3	G10A

## THE WHYS AND WHEREFORES GF STATISTICAL QUALITY CONTROL

By

#### Wade R. Weaver*

During the past two years the published material on Statistical Quality Control has been increasing at an astounding rate. Despite this greatly increased dissemination of not only propaganda but of case histories, one still hears all too often the bewhiskered statement "It's fine for the other fellow but I can't use it." An even more disconcerting fact is the widespread lack of understanding and appreciation of the breadth of scope of Quality Control and its kit bag of varied tools. All to frequently it is referred to as "an Inspection Tool," rather than a Management tool, which means something entirely different. It is the purpose of this paper to dispel both these beliefs and to bring out a few fundamental reasons why any industry, and the Foundry Industry in particular, can and should avail itself of the potential advantages offered by Quality Control.

Management in industry is constantly being called upon to make decisions. That is one of its primary functions. When we say "management" we do not necessarily imply the top, or near the top level of administration. Management extends down through the superintendent, the foreman, the inspector, the engineer, process chief, even down to the single individual performing a special operation or exercising supervision over an integral part of a process. It can and does mean anyone who is managing anything, almost irrespective of his field of endeavor or his level in that field. It goes without saying that anything which will sharpen one's judgment, present facts rather than opinions, and aid in making the right decision at the right time must be of valuable assistance to management. Statistical Quality Control when properly used does exactly that.

#### **Elements of Quality Control**

Modern Quality Control has been referred to as a kit bag of tools. It is not a *single* tool. When the plumber is summoned to repair a disastrous failure he surveys the job at hand, opens his tool kit, selects the proper tools, and attacks the problem. So, also, does the trained Quality Control engineer proceed. It

is true that some of the tools of his trade are quite complicated by comparison to others and, hence, require some specialized training to use effectively, but some of them are almost childishly simple in this respect. Some of the rather basic ones, and certainly the most widely used and effective ones, are founded on principles which are familiar to all of us, if not completely understood.

One of the basic concepts is that there is variability in all things. Instead of ignoring this fact we should recognize it, study it, and profitably use it. Closely allied is the behaviorship of numbers, the laws of chance-a pair of dice is the most common example. These laws require a little study and they have been carefully studied until it is possible to calculate just how often a given value will appear, or a certain event occur, when that occurrence is due to chance causes alone. Further, it is quite simple to calculate the odds against such an occurrence of a specific value oftener than a given percentage of times. Exhaustive study of literally thousands of "events" have shown that there are certain predictable patterns to which there is universal conformance when only chance causes are at work. The mathematicians have worked out very simple ways of presenting these events pictorially, and two of these picture forms are represented by the "frequency distribution curve," and the "control chart." Most of us have at least seen both these types of pictures.

The former shows how often or what percentage of the time any specific value occurred, the "frequency of occurrence," or relative frequency. The latter shows the relative level which those values had as they occurred, or were produced, one after the other, or chronologically. An important part of such a chart are the limit lines, boundaries within which the values would be expected to lie.

Practically everything we can think of "produces" values—measurements of variation. Often we do not actually measure them. Perhaps they are difficult to measure, but it is possible and usually practical. If the study of a set of data—measurements from a process, material or part—indicates the factor we are studying behaves normally, is conforming to normal distribution and normal relative frequency of occur-

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rence, then we are entitled to say only normal chance causes are at work. On the other hand, if the data are definitely non-conforming in nature, we know that some extraneous cause or causes, some non-normal cause, an assignable cause, is at work.

#### The Control Chart

The Control Chart provides a normal level of values as well as top and bottom limits beyond which we should only rarely expect to find values fall. If a process or any measurable situation is behaving normally and only chance causes are at work, it is said to be "in control." Under such conditions values will lie within the upper and lower "control limits" and be rather evenly dispersed about the normal level or average. If there are assignable causes present the process is certain to behave abnormally, values will lie outside the limit lines, and the process is said to be "out of control." Corrective action is indicated and proper action generally results in bringing the process back "in control."

There are generally a rather large number of variables affecting a given situation, each of which contributes but a small share to the result. All of these vary independently and the combined result varies. But, since any given factor usually has but a small share of the contribution, it is generally quite costly and rather fruitless to try to do very much about it. On the other hand, those causes contributing to the non-normal behaviorship are generally small in number, frequently only one or two, and each of them exerts a very strong effect on the result. Due attention paid to assignable causes is generally very productive.

#### **Normal Process Operation**

The properly analyzed past history of a process, operation, quality, situation, enables the suitably trained individual to construct simple pictures of how it should be expected to behave normally. Current situations immediately indicate the presence or absence of those factors which probably will respond to action. These charts are no substitute for experience since they do not indicate what action to take. They only indicate that action is required. Hence, "operating know-how" is still an essential. The charts tell simple stories such as whether today's level is "significantly" different from what should be expected. It may be higher than yesterday's level, but is this merely the result of a combination of several of the higher values which occurred by chance alone? If so, there has been no "significant" change. But, if the process has really and truly changed its level this will be reflected in the chart and the different level can then be said to be significantly different. The charts tell where certain values are occurring more frequently than normally, when some "now" cause has become operative.

Sometimes there is a rather long-range trend to the process, one which is not easily detected. The charts show this condition quite clearly. Whether or not it is advisable, worth while, or necessary to take specific action is always important, and of even greater importance is known when *not* to take action. The charts provide great help in making such decisions.

Thus is management's judgment sharpened. Thus is management's time saved and energies concentrated where they will be most effective. Thus is the necessity to make a decision and the resultant correctness of that decision very greatly aided.

Virtually, there are an endless number of situations in the Foundry Industry where these simple tools are almost revolutionary in their effectiveness. From the variation in raw materials, tools, individuals, and machines down through the sand mixing, coremaking, melting, molding, and casting operations the possible applications are legion. The chemical and physical properties, dimensional characteristics and machinability, temperatures and times are things we will recognize as varying constantly. Control charts are ideal tools to study and control such variables. Clear down to final inspection, packaging, shipping, customer relationship, sales problems, and finally to profits, these simple techniques are most readily applicable. They are available to all levels and kinds of "management" and there is a type to fit nearly every conceivable situation.

#### Quality Control and Research

Behind the functioning of the process as it presently exists—and one can be quite elaborate on the definition of "process"—lies another comprehensive field that may be loosely called research. While it may not be recognized as research, may not be so designated, all kinds of industries are constantly trying to improve their process, trying to sharpen their control. Things must not necessarily emanate from a large department with a heavy door marked "Research Department" to be truly research. Most of us do research and development work every day without realizing it. Our constant striving to produce greater quantities of a better product at lower cost is one big research project.

Accordingly, it seems evident that if there are techniques to aid in research, tools which will help obtain results more quickly and at less cost, means of correctly evaluating the results of research, methods of obtaining facts rather than personal opinions, such techniques should be pursued. Well, there are! The Quality Control kit bag is full of them.

One of the major objectives of research is a new level of a certain factor, such as more production, ' greater hardness, closer tolerances, lower costs, etc. It, therefore, becomes vital to know with a specified degree of certainty whether the "new" level is truly different from the "old," or whether the experimental values just happened to reflect a difference. Remember that variation is always present. How do we know the values exhibited by the "new" situation are not merely other values from the same identical situation as the "old" one? Averages alone have come to have but little meaning. What we really want to know is whether there is a "significant" difference in level. Can the difference in level be merely due to chance, did it "just happen," or can it be counted on to exist? If the difference in levels is actually significant, what is the probable true value of the difference? Between what limits does it lie and how often can we be sure that it will be so? Have the

decisions drawn been based on a large enough amount of data, or "how much is enough?" All these are extremely important questions, and the very best and most accurate answers we can possibly get are re-

quired for correct decisions.

The more probably truly correct answers to these and similar allied questions can be greatly aided by a series of Quality Control tools known as "Difference" and "Significance" tests. They tell us what reliance we can place upon our interpretations. If the proposed change is simple to execute, requires but little money and will not importantly affect the product, one needs but little assurance and may readily gamble on his decision. If, on the other hand, to institute the proposed change involves the expenditure of many thousands of dollars, one needs a high degree of assurance that his interpretations are correct, and the exact extent of the expected change must be known in order to be properly evaluated. The "confidence" tests are a vital part of research, for without them personal opinion becomes too great a factor. All of us have seen programs which had much promise as based on research work fail completely to produce the claimed results when tried on a production basis. The reason seems quite obviousincorrect interpretation of data. These tools give facts.

In examining any situation, any process, with an eye toward improvement, we are constantly tracking down variations from standard, unwanted variations. It frequently becomes necessary to know whether the variations we are finding come "within" lots or "between" lots. What is the source of the variations in dimensions, for example, in a lot of 100 castings made by pouring ten different molds, each having ten separate mold cavities? Are the variations truly random throughout the 100 castings so made, or are the variations in those ten castings from mold No. 3 significantly different from the variations found to exist on the castings made from the other nine molds? In the case of hardness variations, are they greater within one casting, within the castings made from one melt, or is the variation found to exist more importantly from melt to melt? Again, if such differences are found to exist, what is the possibility that these are merely chance variations? No ordinary analysis of complex data of this kind will provide the factual answers but there are techniques known as "Analysis of Variance" that will provide the answers.

#### Variance Analysis Techniques

Variance analysis techniques are extremely valuable in experimental work for, in addition to removing personal bias and providing much factual analysis of the data obtained, they greatly reduce the extent of experimental work. Less than 20 per cent of the number of experiments usually necessary with the conventional approach to data analysis will suffice when the program is planned and conducted to use the variance analysis approach. Thus, considerable saving of time and, of course, an accompanying reduced expense of conducting the research is the reward for a scientific and objective approach to our problem.

Of particular interest to experimenters are the "relationships" that exist between the several variables which are included in our "process" for we are constantly seeking cause and effect situations. Conventional practice is to make a chart, plotting such a relationship between the variables. We wish to see what effect a specific variable has on the results of our "process." The relationship thus depicted would be true only if there were no other variables involved. If a third variable is involved a chart can still be constructed, although it is not readily drawn and less readily interpretable by most of us. When a fourth, fifth, and sixth variable is involved, the chart technique no longer suffices. Here, again, a specific tool in the Quality Control kit bag comes to the rescue of those forced to deal with these multi-variate situations, and because so many of our present day processes are so highly complex, such a tool becomes indispensable to the analysis of such complexities. This tool is called "multiple correlation."

#### The Scatter Diagram

What we are really interested in knowing is the true and independent relationship of one variable to another, regardless of the effects exerted on the latter by other variables present. The conventional approach by the "scatter diagram" method gives only the apparent relationship of the two variables under study, the relationship that exists concurrent with variations of all the other variables present. Hence, conclusions drawn from its picture are frequently misleading. The true net effect of the single variable under study may be quite different from the apparent effect. To make an improvement in a process we must know just how the variations of one specific factor truly affect the process and its product before we attempt to alter or control this specific factor.

The orthodox experimenter tries to hold all variables constant but one and controls variations in that one. He then proceeds to the second variable and controls its variations while trying to hold all others constant, etc. Not only is this quite time consuming, but it is frequently highly impractical, if not impossible. The fact that amazes so many people is that such a procedure is not at all necessary. Let all the variables vary at will-multiple correlation will sort them out and provide unbelievable amounts of information about each and every one of them. The complex operations with which we deal today just simply will not stay put, so why waste time and energy trying to make them? Use the modern technology designed for such occasions and get the facts, not opinions.

Certain of these Quality Control tools are admittedly a little complex and some special training is necessary before they can be used effectively. This is no different from many situations with which we are familiar. New tools, new equipment, new materials, all require some careful training to be used properly and profitably. Quality Control engineering is a new tool.

Surely you will agree that the situations described above are quite commonplace occurrences in the every day life of many of us, regardless of the type of industry with which we are associated. The Foundry Industry is doing all of these things continually, and would probably like to do more but may not know just how. All industries have many unanswered problems, many things over which better control is desired. Like all other industries the Foundry Industry is constantly striving to improve quality and lower costs. The rapid infusion of some of the techniques mentioned, and there are many more useful ones too, into modern industry of almost every conceivable type, makes it possible to state with assurance and definiteness that Statistical Quality Control is proving to be indispensable to a progressive industrial entity.

With the constantly rising cost of a unit of material and a man-hour of labor, many approaches are economically sound today which a few years ago might not have been. Anything which will aid us in conserving man-hours is practically a "must." Anything which will enable us to more effectively utilize the efforts of supervision is a boom to our every-day industrial life. If we can use our raw materials more productively we automatically contribute to the improved standard of living which is our constant goal. Quality Control should not be considered as a drug which will cure any disease, but it certainly helps

cure many of our common troubles. It is a system of analysis which has been accepted by progressive management in practically every type of industry. It has helped ours; it can help you.

In summary, it should be stressed that Quality Control is a broad collection of useful tools rather than a single tool. It might be called "The company system of a scientific and objective approach to any of its problems." Many of these tools are quite simple to use and require very little training to operate. Others are a little more complex and require some fairly careful study for them to be used effectively. Their use is in no sense restrictive, but rather it should be stressed that there is some tool, perhaps several, for practically any kind of a problem that arises. Quality Control does not replace anything. It is a sound supplement to the present methods. The use of this system will sharpen up judgment by providing facts instead of opinions. It aids in obtaining the correct answers to highly complex industrial problems and, in so doing, saves time, energy, and mistakes. One of its strong points is that it tells what action is needed and, of even greater importance, when not to take action. Action on Quality Control is needed by all those not familiar with its potentialities.

## **CORE BOXES FROM CORE PLUGS**

By

James N. Mathias*

An ever increasing amount of interest has been generated these past few years in the process of developing aluminum core boxes from wood or metal master core plugs. There are two important and correlated reasons to which this interest may be attributed.

1. The advantages to the pattern shop previous to obtaining the casting. These advantages are:

(a) Making a master pattern core plug from the pattern layout makes for a greater degree of accuracy since the same amount of shrinkage can be allowed for both master pattern and master core plug.

(b) In a large proportion of patterns requiring cores, master core plugs can be made in less time than a core cavity.

(c) The master core plug can be proved in a plaster mold of the master pattern thus eliminating the making of plaster plugs from core box cavities.

(d) The process can eliminate the making of wood master drier patterns.

(e) One master plug will usually suffice for multigang boxes.

2. Since this method of making core boxes and drier patterns involves the use of plaster molds as a casting medium, the dimensional accuracy and smooth surfaces as cast inherent in this process can substantially reduce the machining time normally required; or, in many cases, entirely eliminate it.

The intent of this paper is to show by actual photographs the pattern equipment furnished for use in the process as constructed by various pattern shops.

#### Single Gang Dump Box

Figure 1 shows the mahogany master core plug and ribbing frame furnished for a single gang dump box. It should be noted that the frame placed over the plug leaves small clearance around the base of the master plug, but that considerable clearance is shown between the plug and the cross rib of the backing frame. This rib clearance should approximate the metal thickness desired. If machine finish should be desired then sufficient stock should be added to the

master core plug and a like amount to the face of the ribbed backing frame. The process then proceeds by taking a plaster from the master core. From this cavity a second plaster is taken to provide a plaster duplicate of the master core (Fig. 2). The ribbed backing frame is placed over this plaster duplicate and the exposed areas are covered with clay sheets. These clay sheets are made to the same thickness as the metal wall desired. This step may be clearly seen in Fig. 3.

Another plaster is taken from this assembly and after ribbing and clay sheets have been extracted the backing half of the plaster mold is complete as shown in Fig. 4.

After oven drying, the mold may be either gravity or pressure cast producing a single cavity dump box as shown in Fig. 5.

Should a multiple gang box be required one master core plug would suffice. One plaster cavity is taken from the plug and as many plaster duplicates as may be indicated are produced from the cavity. These duplicates are then assembled to conform to a layout furnished. The ribbed backing frame should then be made up to embrace the entire gang.

#### Single Gang Split Box (Open End)

Figure 6 shows a split mahogany master pattern plug which is provided with pins for the proper alignment of the two halves. Also shown are backing ribs for both cope and drag halves of the core box. The cope backing frame as shown placed over the cope half of the master core plug incorporates bosses both for blowing and mounting. Pads for pinning the halves are also provided as an integral part of the ribbing structure. It may also be noted that the backing frame in this instance forms a rather snug fit around the master core plug. On split or booked type core boxes this is very important as misalignment of the backs after matching the cavities can render a box useless.

It is not necessary for the backing frame to follow all contours of any given plug but rather to touch the plug in four to six places in order to insure the proper mating of both the cavities and the outside contours of the core box face. The backing frame for the drag half of the box is nearly identical to the cope

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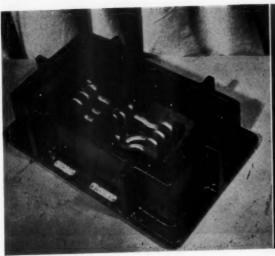


Fig. 1

Fig. 2

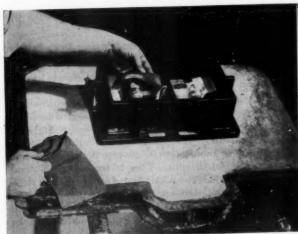
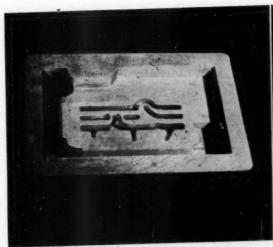




Fig. 3

Fig. 4



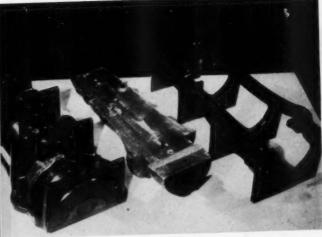
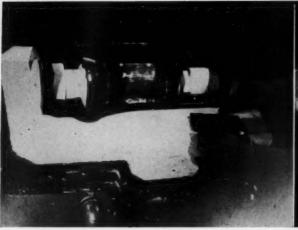


Fig. 5

Fig. 6



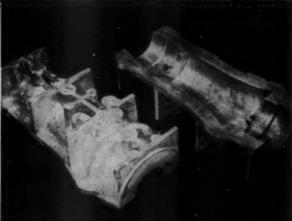


Fig. 7

Fig. 8

half minus the blow bosses and mounting lugs. Therefore, if this core box were engineered to be hand rammed instead of blown it would then only be necessary to furnish one backing frame for both halves of the core box instead of two. Machining stock was not provided on the master pattern plug and therefore was not added to the backing frames.

Obviously, all the expert engineering on this core box was accomplished by a pattern shop in close cooperation with the sand foundry selected to produce the castings. Also, the mahogany master core plug and backing frames reflect an awareness of the problems encountered in the plaster foundry charged with producing a core box.

The steps taken to produce a core box from the pattern material shown is similar to the process followed in making a single cavity dump box. Plaster cavities are taken from each of the halves of the plug. A plaster duplicate of each half of the master pattern plug is taken from the cavities. The backing frames are placed over their respective plaster halves and open areas are backed out with a sheet of clay of a thickness to correspond to the metal wall desired (Fig. 7). A plaster is taken of each assembly

and the clay and backing frame removed to reveal a complete plaster mold of each half of the core box. These molds are then dried and poured to produce the aluminum castings shown in Fig. 8.

#### Single Gang Booked Core Box and Master Drier Pattern

A solid mahogany master pattern core plug having an irregular parting, a ribbed frame each for the cope and drag half of the box, and a ribbed frame for the drier pattern (Fig. 9) form the equipment necessary to produce the single cavity booked box and white metal master drier pattern shown in Fig. 12.

Master pattern core plugs with irregular partings will produce more accurate results if they are made solid rather than split. This is true because wood split pattern halves seldom match each other perfectly before the plasters are made. After plaster slurries have been poured over the wooden halves it is seldom that a perfect mating of the pattern halves can be made.

In the single gang split box (Fig. 6) a split mahogany master pattern core plug was satisfactory and properly designed because the core box is, theoretically, two dump boxes to be assembled as one unit,

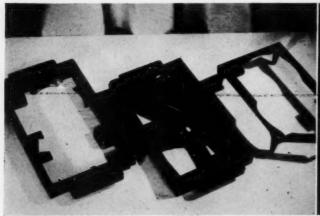




Fig. 9

Fig. 10





Fig. 11

Fig. 12

whereas, in the booked box irregular partings must match perfectly. Should warpage occur in a split pattern plug with a flat or straight parting, it is a simple task to secure the halves to wood or metal flats, but, master pattern plug halves with irregular partings can only be pinned to each other and it is difficult to determine whether the warped half is being brought back into line, the assumed good half brought out of line, or a little bit of compromise by both halves.

The plaster foundry has no way of determining a discrepancy at this point and therefore makes a core box that may produce faulty cores in the sand foundry. Should this occur in an early experience with the process it would only be human to feel that the process leaves too much to be desired for general use.

Returning to the pattern equipment necessary to produce a booked core box and drier pattern (Fig. 9) it can be seen that the solid master pattern core plug is contacted and thereby located by both of the backing frames in five places. Due to the irregular parting only four places were used on the backing for the drier master.

The first operation begins with forming a parting or match for the master core plug. This parting contour must of course conform both to the parting of the core plug and to the confines of the backing frame.

The match on the more complicated type of master core plug should be constructed by the pattern shop as the backing frames can then be constructed on the parting block thus eliminating much guesswork.

A full plaster mold is made from the master plug and plaster duplicates of the respective halves of the master plug are made. Backing ribs are placed on their respective halves and clayed up (Fig. 10) as previously explained, thus completing the halves of the core box mold.

Since this core will necessarily require a drier, the plaster cavity (Fig. 11) taken from the drier half of the master core plug is trimmed back or relieved. A plaster is taken from this relieved cavity, the drier frame is placed thereon, and sheet clay of a specified thickness is used to cover all exposed plaster areas within the frame. A plaster half is poured on this assembly as related heretofore, and after drying, is filled with white metal to become a master drier pattern (Fig. 12).

Large master drier patterns obviously should not be cast of white metal because of excessive weight and lack of sufficient rigidity. Sometimes production requirements necessitate the use of a large quantity of driers which may not be produced rapidly or economically with a loose master drier.

Expanding plaster has been developed and is being successfully used to solve both aforementioned problems. By the proper use of these plasters the shrinkage of aluminum can be compensated for, thereby permitting the casting of either single aluminum master drier patterns or match plates for larger quantities.

It is readily apparent to craftsmen engaged in the foundry and pattern fields that the past few years have indicated but the beginning of vast and undreamed of progress. Core blowing, mold blowing, the shell molding process, dielectric core drying and mechanical equipment and devices became a reality. It therefore forces upon those so engaged a challenge to keep informed, that we may keep pace!

# MALLEABLE IRON ANNEALING TIME REDUCED

By

W. G. Wilson and N. F. Tisdale, Jr.*

#### ABSTRACT

The addition of approximately 0.0015 per cent boron to malleable iron has made it possible to reduce the annealing cycle for that iron in comparison to the annealing cycle used for the boron-free iron of the same base analysis. In the first stage of annealing at 1600 F, the cementite in the iron with boron was entirely decomposed in 8 hr whereas it took in excess of 16 hr to decompose the cementite in the base iron.

In the second stage of annealing at 1350 F, the sample with boron showed almost 100 per cent decomposition of the pearlite to ferrite in 8 hr whereas the iron which contained no boron had appreciable quantities of pearlite after 24 hr at the same temperature.

#### Introduction

A large segment of the malleable industry has accepted the fact that boron in amounts ranging from 0.0010 to 0.0020 per cent makes it possible to anneal malleable irons containing as much as 0.12 per cent chromium with little or no difficulty. If there is no boron in the iron, chromium in excess of about 0.035 per cent causes difficulty in the annealing of malleable iron, and when the chromium content of the iron reaches about 0.10 per cent, it becomes uneconomical in many cases to anneal such irons. Therefore, the use of ferroboron has saved producers of malleable iron a great deal of time and money in annealing chromium-bearing irons. In times like these, when scrap which is definitely low in chromium is becoming more difficult to obtain, the use of ferroboron as an insurance against chromium contamination has become more common.

In addition to the ability of boron to enable malleable founders to anneal irons with high chromium content, a number of producers have found that boron in percentages ranging from 0.0010 to 0.0015 per cent promotes the annealability of malleable iron with chromium contents of 0.03 per cent and under. The effect of boron on annealability has been presented in literature on several occasions.

It seems worthwhile at this time, when most malleable founders are pressing for maximum output, to re-emphasize the benefits gained through the use of boron in reducing annealing cycles.

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#### Procedure

The irons used in this project were obtained from a 40-ton air furnace on two successive days. A nominal analysis of these irons was:

Carbon	2.35	pc
Silicon	1.25	",,
Manganese	0.35	99
Sulphur	0.065	22
Phosphorus less than	0.15	9.9
Chromium less than	0.03	99

On each day an analysis bar, a tensile bar and a set of bars 8 in. x 1/2 in. square, 8 in. x 3/4 in. square, and 8 in. x 1 in. square were poured from a ladle of untreated iron. The next 300-lb ladle of iron going to that section of the foundry was treated with 16.2 grams of ferroboron containing 12.61 per cent boron, which was equivalent to putting 0.0015 per cent boron in the iron. The next day this procedure was repeated. As a result there were four sets of bars poured, two sets with boron additions and two sets from the untreated iron.

This paper will report on the annealing of test samples  $\frac{1}{2}$  in. x  $\frac{3}{4}$  in. square, which were cut from the 8 in. x  $\frac{3}{4}$  in. square bars.

The samples, both with and without boron, used in the high temperature annealing work were cast from the same heat of air furnace iron; this eliminates, as far as possible, the effect of variations in carbon, silicon, etc. analysis on annealability. The samples used in the second stage annealing study were all cast from the heat made the next day.

To determine the effect of boron on the first stage of annealing, the samples, both with and without boron, which were cut from the bars cast on the first day, were placed on a tray and put into an electrically-heated muffle-type furnace which was controlled by thermocouples. After 4 hr at 1600 F a sample containing boron and a sample without boron were taken from the furnace and allowed to air cool. After 8 hr another pair of samples were withdrawn from the furnace and air cooled, and at 16 hr another pair were withdrawn and allowed to air cool. The samples were then split on the 3/4 x 3/4 face and

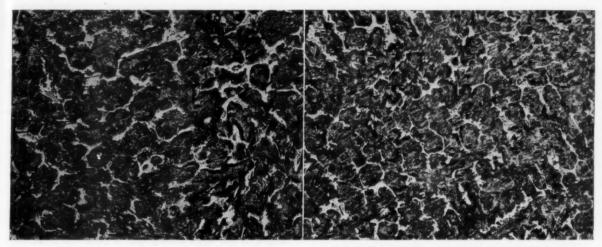


Fig. 1A—Microstructure of white iron which contains no boron. 100x.

Fig. 1B—Microstructure of white iron containing boron. 100x.

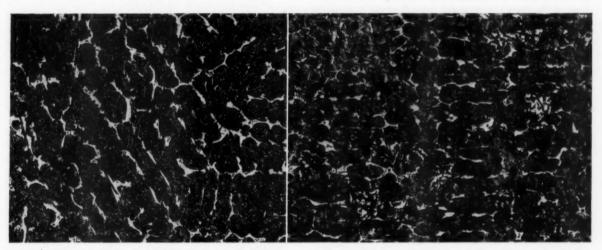


Fig. 2A—Microstructure of malleable iron without boron, annealed 4 hr at 1600 F. 100x.

Fig. 2B—Microstructure of malleable iron containing boron, annealed 4 hr at 1600 F. 100x.

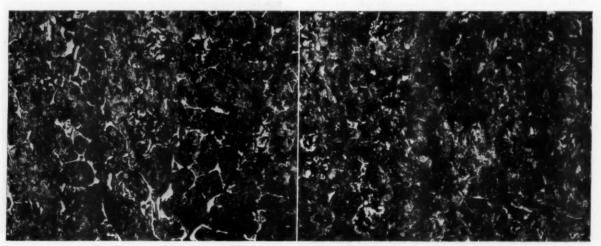


Fig. 3A—Microstructure of boron-free malleable iron, annealed 8 hr at 1600 F. 100x.

Fig. 3B—Microstructure of iron containing boron, annealed 8 hr at 1600 F. 100x.

the  $\frac{1}{2}$  x  $\frac{3}{4}$  face was prepared for metallographic examination. The face used for examination was the one produced by the saw cut.

The samples to be used in the determination of the effect of boron on the second stage of the annealing cycle were cut the same size as those used in the first stage annealing study. However, the 8-in. x 3/4-in. square bars from which these samples were cut were cast on the second day the ferroboron was added to a ladle of iron. Again the samples were placed on a tray in the same muffle-type furnace. After 16 hr at 1650 F, two samples, one containing boron and one without boron, were taken from the furnace and allowed to air cool. The furnace was allowed to drop to 1350 F and 20 min after the furnace reached 1350 F the second pair of samples were withdrawn. Several other pairs of samples were then withdrawn at increasing time intervals.

This second stage annealing experiment had to be repeated because it was found on metallographic polishing that the sample representing the completion of the high temperature anneal of the iron without the boron still contained appreciable amounts of cementite. Upon repetition of this experiment the samples without boron were soaked for 24 hr at 1650 F and those containing boron were held for only 16 hr at 1650 F. At the completion of the 16- and 24hr anneal, a pair of samples, one with boron and the other untreated, were withdrawn from the furnace and allowed to air cool. The furnace was then allowed to drop to 1350 F and pairs of samples were removed from the furnace at successive intervals. Again, the specimens were split on the 3/4 x 3/4 face and the 1/2 x 3/4 face produced by the saw cut on each specimen was prepared for metallographic examination.

# Results

The annealing procedure used to evaluate the effect of boron on the first stage of annealing was given above. By the examination of photomicrographs prepared from the samples with and without boron, which were annealed for various times, as well as a comparison of photomicrographs of the as-cast structure of the boron-treated and untreated irons, it is possible to show the advantages gained through the use of ferroboron in the first stage annealing.

It has been well established that an iron with a small cell size in the as-cast condition and less massive carbide particles will anneal more rapidly than an iron with a large grain size and massive carbides.

A comparison of Fig. 1-A (no boron) with Figure 1-B (with boron) would indicate that the boron-treated iron would anneal much more rapidly than the iron without boron, because of the finer grain size and the slender, well distributed carbide particles. This theory will be substantiated by the other photomicrographs in this series.

Figure 2-A (no boron) and Fig. 2-B (with boron) represent the pair of samples taken from the furnace after 4 hr at 1600 F, and even in this short time the superior annealing qualities of the boron-treated irons are apparent, because:

1. Much greater decomposition of carbides.

2. Greater number of graphite nodules.

3. Larger size of graphite nodules.

From inspection of Fig. 3-A (no boron) and Fig. 3-B (with boron) we may again see the marked effect of boron on the samples annealed 8 hr at 1600 F. The photomicrograph of the iron with boron shows little, if any, cementite in the structure whereas the iron without boron contains appreciable quantities of large heavy carbides.

After annealing the iron without boron for 16 hr it still contained objectionable quantities of cementite. It may be seen in Fig. 4-A (no boron), which is a photomicrograph of the iron annealed for 16 hr, that additional time at 1650 F would be required to produce a structure which would be equivalent to the structure shown in Fig. 3-B, which is a boron-treated iron annealed for 8 hr. Therefore, in this laboratory study, it has been shown that it takes less than half as long to completely anneal boron-treated irons as it does to completely anneal irons without boron.

In the first study of the effect of boron on the low temperature annealing properties, both the treated and untreated samples were held for 16 hr at 1650 F before starting the low temperature portion of the annealing cycle. When these samples were examined, it was found that the iron without boron, after completion of 16 hr at the high temperature, still contained so much cementite that it was impossible to make a comparison of the annealability at the lower temperature.

As mentioned previously, the samples used to study the first stage graphitization were poured on one day and the irons used to study the second stage of the annealing were poured on the succeeding day. Therefore, the fact that the iron produced on the second day without boron did not anneal as rapidly at the high temperature as the iron containing boron further substantiates the work presented above on the effect of boron on the first stage annealing of malleable iron. As a result, in order to have a set of samples in which the cementite had been completely



Fig. 4A—Microstructure of boron-free iron annealed 16 hr at 1600 F. 100x.

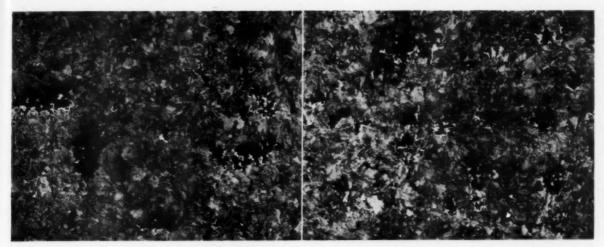


Fig. 5A—Microstructure of malleable iron containing no boron after 24 hr at 1650 F. 100x.

Fig. 5B—Microstructure of malleable iron containing boron after 16 hr at 1650 F. 100x.

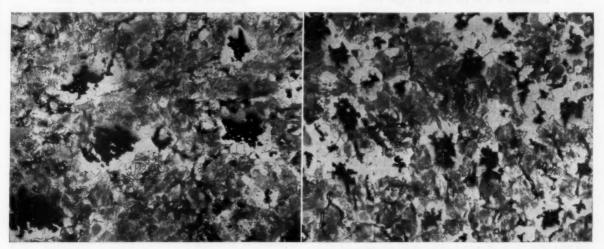


Fig. 6A—Microstructure of iron to which no boron was added after completion of first stage anneal and 1 hr of the second stage. 100x.

Fig. 6B—Microstructure of iron containing boron after completion of first stage anneal and 1 hr of the second stage. 100x.

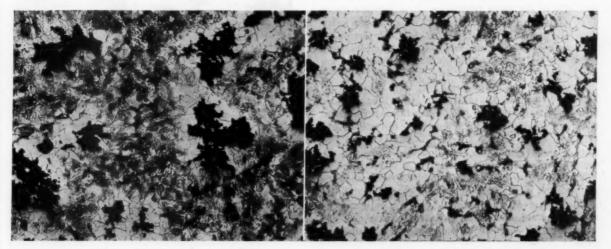


Fig. 7A—Microstructure of boron-free iron after annealing 4 hr in the second stage. 100 x.

Fig. 7B—Microstructure of iron containing boron after annealing 4 hr in the second stage. 100x.

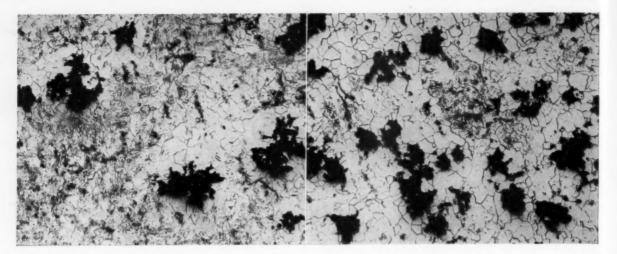


Fig. 8A—Microstructure of iron containing no boron after annealing 8 hr in the second stage. 100x.

Fig. 8B—Microstructure of iron containing boron after annealing 8 hr in the second stage. 100x.

eliminated before starting the second stage of the malleabilizing cycle, it was necessary to hold the ones without boron for 24 hr at 1650 F as compared to a holding time of 16 hr at 1650 F for the ones with the boron.

The photomicrographs representing the samples after completion of the first stage annealing are shown in Fig. 5-A (no boron) and Fig. 5-B (with boron). These photomicrographs show essentially complete elimination of carbides in both the treated and untreated samples. In addition, it may be noticed there are approximately three times as many graphite nodules in the iron containing boron as there are in the iron which was not treated.

Referring back to Fig. 3-B and Fig. 4-A, we can see that, as far as nodule count is concerned, the same condition exists in these samples, poured on the first day, as is shown in Fig. 5-A (no boron) and Fig. 5-B (with boron). Since this increase in nodule count is a qualitative measure of the annealability, we have another indication that the boron acted similarly in the irons poured on these two successive days. The furnace and the remaining samples were then allowed to cool to 1350 F after completion of first stage graphitization and pairs of samples were taken out at 1 hr, 4 hr, 8 hr and 24 hr from the time the furnace first started to drop towards 1350 F.

Figure 6-A (no boron) and Fig. 6-B (with boron) represent the samples taken out after 1 hr. Figure 7-A (no boron) and Fig. 7-B (with boron) represent those taken out after 4 hr. Figure 8-A (no boron) and Fig. 8-B (with boron) represent samples taken out after 8 hr. Figure 9-A (no boron) represents the samples taken out after 24 hr.

A comparison of these photomicrographs shows a pronounced increase in the second stage annealability with the use of ferroboron, which is indicated by the greater amount of ferrite in the structure of the boron-treated iron at all stages of the annealing cycle.

The structure shown in Fig. 8-B (with boron) contains a small quantity of pearlite; however, there is

no more pearlite in Fig. 8-B, which represents a sample annealed only 8 hr, than there is in Fig. 9-A (no boron), which represents a sample that was annealed for 24 hr. It would seem worthwhile to mention that only a portion of the sample containing boron which was annealed for 8 hr at 1350 F showed any pearlite and that approximately 50 per cent of the surface of the sample showed no pearlite whatsoever.

With the photomicrographs presented in this paper, it has been shown that it is possible to complete the low temperature portion of a laboratory annealing cycle of iron with boron in approximately one-third the time required to anneal an iron which contains no boron.

Quite a number of photomicrographs have been shown in this paper which represent irons containing boron. In none of these photomicrographs has there

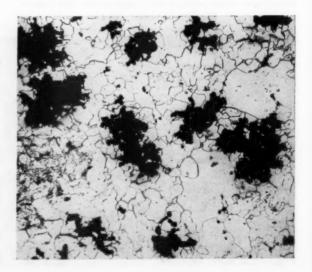


Fig. 9A—Microstructure of boron-free iron, annealed 24 hr in the second stage. 100x.

been any indication of primary graphite. Since primary graphite in any appreciable amount cannot be tolerated in malleable iron, it seems worthwhile to emphasize the absence of this constituent in the boron-treated irons.

# Conclusions

The conclusion drawn from this work is that since boron in this malleable iron has enabled a much more rapid anneal under laboratory conditions, it should be possible to reduce the annealing time of the iron in the large commercial furnaces in this

#### Recommendations

- 1. Boron has been used extensively to combat the harmful effects of chromium on the annealing cycles of malleable iron. This work indicates that the boron when used in iron with low chromium content may be used to accelerate annealing. However, if boron is used to accelerate annealing and then the chromium content of the iron goes up, it would be advisable to increase the length of the annealing cycle to something approaching that used before the introduction of boron into the iron.
- 2. Since the cost of ferroboron additions to malleable iron is only about 16¢ per ton, it would seem logical to recommend the use of ferroboron in almost all malleable foundries operating at the present time for several reasons:
- (a) Scrap with low chromium content is becoming harder to secure.
- (b) The scrap which is obtainable, in many cases, has physical characteristics which are highly undesirable in malleable melting, such as excessive rust and small size, which increases the difficulty of annealing malleable iron produced from such material. An addition of ferroboron should alleviate this prob-
- (c) Even where scrap conditions are good, in many cases the rate of production is so high that the annealing facilities are a bottle-neck in achieving higher production. The use of ferroboron in such instances should materially aid in increasing the production in high strength, high ductility malleable iron.

#### References

H. A. Schwartz, "Boron As An Accelerator Of Malleable Annealing," *The Foundry*, April 1944.
 Norman F. Tisdale, "Boron In Malleable Iron," *The*

Foundry, April 1945.

3. J. E. Miksch, H. A. Fabert and G. M. Cover, "Effect Of Boron Additions On The Malleabilization Of White Cast Irons," AMERICAN FOUNDRYMAN, August 1948, vol. 14, no. 2.

#### Acknowledgments

The writers wish to acknowledge the work of Dr. J. A. Berger, Professor of Metallurgy, University of Pittsburgh, and N. A. DeCecco, Research Metallurgist, Molybdenum Corp. of America, for their cooperation in the preparation of this paper.

#### DISCUSSION

Chairman: W. M. Albrecht, Chain Belt Co., Milwaukee. Co-Chairman: T. F. Ulmer, Link-Belt Co., Indianapolis. Recorder: L. G. OSBORNE, Lakeside Malleable Castings Co., Racine. Wis.

W. D. McMillan (Written Discussion): * The authors of this paper are to be complimented for the concise practical manner in which they have presented the information concerning the use of ferroboron and its effect on annealability of hard iron.

The amount of boron used in percentages of 0.001 to 0.002 is practically the same as that concluded to be an optimum amount by Miksch, Tabert and Cover (AMERICAN FOUNDRYMAN, August 1948). Schneidewind, Reese and Tang (A.F.S. TRANSAC-TIONS, vol. 55), pretty well established that the time required to decompose carbides varied with section size and cooling rate and the basic factor was grain size.

In this paper, Fig. 1B, the grain size is distinctly smaller than in Fig. 1A. The carbide network is also thinner. With a finer cell structure, a greater number of graphite nodules

develop than with a coarser grained metal.

The mechanism by which boron enhances the anneal may be considered mechanical rather than a matter of physical chemistry. Greater rejection of carbon such as effected by silicon in increasing amounts is not a factor. The use of boron, therefore, does not contribute to the formation of primary graphite.

We have been unable to tie in the use of boron with any tendency to give a mottled casting or dark sprue when used in metal cupola air furnace duplex melted and a 40 per cent steel charge. The paper indicates no primary graphite was in evidence in the iron used in these tests which was melted from a charge very likely containing 30 to 40 per cent pig iron.

If there is any greater tendency for a given casting to mottle when poured from iron melted down from a high pig iron charge than from a high steel charge, the iron tested should have been at some disadvantage with respect to mottling.

By outlining the mechanism by which boron enhances the anneal, which is a matter of cell structure and nodule count rather than being the result of a graphitizing carbon rejecting process, the paper should serve to remove the conception that boron in the percentage normally used, 0.001 to 0.002, increases the tendency to mottle.

J. E. REHDER (Written Discussion): 1 The authors have ably presented clear evidence that the addition of very small quantities of ferroboron to a white iron will result in decreased annealing time, and in view of the agreement with data given previously by other authors, it can be accepted that the effect of such additions on annealing rate is as described.

There is a minor slip in the paper where it is stated that with ferroboron additions the iron has both more and larger temper carbon nodules. Obviously for the same total carbon content the nodules cannot be both larger and more numerous.

Additional data, if available, would be appreciated. Particularly the writer would like to know the analysis of the boron addition agent used, especially its aluminum content; the actual boron content by analysis of the treated iron; and the mechanical properties of the fully-annealed irons with and without ferroboron additions.

Although it is agreed that ladle additions of ferroboron will decrease the annealing time of a white iron under suitable conditions, there are other considerations which have not been mentioned in the paper and which are of major importance to the practical foundryman. The principal of these are control and the effect of too high a boron content of the iron. It is well known that when the boron content of a white iron approaches about 0.005 per cent or more, the powerful carbide stabilizing effect of boron becomes dominant, and boron is then a more serious retardant of anneal than is chromium. It is questionable whether all of the boron in remelt is lost in the melting process, and so if boron additions are used on a routine basis, regular analysis for boron content must be carried out to ensure that the medicine does not become worse than the disease. Analysis for boron content of these small orders in iron is not a simple procedure, and it will be found that accurate analyses for boron content are costly and slow.

For these reasons the present writer considers that the use of ferroboron additions must be approached with caution even in foundries with large and well-equipped laboratories. If the money and time necessary to obtain adequate control over boron content were to be spent instead on better over-all metallurgical

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1 Foundry Engineer, Bureau of Mines & Technical Surveys, Ottawa, Ont.,

control and on control of annealing practice, possibly equal benefits in time-saving and in decreased cost of finished product would result.

To summarize, the addition of ferroboron to white iron undoubtedly has the effects on annealing rate described, but a small excess of boron content in the iron works strongly in the opposite direction and control is not believed to be simple, so that malleable foundrymen should proceed with caution in the use of ferroboron additions.

MESSRS. WILSON AND TISDALE (Reply to Mr. Rehder): We would like to take the opportunity to reply to the first paragraph of Mr. Rehder's discussion. We believe that this paper gives, for the first time, some quantitative data on the effect of ferroboron in malleable iron. The other papers to which we referred mention that ferroboron accelerates annealing, but to date this is the first attempt made, that we know of, to show how much effect boron has upon annealing time.

In reply to Mr. Rehder's question on analysis of the boron agent used in this work we are pleased to inform him that it was an alloy containing 10.00 to 14.00 per cent boron, 4.00 per cent max silicon, 2.00 per cent max carbon, and little if any aluminum. This form of ferroboron is manufactured by a carbon silicon reduction process rather than a thermite process so that the aluminum will always be considerably below 0.5 per cent. We did not determine the boron content of the treated iron nor the mechanical properties of the fully annealed iron.

We fully realize that overdoses of ferroboron are quite harmful. As a result, the Molybdenum Corp. of America has always made boron determinations for the malleable foundries that used their products, to enable them to keep in control. By making determinations as frequently as once or twice a month, many foundries have been able to use ferroboron satisfactorily for many years.

As a general rule we find that roughly two-thirds of the boron contained in the sprues is lost on remelting. Many foundries who have accepted this rule of thumb have found that it will work satisfactorily.

We know of no one that has gotten enough boron in their malleable iron to get to the point where they have started to stabilize the carbides. Generally when they get very much in excess of 0.0015 per cent, the amount recommended in this paper, they find that their physical properties tend to drop off. As a result, this gives them adequate warning to reduce the amount of their boron addition before any carbide stabilization occurs. The large number of foundries using ferroboron in malleable iron today attest to its adaptability in the malleable industry.

F. J. WURSCHER (Written Discussion): 2 The usage of boron as an antidote for chromium-contaminated white iron is an expedient for salvaging a defective melt.

Advocating the introduction of boron to the base metal as a standard procedure by pointing to its beneficial effect upon reduction of annealing time should be confined to the production of Grade B specification malleable iron owing to the fact that boron affects adversely the physical properties of the ultimate metal, specifically that of ductility as measured by per cent elongation. The reduction of that property is less critical for Grade B irons than for those of Grade A specification.

The photomicrographs of boron-iron show a somewhat smaller grain size than is displayed by the base-iron, indicating that the tensile properties of the two metals will not differ materially.

The presence of a numerically greater amount of tempercarbon nodules in the B-iron attests to its lower ductility value than is to be expected in the untreated, uncontaminated, base iron because this larger number of temper-carbon nodules break-up more effectively the continuity of the ductile ferrite of the matrix.

Conversely, the B-iron, by reason of the increased nodule count, can be expected to machine more readily than the base metal. Relative to the structure of the temper-carbon, both irons show the same lacy condition.

Resorting to boron as cure-all for unrestricted acceptance of raw materials, specifically that of steel scrap, may lead to failure of the ultimate castings in service. The economy of its use versus the cost of selected steel scrap is to be questioned.

MESSRS. WILSON AND TISDALE (Reply to Mr. Wurscher): Ferroboron has been used in a number of shops producing malleable iron using Grade A specification for some time. When the boron content of these irons is maintained between 0.0010 and 0.0015 per cent they find that the physical properties remain essentially the same as those that may be obtained without the use of boron. Naturally our experience with the Grade A malleable is not as extensive as with Grade B, because of the smaller tonnage of Grade A malleable produced in this country.

There seems to be some difference of opinion among the experts on a malleable iron as to the effect of the size of the temper carbon nodules on the ductility of the annealed malleable. One foundryman who has had extensive experience in malleable iron and has done considerable research on this product, was of the opinion that a greater number of temper carbon nodules enhanced the ductility which might be obtained with malleable iron.

We certainly agree that the better distribution of graphite in the boron treated iron should increase the machinability of these

At the present time there are malleable foundries accepting steel scrap of unknown origin and using this scrap in conjunction with ferroboron to produce malleable iron castings to both Grade A and Grade B specifications. We have seen some of these scrap piles in our travels and are sure that some of the more conservative malleable foundries would say that no acceptable iron could be made with such raw materials. However, acceptable castings are made with these materials, and in appreciable quantity.

W. D. McMillan: Ferroboron is added to the metal leaving the cupola. We use 175 grams of 10 per cent ferroboron every 20 min when melting about 10 tons per hour. This is equivalent to 0.0006 per cent boron. Tests for boron on the castings show 0.001 to 0.0015 per cent boron.

MESSRS, WILSON AND TISDALE: The matter of mottled structure in white cast iron is a rather elusive subject. As mentioned in the paper in the paragraph immediately preceding the conclusions, there is no indication of mottled iron in any of the photomicrographs presented herein, which presents some evidence that ferroboron has little or no effect upon mottling. The most conclusive evidence of the effect of ferroboron on the occurrence of primary graphite is found in the rather large list of people who have been using ferroboron in their malleable iron for quite some time. If there was any tendency at all to produce mottle when using ferroboron, it would seem that these people would have noticed that tendency and because mottle has such a highly deleterious effect on the properties of malleable iron they would have probably abandoned its use years ago. Therefore, this acceptance of ferroboron is perhaps the most convincing evidence that it has no effect on mottling.

MR. REHDER: Boron will decrease the required annealing time, if chromium enters the iron; however, you can still anneal with a longer cycle. Boron may also help annealing due to other causes

One should not refer to a specific level of chromium as the boron effect will also depend upon other properties of a specific plants metal. Increase in the superheat of iron will give a similar beneficial effect.

MR. WILSON: Mr. Rehder's comments are true. However, with the continued use of boron you do not run into the problem of making hard iron for an hour or so, when the analysis is temporarily out of line.

W. G. ETTINGER: 3 How do you add boron to the molten metal?

MR. TISDALE: You can make the addition by envelope to the bottom of the ladle or while tapping. Plunging is not too practical.

Mr. Wilson: In duplex operation, ladle addition is best or a syntron feeder between the ladle and the cupola may be used. In a straight air furnace operation, addition should be made to the furnace after the last skim preceding the tapping.

FRANK CZAPSKI: 4 We use a volume addition into the ladle.

MEMBER: We do not find that boron builds up.

MR. TISDALE: There is a general build-up or recovery of 20-30

ERIC WELANDER; 5 We find that a spectrographic analysis for boron gives inaccurate results.

³ Metallurgist, The Minster Machine Co., Minster, Ohio.

Metallurgist, American Brake Shoe Co., Mahwah, N. J.
 Chief Metallurgist, Chicago Malleable Castings Co., Chicago.
 Metallurgist, John Deere Harvester Works, East Moline, Ill.

# PATTERNMAKING TO COMPENSATE FOR SHRINK ALLOWANCE

By

M. K. Young*

#### Introduction

Use of high expansion plasters for fabrication of expanded patterns, models and molds is an outstanding tooling development of the last five years. Industrial tooling and patternmaking materials such as high-expansion and medium-high expansion gypsum cements have many applications in the aircraft, automotive and foundry industries. Use of expanding plasters has facilitated patternmaking to such a degree that many pattern shops have adopted these materials as standard shop procedure.

Medium-high and high expansion gypsum cements have unusual characteristics. They expand uniformly in all directions and have the highest setting expansion of any known gypsum cements. The amount of expansion can be controlled by the quantity of water used in the mix. Through the proper water to plaster ratio, it is possible to achieve expansion values from

1/16 in. to 1/4 in. per ft.

The end use or application for the expansion plasters are unlimited wherever patterns, matchplates and models are fabricated with shrink allowance. A pattern shop in Chicago utilized the expansion plasters for core dryer patterns by making a cast or impression from the pattern. After the cast has expanded to the desired oversize dimensions, subsequent patternmaking operations were accomplished to complete the core dryer pattern. In Milwaukee a matchplate firm made use of the expansion plasters to add the extra shrink that is necessary where a wood matchplate is to be reproduced in aluminum by the pressure-cast technique.

In Detroit the medium-expansion and high expansion gypsum cements have been used to convert conventional foundry patterns to the metal equipment required for the shell molding or Croning process. This has been accomplished by pouring a mix of the expansion plaster over the existing pattern or foundry equipment. After the plaster has set, it is removed from the pattern or core box and allowed to expand to the dimensions required for the alloy

that is to be cast. The expanded pattern is then rammed in sand to produce the desired metal equipment.

An interesting application of the high-expansion gypsum cement was achieved at one of the large pattern shops in Detroit. An automotive firm approached this shop with a request to reproduce three wood patterns that they had used in the development of a new type engine. The development being successful inspired the automotive engineers to investigate the Croning process as a possible means of fabrication for the production of the new engine. Because of the deadline for the completion of the program, the pattern shop was furnished with the original wood patterns and requested to complete the work in one week. The wood patterns that originally required four weeks to complete were used as master models. Through the use of expanding plasters the new gypsum cement patterns with a double shrink were completed in 36 hr.

While these materials are finding new uses in the automotive and foundry industries, several large aircraft manufacturers have centered their kirksite tooling programs around the expansion plasters because of accuracy, expediency and economy.

### Material Characteristics and Properties

High expansion gypsum cement is a general purpose expansion plaster used in patternmaking to compensate for shrinkage in kirksite, aluminum and bronze. Since the material will expand as much as 1/4 in. per ft it is recommended for the higher shrinking alloys. This product is not highly stable after the cast has expanded to the desired dimensions. Therefore, the expanded pattern or cast should be used as soon as possible. Patterns or casts made from high expansion gypsum cement that have been left to stand over night have been found to be slightly oversize. This is characteristic of any highly expanding material and provision should be made to use the pattern immediately upon attainment of the desired size.

The expansion properties of high expansion gypsum cement are controlled by the water to plaster

^{*} United States Gypsum Co., Chicago.

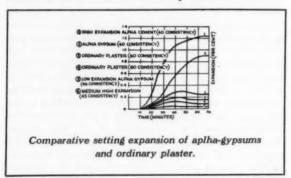
ratio, and therefore the water and the plaster must be accurately proportioned to obtain the desired degree of expansion. Users of expansion plasters have found that they must weigh the plaster and water for each mix, as shown in Table 1.

TABLE 1

High Expansion Gypsum Cement (By Weight)	Parts Water (By Weight)	Approximate Maximum Expansion (Inches per foot)	
100	50-52	1/8	
100	44-46	3/16	
100	38-42	1/4	

Casts must be 3 to 4 in. minimum thickness to obtain these values.

Higher expansion values can be achieved through the use of hot water. Water (120 to 130 F) will increase the expansion approximately 10 to 15 per cent over the values shown on the expansion chart.



Medium-high expansion gypsum cement is an expansion plaster designed for zinc-base alloys having 1/8 in. per ft shrinkage. The material is characterized by higher strength, uniform and consistent expansion and dimensional stability after the desired expansion is achieved. The expanded cast is strong enough after setting to be used as the foundry pattern.

It is also necessary, when using medium-high expansion gypsum cement that the water to plaster ratio be carefully controlled. The best results can be obtained by weighing the plaster and the water for each mix.

TABLE 2-RECOMMENDED PROPORTIONS

Medium-High Expansion Gypsum Cement (By Weight)	Parts Water (By Weight)	Approximate Maximum Expansion (Inches per foot)	
100	50-52	1/16	
100	45-47	3/32	
100	40-42	1/8	

Casts must be 4 to 5 in. minimum in thickness to obtain these values. These expansion values can be increased 10 to 15 per cent by using hot water (120–130 F).

## Mixing the Cement

After the expansion plaster and room temperature water have been weighed as recommended, the plaster is strewn into the water as evenly as possible and left to soak 1½ to 2 min. When the lower water ratios

are used a heavier mix will result and there will not be enough water to completely soak the material. Therefore, mechanical mixing is recommended for the heavier mixes to insure a lump-free slurry.

When mixing by hand it is necessary to force the material into the water and mix vigorously until a smooth mass is obtained. For mixes from 10 to 50 lb a 1/4-in. drill motor with a 4-in. disc has been found superior to hand mixing and should be used to achieve optimum properties.

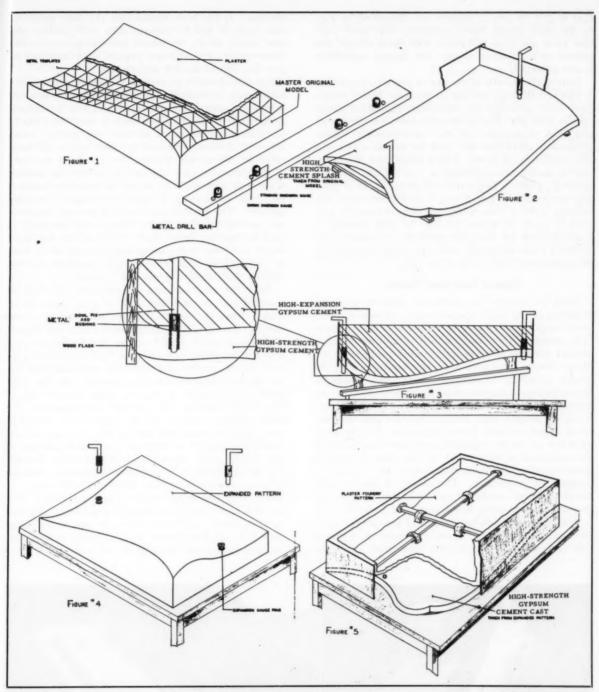
# Removal of Casts from Patterns

Removal of the expansion plaster from rigid patterns has been facilitated when the patterns are sealed with one or two coats of a good grade of lacquer. This is followed by a coating of plaster parting compound such as stearic acid and kerosene that will not alter the detail or soften the face of the plaster cast. Upon removal of the plaster from the pattern, the cast should be placed on a flat surface that has been lubricated to permit free growth of the expanding medium. The surface plate or table top on which the cast is placed should be such that it will not have a chilling effect on the warm plaster during expansion. It has been found that a cold surface plate (below room temperature) will retard growth of the cast. By the same token cold water (below room temperature) will also retard expansion. Consequently, the pattern will not grow to the desired shrink allowance. The water temperature should be approximately 70 to 75 F for optimum results. Care should be exercised to make sure that intricate-shaped patterns are properly supported when placed on the surface plate or table to prevent possible warpage during and after the expansion period.

#### Methods and Techniques for Expanding Patterns

One of the large aircraft manufacturers utilized high expansion plaster for a number of years in his drop hammer tooling program. The experience gained over the period of years is now reflected in their low over-all tooling time for casting kirksite dies. The time and effort put forth by their supervisory staff resulted in time savings of 40 to 65 per cent over the traditional method of pattern shrink allowance and die finishing for their kirksite drop hammer dies.

This company developed a process for determining the expansion of the plaster cast as it expands to the desired oversize dimensions. This technique is a supplement to the standard aircraft tooling procedure since the mock-up or template setup is constructed in the usual manner by filling in the templates with a high-strength gypsum cement. As the gypsum cement hardens, the surface is splined or screeded to form a smooth, accurate full scale model as illustrated by Fig. 1. The trim lines of the part are then scribed in the plaster and the entire surface given two coats of a good grade of clear lacquer. When the lacquer has dried the surface is coated with a parting compound such as stearic acid and kerosene, and then a splash cast or impression is made over the master original model. The cast is usually made with high strength gypsum cement which is approximately 11/2



in. to 2 in, in thickness and reinforced with sisal fibre, and a wood frame as illustrated by Fig. 2. The metal drill bar is then placed on the splash cast and two holes are drilled through the standard dimension gauge bushings. These holes will be transferred to the expansion plaster cast and then used to measure the growth of the cast. Dowel pins and bushings are placed in the drilled holes, and sides are built around the splash cast as illustrated in Fig. 2. The surface of the splash cast and the flask are given the usual coating of lacquer and stearic acid before pouring the expansion plaster.

The high expansion or the medium high expansion gypsum cement is then mixed as outlined in Table 1 or 2. The mix is poured as shown in Fig. 3. The initial set occurs shortly after the material loses its water gloss. After this period has been reached, close attention should be given to the progress of the setting action. The cast will become warm when the set occurs. As soon as this heat can be felt, the dowel pins should be removed and the expanding cast separated from the high strength gypsum cement splash. The expanding cast should be placed on a flat table that has been greased so

that it will be free to expand, as illustrated by Fig. 4. We have found from experience that table tops that have not been lubricated will often restrict the growth of the pattern so that the desired expansion is not achieved.

During the growth of the cast or pattern, expansion is checked through the use of trammel points which have been preset to the shrinkage dimension gauge on the drill bar. Use of the drill bar with the standard drill dimension and the corresponding shrink dimension eliminates the need for shrink scales and the possibility of error. The 0.135-in. per ft expansion is usually accomplished in 2 to 21/2 hr. The expanded pattern is lacquered and given a coat of parting compound. A splash cast is then taken off of the oversize pattern from which the plaster foundry pattern is made, as shown in Fig. 5. This pattern is then checked for size and given the customary two or three coats of lacquer and sent to the foundry, to be reproduced into a zinc-base alloy.

# Gypsum Sand Mold Process

Field evaluation has indicated pattern shop acceptance of the high-expansion and medium-high expansion gypsum cements in all parts of the country. The removal of the expansion plaster from intricate shaped patterns however, presented a problem in the shops where employees were not familiar with the proper handling techniques for the two materials. Immediately, a program was initiated to alleviate the parting difficulties that were experienced on the more complex designs with thin-walled ribs and intersections found on many foundry patterns. It is believed that the use of the expanding gypsum cements would be simplified if they could be cast into a mold or patternmaking medium that would not restrict or confine the expansion. The result of the development program concluded that sand molds were the answer to expanding intricate-shaped patterns.

Use of sand molds has facilitated the use of expansion plasters to such a degree that patterns of almost any design can be successfully made with shrink al-

lowance. It has been found that the finer grade of sand such as No. 00 Albany sand will produce the best surface detail. Expanded patterns poured in the coarse grade of sand often require some finishing to give the desired surface smoothness.

Standard foundry practice is applied to produce the sand molds in which the expansion plaster is poured. When the sand is rammed it should not be tamped as hard as the sand mold which is used for kirksite or other metal alloys. A standard size pattern exact in every dimension should be used as it will later serve as the master pattern for checking accuracy of the expanded cast. Before the face of the pattern is covered with sand, it should be powdered with a fine coat of parting as used in standard foundry practice. The parting can be pearl parting, lacopodium or walnut shell flour. When very fine sand is used no parting is necessary. Lacopodium has been found to be the best parting medium and when used in conjunction with coarse sand, produces an expanded pattern with a smoother surface.

After the sand molding is completed the flask is inverted and the standard scale pattern extracted or drawn from the sand. Caution must be exercised during removal of the pattern so that it is not rapped or bumped which would enlarge the sand mold cavity. Medium-high expansion or high expansion gypsum cement gauged and mixed as recommended for the desired expansion is then poured into the sand mold as illustrated in Fig. 6. Close observation of Fig. 6 will disclose the parting compound on the main channel section of the sand mold. Caution must also be exercised during the pouring of the expansion plaster so that the sand mold surface is not washed away. Sand wash can be avoided through the use of a baffle, a down gate, or by pouring through a spout which is 6 to 8 in. above the sand mold surface. Once the mold surface is covered with 2 to 3 in. of plaster there is no danger of washing. A heavy mix of plaster will result with the low water ratios and will not flow freely, thus, it must be forced or tamped into place. Because of the heavy mix, some



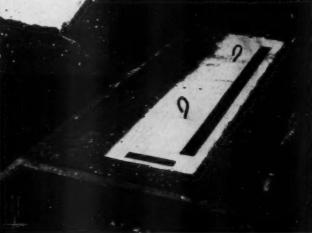


Fig. 6

Fig. 7

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voids may occur in the face of the expansion plaster cast. The voids can be filled after the cast has reached its maximum expansion. The sand mold should be filled completely with plaster so that this thickness is at least 4 to 5 in. See Table 2.

When the expansion plaster begins to cream or become plastic, eye-bolts should be inserted in the cast as shown in Fig. 4. The bolts will facilitate handling of the plaster pattern in subsequent molding operations. During this creamy stage, it is also advisable to imbed two standard steel scales in the plaster cast as shown in Fig. 7. As the plaster sets and expands, impressions made by the steel scales will grow progressively larger to indicate or show the amount of expansion that has taken place. It is advisable to leave the expanding cast in the sand until the desired expansion has taken place and the plaster has cooled to room temperature. Upon shake-out a thin layer of sand will adhere to the surface of the expanded pattern and should be removed through the use of water and a soft brush or cloth. The pattern

is checked for size and then given the customary two or three coats of lacquer before being rammed up in sand to make the sand mold in which the kirksite is poured.

#### Summary

Application of medium-high and high expansion gypsum cements, their characteristics and methods of handling have been set forth to illustrate the versatility of the materials in patternmaking for shrink allowance. Utilization of these materials will afford accuracy, expediency and economy where the traditional equipment is being converted for the shell molding or Croning process. This and other patternmaking applications have been made possible through the use of expanding plasters. The material and techniques have been tried and proven, and will continue to facilitate patternmaking as the patternmaker combines his ingenuity and skill in the use of high-expansion and medium high expansion gypsum cements.

# PATTERN APPRENTICE TRAINING PROGRAM

By

# F. W. Burgdorfer*

Those of us who have been associated with the foundry industry for the past few decades or so have witnessed many changes, particularly within the pattern division of this industry. There is no doubt that if the foundry had shown less ingenuity in meeting customers' demands in past years, a larger portion of work would have been diverted to die fabricating, welding, etc., than was the case.

It is well to remember that nothing in this life is self-sustaining. The very existence of our customers' business depends upon how well they can meet competition. If they fail by using our products, then it is our entire industry that suffers, for more than likely they will turn to other methods of production. With a little more ingenuity and a lot more determination we can prevent further loss of ground and possibly regain some already lost. How? Through proper apprenticeship training.

# Support Apprentice Training

Apprenticeship training discussions are not new within the A.F.S. The first discussion on record took place at the first A.F.S. meeting in Philadelphia in 1896, and it can well be said that the Apprentice Training Committee is one of the oldest functioning committees of A.F.S. However, support of apprenticeship from outside the committee cannot boast of such a record.

A tremendous factor overlooked by management today is that apprenticeship in this country is big business—one of the biggest. Today there are two main types of training in existence. One is the form of training students receive in the schools; the other is the training of apprentices who work for a wage received on the job. However, a vast difference exists between these two types of training. In the schools students are thoroughly screened before they are admitted. They progress step by step through a planned course of study until they acquire the necessary knowledge.

The apprentices, in most cases, are hired without proper screening and spend 4 or 5 years working in a shop without a definite training program. As a result,

at the end of 4 or 5 years we have men on our hands who are classed as journeymen but who cannot fulfill our or our customers' demands. What are we going to do? Produce patterns and castings that lack accuracy, cost more—lose our customers? Or is it our job to eliminate this condition by sound apprenticeship selection and training?

Although management has recognized the importance of acquiring and training a competent supervisory force, for a long time it has overlooked a very vital item—apprentice training. Management is to be commended for its fine work among university and college groups of young people in bringing the industry before them, but unless it includes apprentice training it is only doing a halfway job, since supervisory personnel being trained in universities and colleges can only become a small part of the industry's personnel. If we are to meet outside competition, skilled tradesmen must always outnumber technical experts.

It is impractical to fill the shops with enough technical experts to personally supervise each journeyman. From a purely business point of view it is only sound logic that management should try to get along with as few of these people as possible. Each time a technical expert or such has to stop his work to explain a job or method, management is losing production. By properly training apprentices, journeymen are produced who will require less supervision and who, in a limited way, will be management hired technical experts at a far lower scale.

#### Schools Are Source of Supply

Many of the problems confronting the foundry industry today are the same that confronted it 50 years ago; namely, the finding of young men who are willing to invest a few years of their lives in acquiring a worthwhile trade. Our source of apprentices at present is from the elementary and high schools but, unfortunately, little effort has been made to date to bring our industry to their attention. This is not so with the professional and white-collar people. As a result, we are gradually losing our source of supply.

One can ask a group of 25 or even 75 students, "What is patternmaking and why is it a desirable

^{*} President, Missouri Pattern Works, St. Louis.

trade?" If one gets an answer at all, it no doubt will pertain to dress patternmaking and not the foundry. The only sound answer to this problem is the formulation of joint area-wide apprentice training programs in every area in which the industry exists, by establishing a joint management-labor committee to study industry needs and drafting a program that will supply those needs. For an example, perhaps the problem in one area is that the ratio between apprentices and journeymen is too high. As a representative of management, the chances of an individual in getting a Union to change its ratio rules are very slim, but if the problem is handled from a joint management-labor committee standpoint, the chances are very good



Thomas Ross, Instructor at Rankin Trade School, St. Louis, about to discuss with students the wood patternmaking project in the 1949 A.F.S. Apprentice Contest.

that the condition can be corrected. How? Because we will have the help of the Labor members of the Committee to do the job.

This particular problem was encountered in St. Louis in the metal division of the industry. For years individual employers had tried to get additional metal apprentices, without success. But once the Joint Committee tackled the problem we had all the apprentices we could handle, for both wood and metal.

With the establishment of a committee to supervise established apprenticeship training programs, apprentices can be properly screened before entering the trade; the Committee can examine the apprentices at various times during their terms of apprenticeship and if it is found that they are lacking what it takes to become journeymen, they can be weeded out long before they are classed as such. The committee can recommend apprenticeship training changes that will benefit the apprentice and at the same time supply management with competent journeymen.

The committee can stage "Certificate of Completion Ceremonies" that will bring before the public the activities and work of the industry, particularly to the group of elementary and high school students we desire as future apprentices. Since both management and labor take an equal part in such completion ceremonies, newspaper coverage is readily available. It is this type of publicity that brings the industry before the parents of young men and governs the magnitude of influence which they will exert over their sons to enter the industry.

# Develop Modern Training Methods

In the early 1940's the St. Louis Pattern Manufacturers Association met with the U. S. Department of Labor Bureau of Apprenticeship and requested assistance in banishing the antiquated pick-and-shovel technique of apprenticeship and developing a modern bulldozer method to meet the changing demands of the industry in the area. With this assistance, a joint management-labor apprenticeship committee was formed and named the St. Louis and Vicinity Joint Apprenticeship and Training Committee.

From the start, the committee was mindful of the fact that basic and paramount in the complex relations of modern patternmaking is the skill with which its journeymen can bring to life, in three-dimensional models, live perspective drawings in the least time and with the greatest accuracy man can obtain.

The scope of the committee's activities since its formation includes the screening of apprenticeship applicants; supervision of related training courses conducted in vocational schools during off-work hours; giving of periodic examinations to apprentices; and making training recommendations to employers.

In order that the apprentices in the St. Louis area might be afforded the opportunity of furthering their knowledge of patternmaking, the Joint Apprenticeship Committee purchased and maintains a free lending library covering the fundamental principles of all types of molding, coremaking, and the use of molding machinery. Included in this library are books on wood, metal and plaster patternmaking, as well as books on mathematical principles and fundamentals. We have been informed by the Bureau of Apprenticeship that this is the only apprenticeship committee having a program registered with them that personally owns such an apprentice lending library.

#### Program Makes Journeymen

Many of the problems which existed in the beginning took both the cooperation and patience of labor and management to work out satisfactorily. However, not only did the committee stay with the basic problems and iron them out, but it continued to meet periodically until it has developed an established apprenticeship program which makes journeymen in every sense of the word. However, even though we have come a long way, we are reluctant to even now slacken our efforts to improve our program.

At the present time, the committee is endeavoring to institute a plan whereby the patternmaking apprentices in the area will be able to rotate from one employer to another for 6-month periods after they have completed their third year as apprentices. Most pattern job shops specialize in one class of work. This being the case, it is easy to see why a large percentage

of present job-shop apprentices become journeymen with a very limited knowledge of the industry.

The effect of the apprentice's lack of all-around knowledge has been made very apparent through our system of giving each apprentice a periodic examination every 6 months. We find that many of the apprentices employed in job shops cannot tell how cast aluminum match plates are made, how plate thickness is obtained, or how a squeezer or any other molding machine works. The seriousness of this lack of allaround trade knowledge is shown whenever a job-shop employer hires a journeyman who has only worked in a commercial shop, or when a commercial shop hires a job-shop journeyman. The usual result is that someone has to spend the time showing the man foundry practices in order that he may perform some simple operation-an operation he should have learned as an apprentice.

Training Program Varies With Locality

Since each area in this country is confronted with

different problems as far as apprenticeship training is concerned, no one finished program can be adopted. The complete program followed in St. Louis might be ideal for an area in Michigan, but might be impractical for an area in Pennsylvania or New York. However, we believe that a basic program such as the one we started with should be adopted and used as a blue-print or, using a term of the trade, "bottom board," from which practical, sound apprenticeship programs can be built in individual areas. Thus, each area, although it will start with a basic program, will be able to tailor-make a finished program to fit the needs.

Another policy management should adopt is a program in which each apprentice is placed on a 6-month probationary period before final selection as an apprentice is made. At the end of this period we believe every apprentice should be required to read and answer the questions following the chapter in the article by Paul A. Barrett in *Readings and Trade Literature* 

entitled "Patternmaking."

# HOW TO MAINTAIN FOUNDRY VENTILATION AND DUST COLLECTING SYSTEMS

By

Kenneth M. Smith*

Have you ever been troubled by production equipment shutdowns caused by the failure of dust collecting or ventilating systems? If you have ever had such troubles, you are well aware of how costly such interruptions can be! An adequate preventive maintenance program built upon the design features outlined in the following discussion will pay for itself by eliminating a large part of the losses incurred when production workers and equipment cannot be properly utilized while essential service equipment is being repaired.

Effective maintenance of foundry ventilation and dust collecting systems consists quite simply of three things. First, design rugged systems with the realization that many parts of the systems may require routine cleaning and eventual replacement. Second, install all parts of the systems with the thought "Be sure this part can easily and quickly be replaced or cleaned." Third, establish a routine procedure for the observation and care of all existing systems by competent persons.

The first reaction to these requirements frequently is "What about systems already built which do not fit these requirements?" The best answer to this is "Follow the above steps for normal operation and in planning for or doing any future maintenance." If given the proper encouragement, competent tinners, millwrights and electricians can be of valuable assistance in determining the necessary changes and additions for simplifying and speeding up maintenance of the system parts.

An important rule in system design for good maintenance is "Always provide enough room for the convenient installation, cleaning, or removal of any system part whether it is a hood, duct, dust collector part, fan, or motor." For example, the intentional permanent installation of a simple monorail beam or bridge crane will speed up both the original installation and maintenance work on the fan and motor of any system. Since this hoisting equipment will ordinarily be used at infrequent intervals, it is more economical to have one chain hoist available for use

with a number of bridge cranes or monorails, but a hoist dolly should be permanently installed in each case.

Adequate service catwalks and ladders should be installed for all equipment to provide worker safety, convenient operation, and convenient maintenance.

The first step in the installation of any ventilation or dust collecting system is the making of the necessary completely dimensioned drawings. Since the majority of ventilation and dust collecting systems are built from simple schematic drawings or verbal instructions, too few dimensional records exist for effective use in maintenance work. However, the tinners and millwrights of necessity make a variety of dimensional sketches on the job in order to construct the parts. Rarely is any attempt made to save and properly identify these sketches. Why not supply the craftsmen with tracing paper sketch pads* and encourage them to use them for their fabrication sketches! The set of grid sheets which come with these pads will help the craftsmen make more useful drawings in less time, and the drawings can be blueprinted for future maintenance use. Since these sketches will usually be of business letter size, they are easy to handle at the job where final changes and details can easily be recorded. For those companies which make detailed engineering drawings of system parts the sketch pad system is a convenient one for sending on-the-job changes back to the engineering department.

When a system installation is completed, copies of these construction sketches should be made up into a booklet which should also include the complete name-plate data of all equipment, lists and spare parts reference drawings for all equipment, and actual system operating data such as air inlet or outlet velocities and volumes, suction and pressure readings at various points in the system, fan speed, and motor ampere loadings. Each booklet should contain an index of its contents with a complete description of all drawings including those which may be too large to fasten into

^{*} Foundry Engineer, Caterpillar Tractor Co., Peoria, Ill.

[•] Typical sketch pad examples are:
Dietzgen No. 374W, 8½ in. x 11 in. size; No. 373B, 11 in. x 14 in. size; No. 373T, 14 in. x 17 in. size; Post No. 173A, 8½ x 11 in. size; No. 173C, 12 in. x 18 in. size.



Fig. 1—Multiblade squirrel cage centrifugal fan impeller.

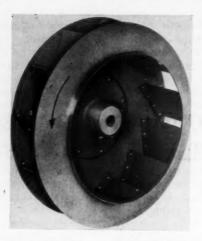


Fig. 2—Backward inclined wide blade non-overloading centrifugal fan impeller.

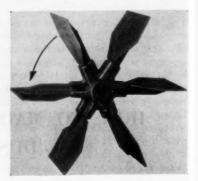


Fig. 3—Radial blade paddle wheel centrifugal fan impeller.

the booklet. Many system records have been crippled by the loss of improperly recorded drawings.

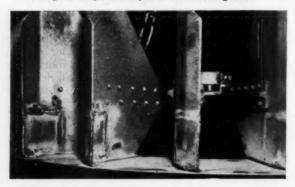
A master file of all tracings, parts lists, and related drawings should be kept in such a place that they will not be lost by maintenance shop use.

#### Fans

Most fan troubles result from four basic causes—(1) dirt build up on impeller or housing, (2) abrasive cutting of impeller and housing, (3) too much heat transmitted to bearings in the case of high temperature systems for core ovens or cupola hot blasts, and (4) indifferent care of fan drive and bearings. While vibration may also seem to be a primary cause of fan failure, the vibration itself is the result of one of the above four basic causes unless the fan was damaged when installed.

Prevention of fan troubles logically follows the pattern of the four basic troubles. First, study the source of any material which may build up on fan surfaces. Can any changes be made in the system to reduce the solid material carried through the fan? Next, consider the type of fan in relation to the type of system. Centrifugal fans with multi-blade squirrel cage impellers (Fig. 1) are not suitable for foundry exhaust systems. The narrow curved blades collect dirt too readily for reliable system operation. Dirt

Fig. 4-Repaired impeller for centrifugal fan.

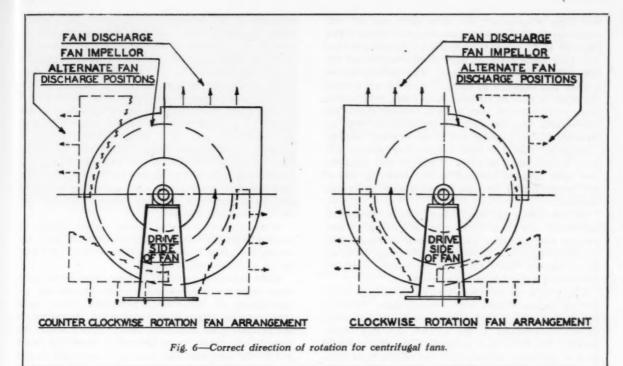


build up will not be uniform on all blades, and the fan will soon vibrate. These fans should be replaced by backward inclined wide blade non-overloading centrifugal fans (Fig. 2) or by radial blade paddle wheel centrifugal fans (Fig. 3). For the same volume of air handled the paddle wheel fan will require more power. Consequently, it tends to be used primarily in systems handling highly abrasive gases, in core oven circulating systems where an accidental fire will not damage the fan, and in pattern shop shavings conveyor systems.

As a means of encouraging more effective removal of any dirt build up in any kind of fan, adequate service doors should be built into the fan itself or into the duct system on both sides of the fan. On centrifugal fans a large section of the scroll sheet

Fig. 5-Worn propeller duct fan.





should be made removable for cleaning and inspection purposes. The removable section should be provided with sturdy clamps and effective seals. While this alteration is being made on a centrifugal fan, it will sometimes be beneficial to rework the fan housing so the fan impeller and also the motor in the case of direct mounted fans can be removed as a unit for maintenance servicing without disturbing the fan duct connections. Replaceable liner plates could also be installed at the same time for those fans where undue abrasive wear is unavoidable.

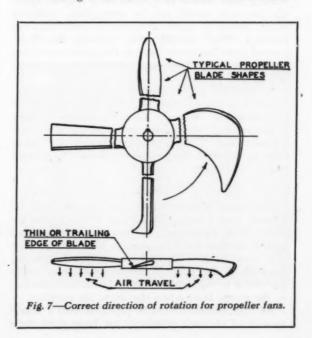
All centrifugal fans should be inspected regularly for the effects of abrasive wear or corrosion. On the impellers watch carefully for the appearance of eroded rivet and bolt heads, and for cutting of the blades next to the rear impeller disc. Steel impellers can be repaired by welding on wear or patch plates (Fig. 4) if the impeller is carefully balanced afterwards. Balancing is easily done by adding arc weld beads to the patches. Make all patches in a uniform pattern both radially and lengthwise of the impeller since patches at different distances from one end of the impeller may balance on the balancing rolls, but the impeller will try to wobble when the fan is run at operating speed.

For all systems having dust collectors the fan should be installed after the dust collector for best fan life.

Propeller fans should not be used for handling air containing large amounts of abrasive dust because most of the impellers are made of aluminum which will wear away quite rapidly. If your propeller fan blades wear like those in Fig. 5, you should consider installation of a dust collector in that system. Convenient service doors should be installed above and below propeller fans to insure their proper maintenance.

Large mesh round wire screens should be installed below or around propeller fans wherever a flying propeller part could cause personal injury. Roof installations are frequent violators of this safety precaution. Propeller fans within ready reaching distance of anyone should have a wire guard with openings small enough to prevent fingers from reaching the fan blades. Always avoid the use of expanded sheet metal for fan guards because of the undesirable resistance to air flow.

If bearing troubles are encountered on fans hand-



ling high temperature air, make sure the fan was originally built for high temperature service. The impeller shaft of high temperature centrifugal fans will have some form of heat dissipating device between the fan and its bearings. This device may be a small fan impeller to circulate cooling air around the shaft; it may be circular plate rings, or a slotted shaft; or in extreme cases a water cooling system for the bearing may be used. Heat radiation from an uninsulated fan housing may cause bearing troubles.

Propeller fans for high temperature service will have water-cooled bearings. Pressure and temperature limit switches installed on the discharge side of any water cooled bearing are desirable protection

measures.

All mechanics and electricians should thoroughly understand the correct direction of rotation of centrifugal and propeller fans as shown in Figs. 6 and 7. Unfortunately most fans will discharge some air when running backwards so air movement is not an

adequate test for correct fan rotation.

The V-belts on all fan drives should be kept tight, and all motors should have take-up bases. When a belt in a multiple belt drive breaks, all the belts should be changed at the earliest fan shut down period. The belts in a multiple V-belt drive should have a matching number variation of not more than two. These matching numbers are stamped on the bolts in addition to the belt size number. When unmatched V-belts are used, some of the belts will not carry their share of the load thus leading to premature failure of the drive.

Foundry fans should be equipped with heavy duty grease lubricated ball or roller bearing units having excellent grease seals. Commercial bearing units are available for rebuilding or replacement purposes on most centrifugal fans. Ring oiled or grease lubricated babbit bearings are too easily contaminated with abrasive dusts in foundry installations. Ball or roller bearing units should be greased on a regular schedule, but they should not be overgreased. Greasing once a month is usually adequate for centrifugal fan bearings. Most propeller fan bearings should be greased more frequently because the grease may drain away from the top bearing of propeller fans mounted in a vertical position.

All foundry motors should be of the totally enclosed or totally enclosed fan-cooled type with grease lubricated ball bearings. Greasing should be done at the regular intervals recommended by the motor manufacturer. The air passages through the motor housing should be blown out occasionally to prevent

motor overheating.

#### Ducts

Galvanized iron or black iron can be used for foundry duct systems. In the case of black iron the inside of the ducts and all other duct parts exposed to the weather must be given a heavy duty corrosion resistant coating. Bitumastic type paints are excellent for this purpose. Pressure duct leaks can be prevented by applying a heavy bitumastic material to the flange faces when the ducts are installed.

All duct systems should be provided with a liberal number of cleanout openings. For large ducts these openings should be large enough for a man to enter. Many overhead ducts can thus be conveniently inspected and cleaned. These openings are especially useful on both sides of all fans. Clamp tight seals should be provided on all pressure ducts, and close fitting covers can be used on suction ducts.

Where duct dampers are needed use sturdy dampers and pivot rods, and heavy duty quadrants which should be installed with the control arm always parallel to the damper itself. When a system has been properly balanced, the damper quadrants should be clearly marked with the proper position. Otherwise incorrect settings will creep into the system, and overventilation of some hoods may cause the unwanted pick-up of fine materials while other hoods will suffer from underventilation.

When altering a dust collecting system, always be wary of overhead unused ducts or ducts too large for the suction pipes leading into them because they may

fill up with dust and collapse.

Mold conveyor pouring, cooling, and break-apart ducts should be cleaned regularly to prevent the accumulation of combustible materials. Often sufficient core oil fumes and moisture are contained in the exhaust gases to cause a sticky deposit inside the duct. These ducts can be cleaned by installing steam jets at the duct entrance and drain openings at the proper places in the ducts to permit periodical steaming out of the sticky coating. If ducts plug up during cold weather only, condensation is the trouble, and the duct surfaces are too cold. Insulation of the duct or heating of the duct should eliminate the trouble.

A dangerous fire hazard frequently exists in core oven exhaust systems where a thick resinous coating which is difficult to remove often builds up on the inside of the ducts. This trouble can be eliminated for ducts where the gas temperature is above 250 F by installing double wall sheet metal ducts (Fig. 8) with a 2-in. dead air space between the inner and outer ducts. Conventional insulation of the duct may seem to be a more simple answer, but when the original duct is replaced, as it should be to secure a completely safe system, the double wall sheet metal duct is the most convenient and economical to install.

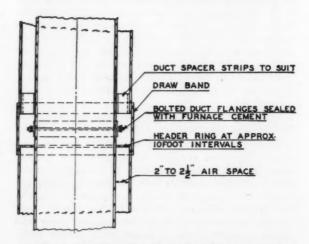


Fig. 8-Double wall exhaust duct for core ovens.

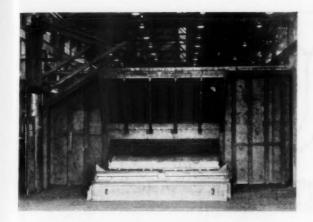


Fig. 9—Heavy duty shake-out side exhaust hood.

#### Hoods

The rules for effective foundry exhaust hood maintenance are (1) build rugged hoods, (2) provide adequate rugged doors for the maintenance or removal of any enclosed equipment and for cleaning the hood, (3) make the hood large enough for convenient maintenance of enclosed equipment, and (4) clean regularly all hoods where dirt build up occurs.

All foundry exhaust hoods should be built to withstand vibration, impact, erosion, and corrosion. The minimum metal thickness for any hood should be 16 gage and 10 gage material can profitably be used for hood sections near floor level. Impact exposure can be expected for most floor level hoods. In order to provide flexibility for miscellaneous maintenance operations hoods, as much as possible, should be made up of angle iron flanged panels bolted together.

Heavy duty crane serviced shake-out hoods should be constructed of 1/4-in, plate unless a rugged set of bumper rails protects the sheet metal hood. Figure 9 shows a crane serviced shake-out hood constructed of 1/4-in, plate which has withstood many years of heavy duty useage. The hood face entrance slots are formed by 3-in, steel pipes welded into a substantial structural steel frame with cover plates welded to pairs of the pipes to allow the hood designer a choice of air inlet slot widths.

This hood also has a sturdy door behind the left wing which provides easy entrance for lubricating and inspecting the shake-out and for cleaning the underground exhaust duct entrance screens. This door has heavy steel block hinges with 5%-in. hinge pins and heavy bar steel latches to withstand the vibration and impact which the hood receives. This entire exhaust hood can be removed in a single piece by the 5-ton overhead crane whenever shake-out replacement is necessary. The wings at the ends of the hood control the effects of strong cross drafts.

A shake-out exhaust hood which includes many desirable features is shown in Fig. 10. This hood has two well built doors large enough for convenient removal of the shake-out by means of the two built in monorails and chain hoists. A service door, adequate room inside the hood, and built in lights provide for

convenient routine maintenance of the enclosed equipment.

Mold conveyor cooling tunnels should be wide enough and high enough for convenient entry for cleaning, inspecting, or repairing of the mold conveyor cars, track, and chain. A dislodged mold in a close fitting tunnel can cause a wreck which is difficult to repair.

Bucket elevator and conveyor belt discharge points should be provided with liberally sized hoods with the suction connection located in the position to provide the greatest gravity settling of the sand fines.

Exhaust hood faces and interiors, especially those serving mold pouring, cooling, and break-apart zones, should be cleaned regularly to reduce the fire hazard from soot or core oil resin deposits. Sand mixer exhaust systems also need regular cleaning and inspection for corroded ducts.

Wheel hoods for heavy duty stand grinders illustrate the application of most of the foregoing rules. The hood should be built of heavy plate for safety reasons. A heavy hinged door with substantial locks provides safety and convenient wheel removal. A well fitted cleanout door or dust box is needed for regular cleaning purposes. The hood side plates should fit closely to the side of the wheel, and an easily adjustable spark guard is needed, to reduce the effects of wheel fan action which may be compared to a cross draft.

#### Air Heaters

The two types of air heaters most used in foundries (1) the unit heater with a propeller fan, and (2) the multiple centrifugal fan heater have similar maintenance problems. The air heating coils should be cleaned twice a year because most air heater fans are also used for air circulation during the summer months. For most effective heater cleaning both sides of the coils and fans should be conveniently accessible. Catwalks should be provided for all heaters which cannot be safely maintained from a portable ladder.

If only dry dust has deposited in the heater coils, an air jet will clean them, but the dislodged dust is

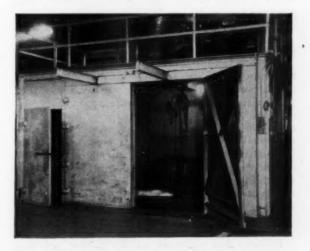
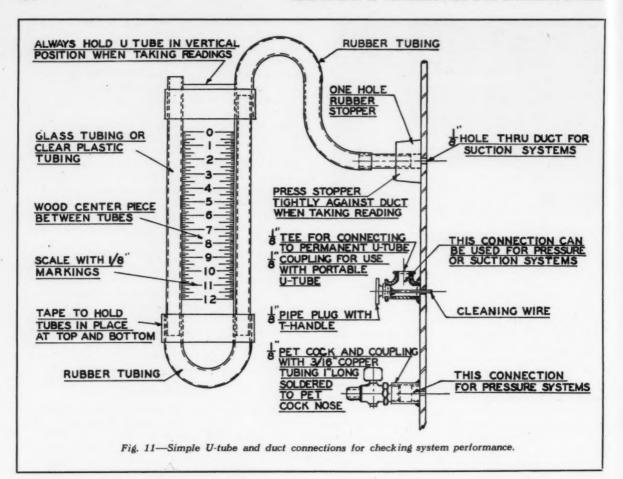


Fig. 10—Shake-out exhaust hood with built in maintenance hoists.



a nuisance. If the dirt is hard or at all sticky, a steam cleaner will do a much better and quicker job. For convenient coil cleaning the bolted on fan mounting used on most unit heaters can be changed to a hinged mount so the fan can be easily swung aside. A simple light weight shield and drip pan can then be hung on the heater to catch the dirt and drippings as the heater is steam cleaned. More carefully fitted drip pans are needed for steam cleaning centrifugal fan heaters.

The centrifugal fan heater impeller will need more frequent cleaning than does the propeller fan.

Wherever possible air heaters should be arranged to heat outside air in order to provide make-up air for the foundry exhaust systems. Non-freeze type coils give greater safety from coil freeze-up in very cold weather. A thermostat installed in the hot air side of a heater system can be used to operate intake and recirculation damper controls as protection against both steam trap plugging and extremely low incoming air temperatures. However, these low temperature controls should be periodically checked for proper operation during cold weather by turning off the steam to the heater for a short time.

#### **Dust Collectors**

Foundry dust collectors can be divided into two general classes (1) the dry collector which collects dusts in the dry state and (2) the wet collector which

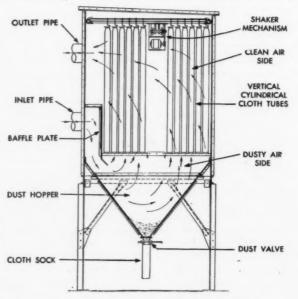


Fig. 12-Vertical cloth tube dry dust collector.

collects dusts by the use of water. Generally, the collected material from the dry collector is removed from the collector and disposed of in the dry state. The collected material from wet collectors is re-

Fig. 13-Felt tube dry dust collector.

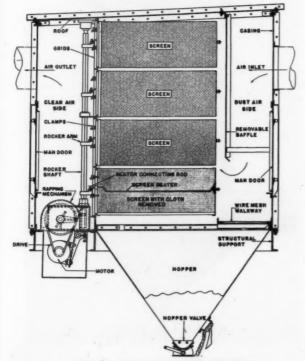


Fig. 14—Flat cloth envelope dry dust collector.

moved from the bottom of the collector by hand or by a chain drag conveyor as a sludge, or it is sluiced from the bottom of the collector to settling tanks or basins where the water can drain off leaving a damp material.

The types of dry dust collectors are vertical cloth tube collector (Fig. 12); heavy felt tube collector (Fig. 13); flat cloth envelope collector (Fig. 14); cyclone collector (Fig. 15); multiple cyclone collector (Fig. 16); and dry rotary centrifugal collector (Fig. 20).

The types of wet dust collectors are wet cyclone col-

Fig. 15-Dry cyclone dust collector.

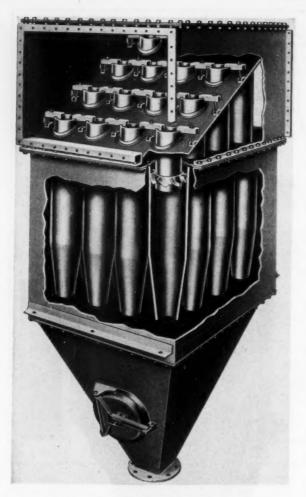


Fig. 16-Multiple cyclone dust collector.

lector (Figs. 17 and 18) wet rotary centrifugal collector (Fig. 21) and turbulent water bath collector (Figs. 22 and 23).

There are three general rules for the proper maintenance of all dust collectors (1) remove collected material regularly from the collector hoppers, (2) inspect all parts of collector on a definite schedule and (3) do not neglect the proper repairs to the collector. Adequate inspection and service doors are a must for all dust collectors.

When the hoppers of dry dust collectors are not emptied regularly, the dry dust will frequently build up to the point where it will be recirculated, thus causing undue localized wear. In the case of the dry cyclone the dust will be carried through the collector if the hopper is too full of material.

If the collected material is not removed from wet collectors often enough, sludge will build up on the collector surfaces. When this build-up reaches the moisture eliminator section, dirt will soon cause fan trouble as discussed under fan maintenance.

A most convenient indication of the general operating condition of a dust collector is given by the pressure drop from the inlet to the outlet of the collector. If there is much change in the typical pressure drop,

look for changes in the collector or its attached exhaust system. A permanently installed U-tube is a convenient method of measuring pressure drop if the U-tube and its connections are properly maintained. However, a simple U-tube which can be used for checking both dust collectors and other parts of air handling systems can easily be made from lengths of glass or clear plastic tubing, rubber tubing, and a 1-hole rubber stopper as shown in Fig. 11. When this U-tube is used, the pressure drop across a collector is the difference between the individual readings taken at the inlet and at the outlet of the collector. For collectors with built-in fans these readings must be taken across the dust collecting section only.

This same U-tube is a useful tool for checking duct systems once a record has been made of the U-tube

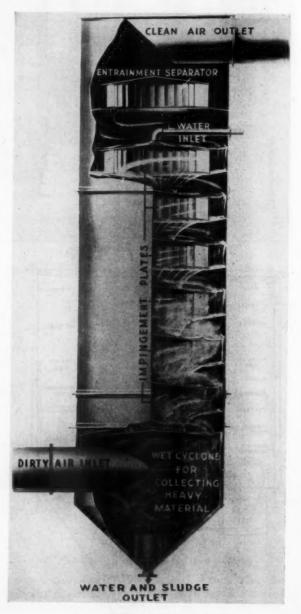


Fig. 17-Wet cyclone dust collector.

readings for definite locations with the system operating properly. Significant variations from the acceptable pressure readings will give warning that a system inspection is needed.

# Cloth Type Dust Collectors

Cloth type dust collectors should not be used in exhaust systems where moisture, core oil vapors, or gummy materials are present in the air handled.

All cloth tubes and envelopes for foundry dust collectors should be given a flame-proofing treatment because of the widespread use of welding for construction and maintenance work. All cloth tubes and envelopes should be securely sealed to their inlet header or grid. To prevent premature wear or tearing of cloth envelopes they should not be stretched too tightly when assembled into the collector. Cloth tubes should not be under tension during any part of the shaker mechanism stroke.

Cloth collector fabric leakage checks should first be made from the clean air side of the dust collector because dust leaks will reveal themselves by staining of the tubes and envelopes or by dust accumulations. Ordinarily a set of cloth tubes or envelopes will last for several years of 24-hr per day service. A worn fan impeller indicates gross neglect of fabric maintenance if the fan is installed on the clean air side of the collector as it should be.

All cloth collectors (Figs. 12 and 14) should be rapped or shaken down regularly with the fan turned off. An automatic timed relay is very useful for operating the rapping device for a set time after the fan is turned off. Since a shakedown every 8 hr is usually often enough, lunch periods make an ideal

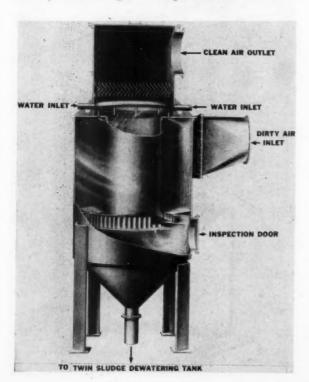


Fig. 18—Wet cyclone dust collector.

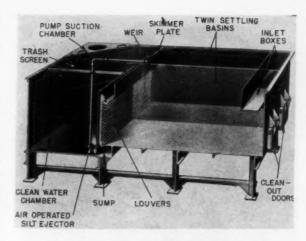


Fig. 19-Twin manual sludge dewatering tank.

time for shakedowns when the collector is used three shifts a day.

If the collector has spark screens, they should also be cleaned at regular times.

The pressure drop through the collector should not be more than 4 in. of water just before the shakedown cycle. If the pressure drop is greater than 4 in., the collector should be shaken down more frequently. If the collector pressure drop is more than 4 in. of water after a thorough shakedown, the collector is handling too much air or the cloth pores are clogged by sticky materials or from moisture in the air. The usual design volume of air per square foot of collector cloth for foundry use is 3 cu ft per min.

Points to check in a routine inspection of cloth type collectors are (1) cloth envelopes or tubes for leakage or holes, (2) complete rapping mechanism including the individual rapping hooks or cranks for insufficient rapping action, (3) collector shell for holes or material build up, (4) collector baffle plates for wear, (5) spark screens for plugging or wear, (6) hoppers for wear or dirt build up, (7) hopper valves whether rotary or dribble type for wear or defective operation which could allow air or dust leakage, (8) condition of discharge screw conveyor if one is used, (9) fan as described under fan maintenance.

# Felt Tube Dust Collectors

Since the felt tube dust collector (Fig. 13) has air jet cleaning rings around each tube which travel intermittently up and down the outside of each tube as required, the pressure drop across the collector should remain almost constant from day to day. Consequently this type of collector should be equipped with a permanently connected U-tube for measuring the pressure drop because a change of pressure drop can be the signal for possible trouble. A decrease in the pressure drop can indicate a hole in the collector tubes, an obstruction in the attached exhaust system, or plugging of the U-tube piping. An unusual increase in the pressure drop can indicate a failure of the air jet traverse equipment or air supply hose, or a failure of the air jet blower or blower intake filter. The pressure drop will also change if the inspection doors are left open.

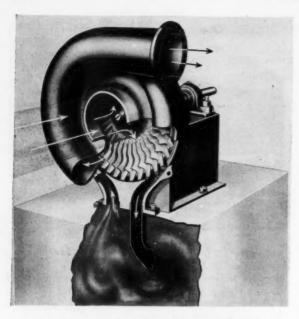


Fig. 20-Rotary centrifugal dry dust collector.

# Dry Cyclone and Multiple Cyclone

Cyclone collectors (Fig. 15) for foundry service should be built of ½-in. and ½-in. steel plate because of the abrasive nature of foundry dusts. Flanged sectional construction, wear plates, and in some cases rubber liners are useful features for maintenance convenience. Cyclones serving as sawdust collectors need not be as heavily built.

Multiple cyclones (Fig. 16) should have their inlet and outlet connections tightly sealed into their respective header plates to prevent dust from bypassing the cyclones. Multiple cyclones using short multivaned swirl rings should be protected against wood pieces, paper, and rags by installing 3/4-in. mesh round wire screens at the duct inlets. When necessary, chains suspended in the center of the individual small cyclones will keep them relatively free of deposited material.

# Dry Rotary Centrifugal Collectors

Dry rotary centrifugal collectors (Fig. 20) combine the dust separator and fan functions into a single unit. For foundry service these units should have liners in the back housing at the rear of the scroll casting to eliminate wear on this housing. In many foundry installations a precleaner, also with wear plates, should be installed on the collector inlet.

Empty the dust hopper regularly so the collected dust will not be carried back into the impeller. Be sure the dust hopper is vented back to the collector. Inspect the impeller and wear liners at 6-month intervals. If the tips of the impeller blades wear off, the entire impeller should be rebladed.

#### Wet Type Dust Collectors

While the wet type dust collector provides a nonhazardous collected material discharge, the collector and its material handling equipment must be properly maintained to prevent undue wear, corrosion, or material build-up.

All collector surfaces in contact with water or moist air should be coated with a high quality water resistant, abrasion resistant, and rust preventive coating. These coatings may be paints, sprayed plastics, or bituminous base materials depending upon the severity of the exposure. Since a material build-up tends to take place slowly on the internal surfaces of wet collectors, these surfaces should be carefully cleaned and the protective coatings repaired once a year. Material which can be dislodged by flushing should be removed at more frequent intervals. The wet collector should be repainted inside and outside at least every 5 years.

# Wet Cyclone Collectors

The wet cyclone collector (Figs. 17 and 18) must be supplied with an adequate flow of clean water. A fresh water supply reduces the maintenance cleaning of the collector, but in many localities fresh water is too expensive. Where a water recirculation system is used adequate settling tanks must be provided in keeping with the equipment manufacturers specifications.

The fan of a wet cyclone collector should not be run unless water is flowing through the collector. When the collector is being shut down, the water should be allowed to flow through the collector for a half hour after the fan has been turned off in order

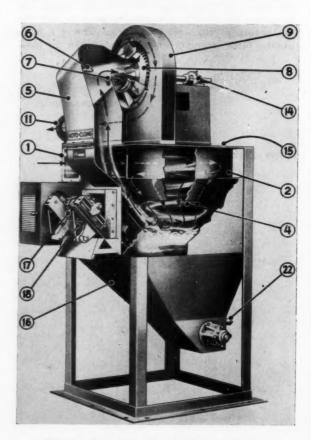


Fig. 21-Rotary centrifugal wet dust collector.

to adequately flush the collector.

The collector should be checked throughout at 3-month intervals for unusual accumulations of materials. The collector should be thoroughly cleaned at least once a year.

If spray nozzles are used in the collector, they should be checked for plugging at least once a week.

## Sludge Settling Tanks

Proper operation of sludge settling tanks is essential. Do not try to completely fill the settling chambers of dewatering tanks (Fig. 19) with sludge. Small tanks should not be filled over half full and large tanks not over two-thirds full in order to provide a very low water velocity across the tank. Silt should be removed from the bottom of the clean water chamber at regular intervals. Not over 7 in. of sludge should be allowed to accumulate in the bottom of wet collectors with built in sludge settling tanks.

Sludge settling tanks with drag chain sludge conveyors are built as separate assemblies or as integral parts of dust collectors as shown in Figs. 22 and 23. These sludge conveyors should be equipped with an automatic timer which will run the conveyor for at least 45 min after the fan is turned off in order to remove sludge which may harden in the bottom of the tank during a weekend shut-down. In any case the bottom of the tank should be equipped with replaceable wear plates or replaceable guides.

If a bath of used oil is kept floating on top of the water in the section of the tank where the drag chain emerges from the water, wear and corrosion of the

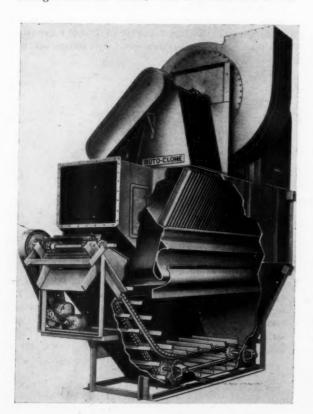


Fig. 22—Turbulent water bath dust collector.

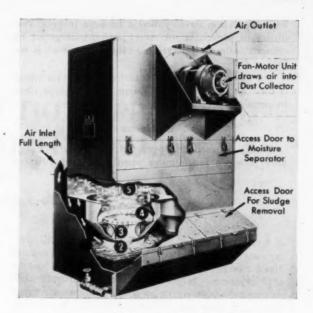


Fig. 23-Turbulent water bath dust collector.

drag chain will be greatly reduced. Where necessary a simple baffle can be installed to prevent the oil from flowing out the overflow connection.

Pumps for handling water from sludge settling tanks should be of the type designed for handling abrasive materials.

#### Wet Rotary Centrifugal Collector

The wet rotary centrifugal collector (Fig. 21) is quite similar to the dry type (Fig. 20) with the addition of a wet precleaner and a water spray at the impeller inlet. Clean water should be supplied to the spray nozzles. The impeller spray nozzle should produce a symmetrical pattern to adequately flush the impeller. All nozzles should be checked periodically for plugging or wear.

The drain pipe connected to the collector air outlet should be checked periodically for plugging. A plugged drain may cause the formation of a water trap which in turn may reduce the volume of air handled by the collector.

Sludge ejector equipment and collector surfaces' should be maintained as outlined in the sections on sludge settling tanks and on wet type dust collectors.

Turbulent Water Bath Collectors—Turbulent water bath collectors (Figs. 22 and 23) use baffles to force the air to travel through the water bath. These baffles and the moisture eliminator plates should be checked regularly for dirt buildup. A high pressure water jet with an ample length of hose should be provided for periodically washing down most wet collectors.

When baffles are removable for cleaning, they must be replaced correctly. When the moisture eliminators are replaced, the angle lips should face against the air flow.

The water overflow pipe should always be kept open to prevent too high a water level which will reduce the air capacity of the collector. The water level regulation devices should be kept in good condition to prevent unnecessary water wastage or inefficient collector operation if the water level is too low.

The power driven rotary turbulence device which is used in some collector designs must be kept clean to prevent vibration.

#### Conclusion

A well planned and operated preventive maintenance program for foundry ventilation and dust collecting equipment will not only permit full use of production equipment, but it will provide cleaner more comfortable working conditions in the foundry.

# Acknowledgment

The writer would like to express appreciation to the Sturtevant Div., Westinghouse Electric Co., American Wheelabrator & Equipment Co., Newcomb Detroit Co., Pangborn Corp., Buell Engineering Co., Western Precipitation Corp., Claude B. Schneible Co., Whiting Corp., American Air Filter Co. and Caterpillar Tractor Co., for illustrations used in this paper.

# REFRACTORY SELECTION APPLIED TO BATCH TYPE AIR FURNACES

By

C. O. Schopp* and P. F. Ulmer*

#### ABSTRACT

This paper deals with a specific set of conditions representing experience in the authors' plant.

It is believed that the experience here could be applied to similar operating conditions in areas suitably located to sources of refractories. Methods of selecting refractories manufactured from raw materials by processes that result in a product which will perform its intended function for the user are desired.

For a given locality, the refractory producing the most metal per unit, i.e., in terms of tapping block, sidewall brick or bung brick, will be the best. The chemical and physical make-up of the brick may be of secondary importance to proximity to the refractory plant, should transportation cost of the superior product render it uneconomical. The selection is properly made when the most economical refractory that will safely withstand the desired cycle of operations is found.

This paper is designed to outline the experience in a specific plant when using refractories selected after many years testing by actual operations. During these years numerous tests have been suggested by suppliers, In this operation white cast iron is produced in batch-type furnaces, generally two successive heats per furnace per day. After a 5-day cycle of ten heats, major repairs generally are found necessary. The furnace is rapidly cooled by slowly flowing water on the bottom after the last heat of the week. This rather drastic cooling procedure is necessary in order to reduce the temperature sufficiently so that repair crews can inspect the sidewalls and either completely



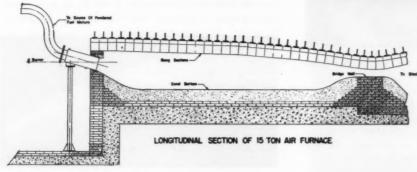


Fig. 1 (above)—Fifteen-ton air furnace ready for operation.

Fig. 2 (left)—Longitudinal cross section of air furnace.

Fig. 3 (below)—Cross section of air furnace through the tapping block.

and it is felt that these tests have been evaluated with but one thought—that of obtaining a more satisfactory refractory for the application, from both operating and cost standpoints.

The term "air furnace" is restricted to apply to a reverberatory type powdered-coal-fired-furnace, in this case of 15-ton capacity, used for melting white cast iron. This iron when properly heat-treated results in the blackheart malleable iron, which is an important engineering construction material of today.

Sord Bottom

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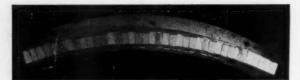


Fig. 4—Lined air furnace bung ready for use.



Fig. 5-Metal bung frame for air furnace.



Fig. 6—Take-up clamp for holding air furnace bung brick in metal frame.

reline or repair the lining. The repair foreman must have experience enabling him to determine whether the lining must be replaced or whether minor repairs will permit another cycle of service to be obtained.

Repair of the sidewalls consists of replacing such sections of the wall that appear to be too thin to withstand another operating cycle.

Bungs used in the roof construction are repaired when necessary, and may need to be replaced daily or during the heat should the occasion arise.

The bottom is replaced each time the furnace is repaired or relined.

The tapping block generally is replaced each time the furnace is cooled for repairs.

Construction of the air furnace used for melting follows the general plan of published diagrams. Figure 1 shows the general exterior structure. Sectional and cutaway views are shown in Figs. 2 and 3. Figure 2 shows a longitudinal cross section, while Fig. 3 shows a cross section through the tapping block area.

The metal shell and outer layer of brick serve to maintain the general shape of the furnace and to support such accessories as burners, skimming-bar supports and tapping blocks, as well as the refractory lining. The lining should be readily removable, otherwise the cost of maintenance would be unduly increased. The inner lining need only support the roof and confine the melt and slag. Forces are either down or outward toward the permanent shell, which has ample strength to back up the refractory.

The structure of the bungs is shown in Fig. 4, which illustrates a completed bung ready to go into service. The metal frame is shown separately in Fig. 5, and a closeup of the bung "shoe" clamping arrangement in Fig. 6. This arrangement holds the brick in the arched form.

Figure 7 shows a bung which has been in service. Enough refractory remains in the frame to allow its use for limited service such as would be encountered in the roof of the furnace near the stack.

Figure 8 shows a bung in which the brick has worn thin. Part of the one side has dropped out, rendering replacement imperative. As indicated above, the bung shoes must be kept tight as the bricks may loosen during the heating and cooling cycle and the entire lining will fall out of the bung.

The tapping block used in this operation is a refractory form with a wedge block which is used to hold the block in place. Considerable difficulty was encountered in selecting a refractory for this service. It is highly desirable that this section perform satisfactorily throughout a cycle of heats, in this case ten, as it is very difficult to repair or replace this part without cooling the furnace to allow access to the interior.

Experience in this plant indicates that a hardburned stiff-mud process brick made of a Missouri clay is most satisfactory for sidewall use. In this lo-



Fig. 7—Bung brick suitable for further limited service.



Fig. 8—Bung requires relining due to extensive wear on bricks.

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cation the brick must resist penetration by molten iron and slag, as well as withstand a high flame temperature and the fluxing action of the ash.

Bricks used in the bung sections which form the roof of the furnace are of a dry-pressed Kentucky clay. This brick must withstand severe heat shock; otherwise spalling will cause the loss of refractory. Access for charging the furnace is obtained by raising sections of the roof immediately after the previous heat has been removed.

The roof is subjected to a wide range of operating temperatures. Immediately in front of the burners, where the fuel burns intensely and the radiant energy of the flame is high, the roof temperature will be very high. In other zones, it is quite likely that the temperature will be relatively low. Brick suitable for general use throughout the roof section must have a minimum amount of expansion or contraction under operating conditions. Otherwise, the bricks are likely to fall out of the frame or, if expansion should be high, may crack the frame.

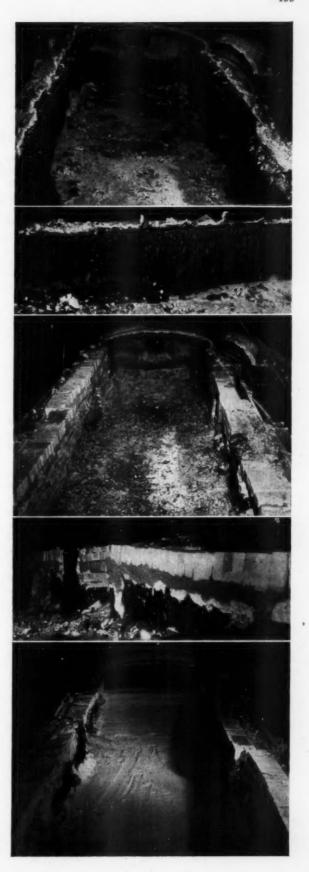
Fluxing action of the ash will vary as the quantity and composition of the ash and the roof temperature are quite different in various parts of the roof. Bung bricks are a very specialized product and are rightly used only for this application.

# Sand Bottom Withstands Cooling Stresses

The furnace bottom is made of silica sand for economic reasons. A brick bottom would be destroyed by the cooling water necessary to reduce the temperature for repair purposes. Certain precautions are necessary to prevent the vitrified layer next to the melt becoming damaged so metal will penetrate into the sand. Apparently, cooling stresses or cracks form under a large mass of metal allowed to freeze in the furnace so that sections of the bottom rise with the solid metal as the heat is melted. Large portions of the bottom can be lost in this manner. Metal filling the holes in the bottom cannot be removed since the level of these holes is below that of the tapping block. A new bottom is covered with boards to prevent damage when the first charge is dumped in.

The tapping block is likewise subjected to very severe service. In reality, the tapping block is expected to play a double role. It should prevent the conduction of heat to the outside of the furnace while the heat is being made, and since refractory is not a perfect insulator, the block should be a high conductor while the stream of metal is flowing through it so that the stream does not freeze. While the melt is being removed from the furnace through the tapping block, the temperature can become quite high compared to the refractory surrounding the block. It is obvious that large temperature gradients exist which react to cause spalling and cracking of the tapping block. The refractory used in construction of this block should satisfactorily resist the fluxing action of hot metal and slag for a period necessary to produce

Figs. 9 to 13 (right, top to bottom)—Fig. 9—Air furnace before starting repairs. Fig. 10—View of worn side wall. Figs. 11, 12, and 13—Various stages in the repair and replacement of the furnace lining.



a cycle of operations. Selection of the refractory material entering into the construction of these blocks is very important because of the difficulty of replacement should a failure occur during the operation cycle.

Considerable testing of various compositions of refractory were necessary before a satisfactory block was obtained. The block used currently is made of Kentucky clay. At present, an occasional failure occurs before completion of a ten-heat cycle.

After the roof has been removed and the furnace cooled sufficiently to allow access, the repair foreman must decide the extent of repairs necessary. Figures 9 and 10 show the condition of the lining after a cycle of service. Various stages in the repair and replacement of the lining are shown in Figs. 11 and 12. In Fig. 12 one sidewall has been replaced and the other patched. Figure 13 shows the same furnace as in Fig. 11, but looking toward the stack after the bottom has been smoothed so that it tapers gradually to the bridge wall in the back of the furnace and to the tapping hole on both sides.

A measure of refractory performance is best expressed in terms of refractory per ton of metal melted. Average costs in this plant, in terms of refractory units per ton of metal melted, are as follows:

.....4.0 Bricks/Ton Bung Brick 

From a physical standpoint, the refractories used at present show maximum resistance to spalling both from normal cooling and accelerated cooling, and to metal penetration.

Chemical properties such as melting point and resistance to fluxing or slagging by ash and slag, are not controllable because of the changing conditions during production of the heat. It is believed that the attack by ash and slag is accelerated as the temperature increases, and that fluxing action by the ash will vary as the quantity of ash impinged on the brick varies. The quantity of ash contacting the refractory depends on the location in the furnace and with the shape of the charge which will deflect an undetermined amount of furnace gas toward the refractory. Efforts are made to minimize the chemical attack of the ash by properly directing the flame on the surface of the bath. This condition of proper flame impingement is only obtainable after the charge is melted. Erosion and attack by the slag, which circulates because of the action of the flame, are likewise not controllable because of variable quantities and composition, depending upon the make-up of the charge. Variations in the composition of the coa and the resulting ash, and the completeness of the removal of slag when skimming likewise affect the rate of attack on the refractory.

#### Conclusion

It is recognized that in the foregoing discussion certain theoretical aspects have been made secondary. It is felt that a proper balance between the ideal properties desired in the refractory and the delivered cost is the happy medium in the selection problem.

The purpose of this paper is to describe the steps necessary to select a suitable refractory for a given operation when precise methods are not applicable, due to the lack of control of a major number of the conditions affecting the life of the refractory. On the job tests of sufficient duration to establish the suitability of carefully selected samples should, in a short time, give enough data to intelligently evaluate a new sample. Considerable weight should be given to suggestions of refractory engineers and experienced

It is hoped that the procedure given will be helpful to those faced with the problem of selecting materials under similar conditions, whether it be melting metal or one of the many problems frequently confronting the practical and technical worker alike.

#### DISCUSSION

Chairman: W. G. FERRELL, Auto Specialties Manufacturing Co., St. Joseph, Mich.

Co-Chairman: ERIC WELANDER, John Deere Malleable Works, East Moline, Ill.

Recorder: Wm. ZEUNIK, National Malleable & Steel Castings Co., Indianapolis.

J. H. Lansing: 1 This paper shows that brick bottoms are not necessary in small tonnage air furnaces.

H. E. LEICKLY: What is the height of the sidewalls?

Mr. Schopp: Height of sidewalls is 28 in.

CHAIRMAN FERRELL: Why do you have difficulty in getting 10day life from tap-out blocks? We used to get 20 heats.

Mr. Schopp: We are pouring small castings. The average weight of our castings is less than 1/2 lb and we do our tapping in small hand ladles. Therefore, the hole has to be plugged more often. We feel that it is bar erosion that causes our short life.

R. A. WITSCHEY: 3 Are your figures expressed as units or as 9-in. equivalents?

Mr. Schopp: Our figures are expressed as units.

Malleable Founders' Society, Cleveland.
 Fanner Manufacturing Co., Cleveland.
 A. P. Green Fire Brick Co., Chicago.

# INFLUENCE OF PHOSPHORUS ON HOT TEAR RESISTANCE OF PLAIN AND ALLOY GRAY IRON

By

John C. Hamaker, Jr.* and Wm. P. Wood**

#### ABSTRACT

In order to study the properties of gray iron under actual hot tearing conditions, a procedure was developed for tensile testing on direct cooling from the liquid state in a sand mold. Forty-four tests at temperatures of 2050 to 1650 F on a 2.90 per cent T.C., 2.10 per cent Si base iron containing various phosphorus and alloy additions provided hot strength and plastic deformation data for the determination of the hot tearing temperature range in gray iron. Irons containing 0.04 to 0.16 per cent phosphorus were found to be most susceptible to hot tears in the range 1900 to 1800 F, while 0.25 per cent phosphorus irons were susceptible over the extended range 1900 to 1700 F. At these temperatures, the various compositions possessed tensile strengths of 1600 to 3500 psi, thus indicating the magnitude of hindered contraction stresses necessary for hot tear formation.

Due to the presence of liquid phosphide eutectic, the ultimate strength values in the hot tearing range varied consistently with phosphorus and alloy content. The hot tearing tendency of plain gray irons did not increase appreciably until the phosphorus content exceeded 0.16 per cent, while alloy irons exhibited marked strength reductions when the phosphorus was increased from 0.04 to 0.07 per cent and lost over 30 per cent of their base strength at 0.16 per cent P. Molybdenum and chromium additions sharply reduced the hot strength of plain iron at phosphorus levels as low as 0.10 per cent by segregating to the phosphide eutectic to increase its volume. Nickel also decreased the hot strength to some extent at phosphorus levels above 0.12 per cent.

In the early days of gray iron founding, hot tears or hot cracks were seldom encountered simply because the high carbon compositions experienced very little net contraction on cooling after solidification. However, the present trend toward more complex casting designs with high strength specifications has resulted in internal contraction stresses approaching those of steel. If these stresses are sufficiently concentrated to exceed the strength of the iron at some temperature after solidification, a hot tear is formed. In many cases, small cracks arising from this source are not detected until the casting has been completely machined, when they may be falsely attributed to hot shake-out, rough handling, or residual stresses.

This type of defect is a particular problem in the production of heavy castings, where very low carbon equivalents are necessary to obtain the desired properties in a heavy section. Since much of this work is of the jobbing type, extensive pattern changes to remove points of high stress concentration are frequently out of the question. Some reduction in metal contraction may be achieved by adjusting the carbon-silicon ratio, but this is also limited by practical aspects of cupola operation and the efficiency of inoculation treatments.

The problem of further reducing the hot tearing tendency is thus resolved to increasing the strength of the iron shortly after solidification. Unfortunately, there is little or no data available on the strength of gray irons above 1200 F, which is considerably below the temperature range in which hot tearing would be expected to occur. Due to the low melting point of the phosphide eutectic and its tendency to form a liquid network around the iron-carbon eutectic cells, the phosphorus content would be expected to have the most profound influence on the strength of gray iron in this range. This has been indicated by industrial studies in which reductions in base phosphorus content have eliminated hot tearing problems. It has also been observed in several instances that balanced alloy additions may increase the tendency for hot tear formation, although the chill and graphitic carbon content remain essentially constant. Another investigation of phosphide eutectic properties revealed that both molybdenum and chromium segregate strongly to the liquid phosphide, increasing its volume considerably and, in the case of molybdenum, lowering its freezing point.

Based on these observations, an investigation was undertaken to determine the influence of phosphorus and alloys on the strength of a selected base iron at temperatures within a few hundred degrees of the iron-carbon eutectic.

#### Preliminary Experiments

In order to determine the reliability of using standard high temperature tensile tests to study the properties of gray iron under hot tearing conditions, the microstructures of test bars quenched directly from the mold were compared with similar specimens slowly

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This paper is based on a thesis submitted in partial fulfillment of the requirements for the degree of Doctor of Philosophy in the Horace H. Rackham School of Graduate Studies at the University of Michigan.

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 Professor of Metallurgical Engineering, University of Michigan, Ann Arbor, Mich.

cooled in the mold and reheated for various times at corresponding temperatures before quenching. In all cases, the microstructures were vastly different due to the irreversible reactions experienced by the graphite flakes and the phosphide eutectic on cooling to room temperature. Much of the phosphide eutectic which is present as a fairly continuous liquid network above 1750 F was redissolved in the ferrite on slow cooling to room temperature due to a peritectoid gamma-alpha transformation, and further reheating resulted in a dispersed phase which would have little effect on high temperature properties.

This conclusion was confirmed by the results of short time tensile tests at 1850 F shown in Table 1.

TABLE 1-TENSILE PROPERTIES OF PLAIN AND ALLOY GRAY IRON ON REHEATING TO 1850 F

		Tensile	Breaking	
Specimen Number	Composition	Strength, psi	Load, lb	% Elongation in 3 In.
T-1	Plain	2303	500	6.35
T-2	2.00 Ni	2180	500	4.65
T-4	1.00 Cr	2715	400	5.00
T-3	1.00 Mo	2575	500	4.41
T-9	0.54 Mo	2325	500	5.09
T-11	1.50 Ni, 0.50 Mo	2303	500	4.52

For these tests, 1.2-in. bars were poured from Detroit indirect arc furnace heats of base composition 3.17-3.25 per cent total carbon, 2.22-2.27 per cent silicon, 0.69 per cent manganese, 0.12 per cent sulphur, and 0.13 per cent phosphorus, inoculated with 0.50 per cent of the total silicon, and alloyed as shown. The specimens were machined with a gauge length 1 in. in diameter and 3 in. long, and inserted in an electric resistance furnace mounted on a hydraulic testing machine. After reaching the test temperature in approximately 30 min, they were held at temperature 1 hr, and pulled at  $1850 \pm 5$  F with a head speed of 0.063 in./min.

In every case, the variations in tensile strength associated with the different alloy conditions were directly related to the graphite configurations they produced in the original casting. The 1.00 per cent Cr addition produced considerable free carbide and a relatively small amount of graphite for the greatest increase in strength, while the 2.00 per cent Ni iron exhibited long interlocking flakes with a resulting strength reduction. The 0.54 per cent and 1.00 per cent Mo additions refined the flake structures for corresponding strength increases, while the balanced Ni-Mo iron had a graphite structure almost identical to that of the plain iron and produced the same tensile strength. Although the molybdenum and chromium bearing irons contained considerably larger quantities of phosphide eutectic, the original network had been broken up on initial cooling in the mold and the dispersed liquid phase obtained on reheating had little or no effect on the tensile strengths. The close relationship with graphite configurations indicates that alloying elements exert very little influence on the strength of the matrix austenite in the hot tearing temperature range.

It was apparent from these studies that a procedure for testing gray iron on direct cooling from the liquid state would be necessary to obtain reliable data on its properties under actual hot tearing conditions. In addition to providing the same phosphide eutectic and graphite structures as those found in a casting in the hot tearing temperature range, this method would completely eliminate the potential errors produced by oxidation and growth in ordinary high temperature tensile tests.

#### **Experimental Procedure**

A direct cooling tensile testing procedure of this type was developed by Hall^{1,2} in an investigation of the hot tear resistance of various cast steel compositions. Using a tensile machine of special design, 1-in. specimens cast around bolts placed in the mold cavity were pulled at a series of temperatures following solidification. In order to limit the cooling range during the tests, it was necessary to employ extremely high rates of strain which probably resulted in abnormally high strengths, although the relative values were valuable for the comparison of various compositions.

For this investigation, the mold shown in Fig. 1 was designed to provide a very slow cooling rate to a

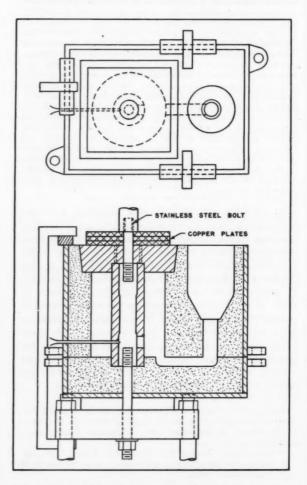


Fig. 1—Mold design for direct cooling high temperature tensile tests.

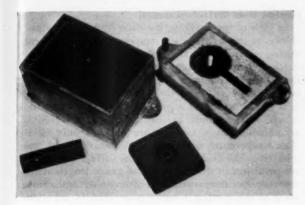


Fig. 2-Mold components.

specimen for testing in a standard beam-type tensile machine at normal rates of strain. Further details of the design may be seen in photographs of the mold and typical casting, Figs. 2 and 3. The specimen was cast in a hollow cylindrical core placed in the center of a mass of metal 4 in. in diameter which provided the slow cooling rate. The core was drilled on a lathe to produce a specimen 1 in. in diameter, 51/2 in. long, with a gauge section 7/8 in. in diameter and 2 in. long. An ingate 5/8 in. in diameter and a thermocouple hole were drilled below the gauge length on a jig, and the core set firmly in place between core prints in the drag and top core block. Sandblasted 1/6-in. stainless steel studbolts inserted 1 in. into each end of the mold cavity were fastened to the components of the tensile machine. Three 1/4-in. copper plates were placed around the top bolt to equalize the chill produced in the bottom bolt by the base of the machine. An alundum-tipped chromel-alumel thermocouple cemented 1/16 in. into the mold cavity and connected to a potentiometer recorder provided temperature control.

The complete experimental setup, with the mold inserted in the tensile machine, may be seen in Fig. 4. After checking the metal temperature with a platinum-platinum rhodium immersion thermocouple, the mold was poured at a constant rate from a platform set before the machine. A stopwatch, started at the instant the mold was filled, provided a basis for all further adjustments and readings. The mold clamps, weights, and support bar for the bottom bolt were removed after 12.0 min to maintain constant cooling conditions for successive specimens. Shortly before testing, the tare weight was adjusted to balance the beam at zero load, and an initial load of 100 lb was set on the beam. In order to remove the play and obtain an initial stress at the desired temperature, the test was begun at a head speed of 0.351 in. per min until the beam deflected at the 100-lb load. The head speed was then immediately reduced to the test value of 0.038 in. per min and readings were obtained every 0.05 to 0.10 min until fracture.

Calibration of the tensile machine with a ring gauge revealed that it was accurate within  $\pm$  25 lb under loads up to 3000 lb, with the exception of a short range between 800 and 900 lb. In this range, the re-

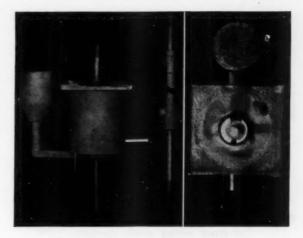


Fig. 3-Typical casting and specimen.

moval of play in the threads of the machine produced values of 65 to 75 lb above the actual load. The test head speed maintained a value of  $0.038 \pm .001$  in. per min under loads ranging from 500 to 5000 lb.

In order to determine the temperature gradients in the specimen, a series of molds were made up with thermocouples placed at various points in the gauge length and poured from the various compositions to be studied. In preliminary tensile tests, it had been found that a pouring temperature of 2700 F produced a sound specimen which cast well around the bolts, possessed a smooth surface, and turned the cylindrical core completely to powder to remove any friction from this source. When the pouring temperature was maintained constant at this value, the complete range of compositions produced nearly identical cooling curves at the top, center, and bottom of the gauge

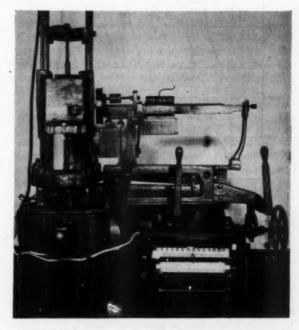


Fig. 4—Experimental setup for direct cooling tensile test.

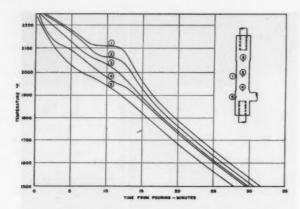


Fig. 5—Cooling curves at various points in gauge length of direct cooling tensile specimen.

length. At temperatures below 2000 F, these curves, shown in Fig. 5, did not vary by more than  $\pm$  10 F at a given time interval from pouring. This constancy was produced by the strong cooling influence of the bolts which were accurately adjusted to the same position in the mold cavity before each test. It is interesting to note that their cooling action extended into the center of the gauge length, where the surface of the specimen was maintained at a higher temperature than the center by the mass of metal surrounding the core. This reversed temperature gradient was confirmed by a slight increase in the fineness of the

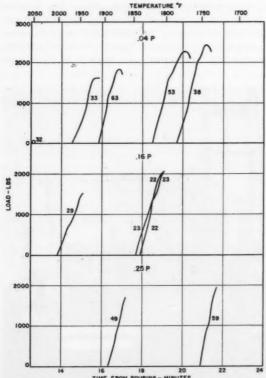


Fig. 6—Experimental load curves from direct cooling tensile tests of plain gray irons.

pearlite from surface to center.

In view of these results, the time from pouring was chosen as the criterion for temperature measurement during testing, with the thermocouple serving as a check on proper cooling conditions. The duration of the tensile tests ranged from 0.5 to 2.0 min, depending on the ductility of the metal. Referring to Fig. 5, it can be seen that a maximum interval of 2 min, which occurred only at the lower temperatures, is accompanied by a temperature drop of only 50 F. The temperature of the strength measurement was selected as the average between the surface and center temperatures in the center of the gauge length at the time the specimen reached its ultimate strength. Since none of the specimens broke more than 1/4 in. above or below the center of the gauge length, the accuracy of the temperature determination should lie within ± 25 F.

For the direct cooling tensile tests, a series of 43 181/2-lb heats of base composition 2.86-2.95 per cent T.C., 2.07-2.16 per cent Si, 0.70 per cent Mn, and 0.10 per cent S were melted in a 30-lb induction furnace from charges of wash metal obtained by melting 100 per cent steel in the cupola, ferro-85-silicon, ferro-88-manganese, and coke. The base phosphorus levels were adjusted to 0.04 per cent, 0.07 per cent, 0.16 per cent, and 0.25 per cent, and straight alloy additions of 1.85 per cent Ni, 1.00 per cent Cr, and 1.10 per cent Mo were made to the furnace. The melts were superheated to 2850 F, inoculated with 0.50 per cent of the total silicon, and poured at 2700 F.

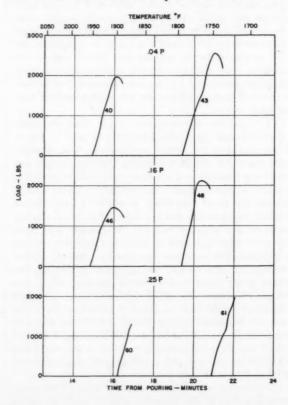


Fig. 7—Experimental load curves from direct cooling tensile tests of 1.85 pct Ni irons.

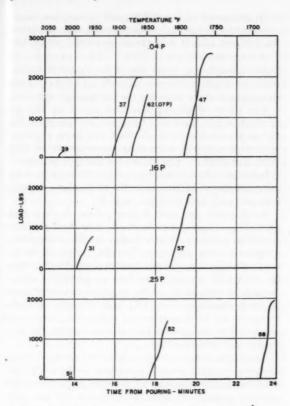


Fig. 8—Experimental load curves from direct cooling tensile tests of 1.00 pct Cr irons.

## Results

The experimental load versus time curves at various phosphorus levels for the plain, 1.85 Ni, 1.00 Cr, and 1.10 Mo irons are shown in Figs. 6, 7, 8, and 9, respectively. The corresponding average temperatures at the center of the gauge length are included at the top of the plots. In this temperature range from 2000 to 1700 F, the structure of a 2.90 per cent T.C., 2.10 per cent Si gray iron consists of 1.55 to 2.00 per cent graphitic carbon, saturated austenite containing 1.35 to 0.90 per cent carbon, and a quantity of liquid phosphide eutectic dependent on the phosphorus and alloy content.

A survey of the data for the four irons reveals that phosphorus exerts a profound influence on both the ultimate strength and the ductility in the hot tearing temperature range. As the phosphorus was increased from 0.04 to 0.16 and 0.25 per cent, the elongation at the point of maximum load and the decrease to a breaking load were completely removed, indicating a drastic reduction in the ductility of the iron. The various alloys also produced a marked change in the hot tensile properties. The nickel irons exhibited considerably more ductile fractures than plain irons at the 0.16 per cent P level, while the chromium and molybdenum additions reduced the elongation at maximum load to some extent at even 0.04 per cent P. It is interesting to note that an increase in phosphorus from 0.04 to 0.07 per cent in the chromium and molybdenum irons, as shown by Curves 62 and 64

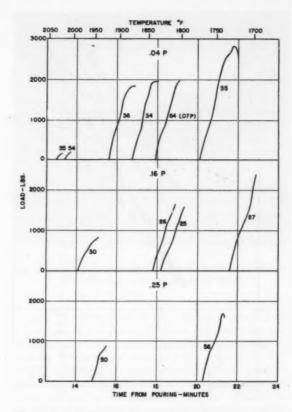


Fig. 9—Experimental load curves from direct cooling tensile tests of 1.10 pct Mo irons.

(Figs. 8 and 9) produced a noticeable reduction in tensile properties. This corresponds with the results of another investigation, in which chromium and molybdenum were found to reduce the solubility of phosphorus in austenite from about 0.08 per cent to 0.03-0.04 per cent. In fact, with the exception of the increase in ductility by nickel, all of the changes in the tensile curves produced by phosphorus and alloy additions may be explained by an increase in quantity or change in the properties of the liquid phosphide eutectic.

The form of the curves at points below the maximum load presents an interesting indication. The slight discontinuities between 800 and 1000 lb must first be eliminated from consideration, since calibration of the tensile machine showed that some error would be introduced in this range due to the removal of play in the machine. However, the curves for the plain irons (Fig. 6) show a marked change in slope with loads of 2000 lb at 1760 F, 1600 lb at 1800 F, and 1200 lb at 1900 F in the 0.04 per cent P series, and also with loads of 1300 lb at 1820 F in the 0.16 per cent P series. This change in slope is probably due to a spontaneous recrystallization of the saturated austenite, which might be expected to vary with temperature and the saturation carbon content in like manner. Similar discontinuities above 1000 lb load are also indicated in the curves for the alloy irons, although the influence of the phosphide eutectics reduced their relative intensities considerably.

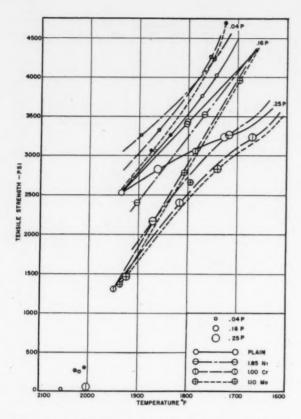


Fig. 10—Tensile strength of plain and alloy irons at 2050-1650 F on direct cooling from the liquid state.

It is also noteworthy that the average slopes of the load curves decreased consistently as the test temperatures approached 1950 to 2000 F. This would indicate that gray iron can undergo greater plastic deformation under a given load in this temperature range, thereby tending to relieve the contraction stresses before they can reach sufficient magnitudes to produce hot tears.

A summary plot of the ultimate tensile strengths against the temperatures of maximum load, shown in Fig. 10, illustrates the strength trends produced by phosphorus and alloys at various temperatures in the hot tearing range. At 1600 to 1700 F, curves drawn through points of the same phosphorus and alloy composition tend to concentrate into three distinct families based on the phosphorus content alone. At still lower temperatures, these families would be expected to approach each other and eventually intersect.

Above 1700 F, the effect of alloy additions in changing the quantity and properties of the phosphide eutectic begins to manifest itself. The plain irons approach a single point to 2500 psi at 1940 F with all phosphorus contents, while the 0.16 per cent and 0.25 per cent P molybdenum and chromium irons exhibit drastic strength reductions to approach a value of 1300 psi at 1950 F. The curves for the nickel irons fall between those of the plain and molybdenum or chromium irons at corresponding phosphorus levels.

As the temperature approaches 2000 F, the strengths

of all compositions drop off rapidly to very low values, as indicated by the sharp reduction in the strength of the 0.04 per cent P irons from over 3000 psi at 1900 F to less than 500 psi at 2000 F. The curves for the higher phosphorus irons would be expected to converge to a point of zero strength at slightly lower temperatures than those for the low phosphorus irons.

The effect of phosphorus and alloys on the hot strength of gray iron may be more clearly seen in Fig. 11, in which the strength data has been replotted against phosphorus content at temperatures of 1900, 1800, and 1700 F. The values for the 0.07 per cent P chromium and molybdenum irons were used to determine the slopes of the curves for these alloys between

0.04 per cent and 0.16 per cent P.

At the 0.04 per cent P level, there is very little phosphide eutectic present in the iron, and the alloy additions produce slightly higher tensile strengths by solution in the austenite. However, part of the increase produced by chromium and molybdenum might be attributed to the reduction in graphitic carbon content of refinement of the graphite structure noted in the preliminary tests. The high strength exhibited by the nickel irons at this level is notable, since nickel tends to increase graphitization and coarsen the flake structure to some extent.

As the phosphorus content is raised above 0.04 per cent, the strength increase produced by the alloy additions is rapidly removed by the change in phosphide eutectic properties, and the strengths of the alloy irons fall far below those of the plain iron, particularly in the higher temperature range. The close correspondence between the curves for the molybdenum and chromium irons agrees with their similar behavior in segregating to the phosphide eutectic to increase its quantity. The slightly greater strength reduction exhibited by the molybdenum irons is probably associated with their weaker tendency for carbide stabilization and the low freezing point of the molybdenum phosphide which increases its fluidity for a more complete network around the iron-carbon eutectic cells. Although nickel does not increase the quantity of phosphide eutectic, it also reduces its freezing point to result in appreciable strength reductions at higher phosphorus levels. The slight strength reduction experienced by the plain irons at higher temperatures

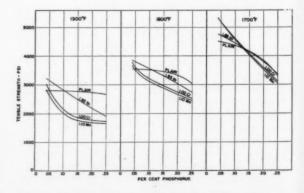


Fig. 11—Influence of phosphorus on tensile strength of plain and alloy irons at 1900, 1800 and 1700 F.

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corresponds with the well dispersed phosphide phase observed in similar specimens quenched directly from the mold at these temperatures. However, an increasing network formation is evident as the phosphorus is raised to 0.25 per cent, and the strengths of the plain irons would be expected to closely approach those of the alloy irons at still higher phosphorus

When the temperature is reduced to 1700 F, the strength of the austenite has increased appreciably and the fiber stresses attain much greater values before fracturing. This tends to magnify the effect of liquid or recently solidified phosphide eutectic in breaking up the continuity of the matrix and providing stress concentrations. Consequently, the phosphorus content exerts a very strong influence on the properties of all irons at this temperature, overshadowing the individual changes in phosphide eutectic properties by alloy additions. As the phosphide gains strength at still lower temperatures, it would be expected to become a continuous part of the matrix and have very little effect on the properties except for changes in the matrix structure due to impoverishment by molybdenum or chromium segregation.

A comparison of the ductility of the various irons could be obtained by computing the per cent elongations from the head speed of the tensile machine and the duration of the tests. These values are plotted against the temperatures of maximum load in Fig. 12, with the points for the 0.04 per cent, 0.16 per cent, and 0.25 per cent P irons divided into approximate regions indicated by their distributions. An increase in phosphorus from 0.04 per cent to 0.25 per cent

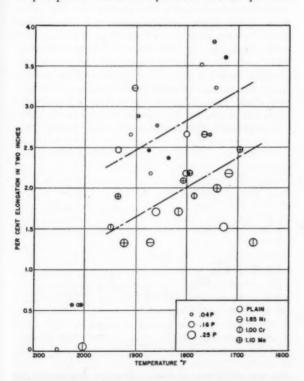


Fig. 12-Ductility of plain and alloy irons at 2050-1650 F on direct cooling from the liquid state.



Fig. 13-Photomicrograph of typical direct cooling tensile fracture. Unetched. Mag. 50x.

reduced the elongation to approximately one-half its original value in most of the irons. The molybdenum and chromium irons generally exhibited elongations near the bottom of the average regions for their phosphorus content, while nickel irons tended to produce elongations on the high side of their regions. These variations correspond with the marked changes in ductility at the points of maximum load observed in

the experimental load curves.

A photomicrograph of a typical direct cooling tensile fracture is shown in Fig. 13. In general, there were no significant differences in the appearance of the fractures from the various direct cooling tensile compositions or the preliminary reheated tensile specimens. The graphite flakes were slightly larger in the slowly cooled direct cooling specimens, but the proportion of carbide in the chromium irons and graphite distributions in the other irons were nearly identical from both types. Consequently, the marked difference in the relative strengths of the various alloy irons obtained by the two methods of testing could be attributed almost entirely to the difference in the quantity and distribution of the phosphide eutectic. Since the direct cooling tensile test reproduces the actual structure of the metal in a casting experiencing contraction stresses in the mold, the data obtained by this method can be used to compare the hot tearing tendencies of various phosphorus and alloy irons.

#### Theory of Hot Tear Formation

The formation of a hot tear can be most readily understood by tracing the steps in the cooling of a casting. Shortly after solidification, the metal begins to contract in the solid state at a rate dependent on its composition. This contraction may be resisted by either the sand in the mold or another part of the casting at a different temperature and contracting at a different rate to produce tensile stresses in the section under consideration. At temperatures just below the freezing point, the metal is in a semi-plastic condition and will deform readily under small loads without

fracture. This deformation tends to relieve the stresses as rapidly as they form, in a manner similar to the relaxation of bolts in high temperature applications. However, as the casting cools further, the metal increases in rigidity (or the plastic deformation/unit load decreases) and the stresses begin to build up in sections experiencing hindered contraction. If the deformation/unit load falls to a low value in a temperature range in which the metal is still relatively weak, the stresses may reach sufficient magnitudes to exceed its tensile strength, and a hot tear is formed.

#### Hot Tear Formation Factors

Thus, hot tear formation is a complex phenomenon involving the relationship between the following factors:

- 1. The linear contraction of the metal produced by a given decrease in temperature, or its "contraction rate." This varies with the composition and, to some extent, with the temperature range under consideration.
- 2. The amount of resistance to normal contraction provided by the mold or another part of the casting. Mold resistance depends on the hardness of ramming, the expanse of sand between two sections of the casting, and its compressive strength at high temperatures. Resistance arising from another part of the casting is produced by thermal gradients which depend on casting design, gating and risering, and pouring temperature and speed.
- 3. The deformation/unit load, which determines the rate at which stresses are built up in the casting under a given contraction rate and resistance. This factor varies considerably with temperature and a sharp decrease to a low value would define the temperature range in which hot tearing would be most likely to occur.

4. The tensile strength of the metal in the hot tearing temperature range.

The factors affecting hot tear formation in steel castings have been quite extensively investigated and reviewed in considerable detail by Briggs.3 Steel is generally considered to be most susceptible to hot tears in the temperature range 2500 to 2300 F, although some evidence has indicated hot tear formation at or near the freezing point. By hindering the contraction of solidifying bars with springs and measuring the resulting stresses, Briggs and Gezelius4,5 obtained relative values representing the combined contraction rate and deformation/unit load of various cast steel compositions in this range. These studies revealed that uniform stresses of up to 1200 lb could form at hot tearing temperatures without fracture, and it was concluded that local stress concentrations may produce stresses of much greater magnitude for hot tear formation.

By relating this data to the direct cooling tensile strengths obtained by Hall, 1,2 the relative hot tearing tendencies of various compositions could be evaluated. Although higher carbon steels exhibited lower strengths, they also experienced greater plastic deformation to result in reduced hindered contraction stresses, so that no marked change in hot tearing ten-

dencies with carbon content was indicated. Nickelchromium additions, on the other hand, reduced the hot strength appreciably without producing a noticeable change in the hindered contraction stress, thus confirming the hot tear susceptibility of these steels observed in foundry practice.

#### Hot Tear Formation in Gray Iron

Hot tear formation in gray iron follows a pattern similar to that of a very high carbon steel with two notable exceptions. First, the precipitation of graphite from saturated austenite after iron-carbon eutectic solidification produces an expansion which would relieve any external hindered contraction stresses and prevent hot tear formation in the temperature range immediately after solidification. The volume change data of Ash and Saeger^{6,7} reveals that this expansion reaches a maximum at 1900 to 1950 F and then drops off to a fairly constant linear contraction rate of 0.05 to 0.06 per cent per 100 F in medium to high carbon irons. Experience in foundry practice indicates that the contraction rate is considerably higher with lower carbon and silicon contents. Secondly, the presence of graphite flakes and liquid phosphide eutectic breaks up the continuity of the matrix to reduce the hot strength appreciably. These phases also provide local stress concentrations within the metal itself which may act as nuclei for hot tears.

The data obtained in the present investigation provided an excellent indication of the critical temperature range for hot tear formation in gray iron. It will be recalled that the slopes of the experimental load curves (Figs. 6, 7, 8, and 9) indicated a variation in deformation/unit load with temperature. The actual value of this factor could be obtained from each load curve by measuring its average slope from zero load to the change in curvature near the point of maximum load, and converting this slope to a strain/stress relationship from the head speed of the machine and the gauge length and diameter of the specimen. On plotting these values against the average test temperatures, the points for the various phosphorus and alloy compositions followed the single curve in Fig. 14.

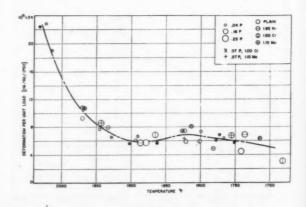


Fig. 14—Influence of temperature on deformation per unit load of gray iron on direct cooling from the liquid state.

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The independence of the deformation/unit load to variations in composition may be explained by the fact that its reciprocal corresponds with the elastic modulus in terms of room temperature tensile properties. Since the modulus of a metal is primarily dependent on the properties of the matrix, variations in the dispersed phosphide and graphite phases produced by phosphorus and alloy additions had little effect on its magnitude. This curve should be applicable to a wide range of carbon contents in gray iron because the matrix austenite is saturated with approximately the same amount of carbon at corresponding temperatures regardless of the base carbon composition. The indicated values for the "modulus" of gray iron in this temperature range are 43,000 psi at 2030 F, 172,000 psi at 1880 F, 156,000 psi at 1820 F, and 200,000 psi at 1700 F. At lower temperatures, the modulus would be expected to increase more rapidly in an approach to the room temperature value of 12-20,000,000 psi.

#### Hot Tearing Temperature Range for Gray Iron

The sharp reduction in the deformation/unit load to a minimum at 1880 F, together with the graphite expansion at temperatures above 1900 to 1950 F, defines the hot tearing temperature range in gray iron quite precisely. Above 1880 to 1900 F, plastic flow would be expected to relieve the small contraction stresses which could form after the completion of graphite expansion. Referring to Fig. 10, it can be seen that the strengths of the low and medium phosphorus irons approach relatively high values below 1800 F, while the high phosphorus irons retain relatively low strengths to 1700 F. Thus, hot tearing in the low and medium phosphorus irons might be expected to occur within the relatively short range of approximately 1900 to 1800 F, while the higher phosphorus irons could be susceptible to hot tears from 1900 to 1700 F. These temperature ranges, which lie 150 to 350 F below the solidification temperature, correspond with the hot tearing temperature range in steel, which lies 150 to 400 F below its solidification temperature. The tensile strengths in this range, varying from 1600 to 3500 psi, are also of the same magnitude as the stress values which are believed to produce hot tears in steel castings. The strengths are somewhat higher than the values of 1200 to 2500 psi reported by Hall^{1,2} for steel in the hot tearing range and this difference, coupled with the lower contraction rate of gray iron, accounts for the generally reduced hot tearing tendencies of gray iron in comparison to steel.

The effect of individual elements on the hot tear susceptibility of gray iron may now be seen from an examination of Figs. 10 and 11. An increase in phosphorus from 0.04 to 0.25 per cent reduces the strength of the alloy irons at 1800 to 1900 F by 30 to 40 per cent. It also extends the hot tearing range 100 F, thereby permitting the contraction stresses to build up to greater magnitudes while the iron is still relatively weak. In plain irons, a similar increase in phosphorus does not materially decrease the strength at 1900 F and produces a reduction of only 14 per cent at 1800 F. However, the marked strength reduc-

tion at 1700 F indicates a considerably greater hot tearing tendency for higher phosphorus plain irons under high contraction stresses.

At a phosphorus level frequently encountered in high strength gray irons, for example 0.12 per cent, the addition of 1.85 per cent nickel to a plain iron does not affect its hot tearing tendency, while 1.00 per cent chromium or 1.10 per cent molybdenum additions reduce the strength in the hot tearing range by 12 to 30 per cent. At higher phosphorus levels, the strength reductions by chromium and molybdenum increase considerably, and nickel also shows a tendency to reduce hot strength. Since balancing nickel additions do not materially change the properties of the molybdenum or chromium phosphide eutectics, the apparent increase in hot tearing tendencies due to balanced alloy additions in foundry practice may be understood in view of the drastic strength reductions produced by these segregated alloys.

The close correspondence between the strengths of the chromium irons, containing relatively small quantities of graphite, and the fully graphitized molybdenum irons indicates that graphitic carbon content assumes a very minor role in the strength of gray iron at hot tearing temperatures. Thus, since the deformation/unit load and hot strength are both quite independent of carbon content, the greater contractions experienced by low carbon irons would be expected to increase their hot tearing tendencies proportionately.

#### **Elimination of Hot Tears**

A hot tearing problem can best be solved by considering the four factors involved in hot tear formation and making the necessary changes in the factor which may eliminate the trouble. Since the deformation/unit load does not appear to vary with composition in gray irons, this factor may be immediately eliminated from consideration. When the strength specifications permit, hot tearing problems in plain irons may be frequently solved by changing to a softer grade which exhibits a lower rate of contraction. However, if this change involves a marked increase in the phosphorus content, the decrease in hot strength at 1700 F might offset the advantage of reduced contraction rates.

With fixed strength specifications, a reduction in phosphorus content appears to be the only compositional change of value to hot tear prevention. The alloying of softer grades to reduce contraction stresses may actually increase hot tearing due to the hot strength reduction by alloys and the potential increase in phosphorus from lower percentages of steel in the charge. Conversely, changing to a lower carbon and silicon content and reducing alloy additions may produce greater contraction stresses which exceed the improved hot strength. In general, the greatest security against hot tear formation from a standpoint of composition is obtained by keeping the phosphorus content at a minimum in all irons poured.

Frequently, improvements can also be made on the resistances which produce the hindered contraction stresses. Mold resistance may be reduced by decreasing the hot strength of molding or core sands. York®

has found that some sand compositions may reach compressive strengths of over 1000 psi at 2000 F. In extreme cases, relieving blocks inserted behind the facing sand and removed during final solidification have been successful in preventing hot tears from this source.

Since gray irons exhibit both an expansion and a contraction after solidification, they are susceptible to very high hindered contraction stresses from thermal gradients. It is relatively easy to visualize one section of a casting expanding at some temperature above 1950 F, while another section is attempting to contract at a slightly lower temperature. Hot tears arising from this source are frequently encountered in the spokes of flywheels and gear blanks. Occasionally, cracks may even be obtained on the surface of a uniform cylindrical section due to stresses in the cooler surface layer produced by an expansion in the center. Improvements in casting design to remove wide variations in section size are the best solution to most thermal gradient stresses. However, the foundryman can often eliminate hot tears from this source by gating into light sections, chilling heavy sections, or adjusting pouring speeds and temperatures. Hot tears are also frequently associated with internal shrinkage defects, since they provide local stress concentrations which serve as nuclei for tear formation. Heavy feeding of the affected section would be indicated in such cases.

#### Summary

1. A procedure has been developed for tensile testing gray iron on direct cooling from the liquid state in a sand mold at normal rates of strain. This method provides a means of measuring the properties of gray iron shortly after solidification under actual hot tearing conditions.

2. Direct cooling tensile tests at 2050 to 1650 F on a 2.90 per cent T.C., 2.10 per cent Si iron provided hot strength and plastic deformation data for the determination of the hot tearing temperature range in gray iron. Gray irons containing 0.04 to 0.16 per cent phosphorus were found to be most susceptible to hot tears in the range 1900 to 1800 F, while 0.25 per cent phosphorus irons were subject to hot tearing over the extended range 1900 to 1700 F.

3. The tensile strength of gray iron in the hot tearing temperature range varies from 1600 to 3500 psi, indicating that stress concentrations must reach values of this magnitude for hot tear formation. The stress strain "modulus" of gray iron increases from approximately 43,000 psi to 2000 F to an average value of 170,000 psi in the temperature range 1900 to 1700 F, with little or no dependence on composition.

4. The liquid phosphide eutectic in gray iron greatly increases its hot tearing tendency by reducing the strength in the hot tearing temperature range. This is particularly true for alloyed irons, in which

an increase in phosphorus from 0.04 to 0.16 per cent may produce a hot strength reduction of over 30 per cent.

5. Individual or balanced additions containing molybdenum or chromium sharply reduce the hot strength at phosphorus levels as low as 0.10 per cent by segregating to the phosphide eutectic to increase its volume. Nickel also decreases the hot strength to some extent at phosphorus levels above 0.12 per cent.

6. Low carbon irons would be expected to exhibit greater hot tearing tendencies because they produce greater contraction stresses with very little apparent change in the tensile or plastic properties in the hot tearing range.

7. With fixed strength specifications, a reduction in base phosphorus content appears to be the only compositional change of positive value in the prevention of hot tears. The alloying of softer grades will reduce contraction stresses but may also reduce hot strength proportionally, while a reduction in alloy requirements obtained by lower carbon and silicon contents may increase contraction stresses beyond the improved hot strength.

## Acknowledgment

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#### Bibliography

- 1. H. Hall, "The Strength and Ductility of Cast Steel During Cooling From the Liquid State in Sand Molds, Part I," Iron and Steel Institute (London) Special Report 15, pp. 65-93 (1936).
- and Steel Institute (London) Special Report 15, pp. 65-93 (1936).

  2. H. Hall, "The Strength and Ductility of Cast Steel During Cooling From the Liquid State in Sand Molds, Part II," Iron and Steel Institute (London) Special Report 23, pp. 73-86 (1938).
- and Steel Institute (London) Special Report 23, pp. 73-86 (1938).

  3. C. W. Briggs, "Hot-Tear Formation," The Metallurgy of Steel Castings, McGraw-Hill Book Co., Inc., pp. 317-337 (1946).
- 4. C. W. Briggs and R. Gezelius, "Studies on Solidification and Contraction and Their Relation to the Formation of Hot Tears in Steel Castings," Transactions, American Foundrymen's Association, vol. 41, pp. 385–448 (1933).
- 5. C. W. Briggs and R. Gezelius, "Studies on Solidification and Contraction in Steel Castings, Part II—Free and Hindered Contraction of Cast Carbon Steel," Transactions, American Foundrymen's Association, vol. 42, pp. 449-476 (1934); "Part IV—The Free and Hindered Contraction of Alloyed Cast Iron Steels," Transactions, American Foundrymen's Association, vol. 44, pp. 1-32 (1936).
- 6. E. J. Ash and C. M. Saeger, Jr., "Methods for Determining the Volume Changes Undergone by Metals and Alloys During Casting," TRANSACTIONS, American Foundrymen's Association, vol. 38, pp. 107 (1930).
- 7. E. J. Ash and C. M. Saeger, Jr., "Volume Changes of Cast Irons During Casting," Transactions, American Foundrymen's Association, vol. 40, pp. 172-200 (1932).
- Association, vol. 40, pp. 172-200 (1932).

  8. H. L. York, "Report of Progress of Sand Research on Steel Sand Mixtures at Elevated Temperatures," Transactions, American Foundrymen's Association, vol. 47, pp. 805-830 (1939).

# FOUNDRY MANAGEMENT LOOKS AT STATISTICAL QUALITY CONTROL

By

E. L. Fay*

#### ABSTRACT

In presenting the case for Statistical Quality Control the writer endeavors, first to point out its value to the foundry manager in controlling and increasing quality, in the reduction of scrap and salvage costs, and in the indirect and secondary benefits to be derived from such a program. Second, he discusses several of the fundamental steps in setting up a successful quality control program. And third, he points out briefly, some of its results. The writer discusses these topics under the following headings:

1. Why Should Management Look at Statistical Quality Con-

trol?

2. How Can Such a Program Be Set Up?

3. What Are the Results?

4. Summary and Conclusions

#### 1. Why Should Management Look at Statistical **Quality Control?**

Any discussion that could, logically, be offered under this subheading must, of necessity, present some of the atttributes and advantages of a sound, well-functioning Statistical Quality Control program. In other words, when we examine this subject in the light of management responsibility and management activity, does Statistical Quality Control offer any advantages that would, or could, influence its adoption

Some of the broader goals which the foundry manager is interested in are a reduction in scrap castings, reduced salvage costs, increased efficiency, and surely, a higher quality casting at a cost that is, at least, commensurate with existing competitive prices. To the foundry manager there are, without question, other aims and objectives besides these; but, for the present we are satisfied to enumerate only those things that are also the aims and objectives of Statistical Quality Control. Hence, to begin with, we find that the sights of management and the sights of quality control are trained upon the same target.

Statistical Quality Control is a function used to present accurate information about the quality of any specific process. It is a function of industry because, by its very nature, it is inevitably linked with the same ideals and objectives as industry. Not only does

it have these things in common, but it is readily applicable to foundry processes and, as a matter of fact, is being used today on an ever increasing scale. The accumulated data and the control charts portray a process as it actually is, without the distorting, and often inaccurate, influences of personal opinions, or what somebody else thinks about a job or process. Not only is the control chart a picture of the quality of a given process, it presents the information while the process is in operation; not at some later time. If bad castings are being made, it is much more economical to make a correction during operations rather than waiting until all of the castings are made, and then having the trouble discovered two or three days later in the machine shops.

For the reasons given above, a successful quality control program can aid materially in reducing the amount of salvage. There is no point in paying out good money to produce a run of bad castings, and then spending more good money to rework the bad castings into mediocre ones. If errors can be stopped before they occur, or at the time they begin, we are taking a long step toward attaining one of our major objectives. And Statistical Quality Control contains the

necessary potential for doing this.

A second objective of both foundry management and Statistical Quality Control is to increase the efficiency of the personnel, as well as to increase the efficiency of and cooperation between the various de-

partments.

First, a reduction in the amount of scrap means a corresponding increase in the capacity of existing cupolas, coremaking machines, core ovens, molding equipment, etc. It also means a reduction in the cost of materials and overhead. A reduction in scrap contributes toward a better flow of material; which means that production schedules can be met more easily and with more assurance.

Second, in foundries where a successful Statistical Quality Control program is in operation, there is a marked tendency toward a better spirit of cooperation between foremen, and between employees of different departments. The writer is not inferring that cooperation does not, or cannot, exist where there is

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Fig. 1-Coremaking

no quality control; but he does mean that increased cooperation is certainly a by-product of Statistical Quality Control; and there is considerable evidence to support this opinion in foundries where quality control is now functioning.

The problems of the molding foreman are many times the problems of the coremaker, the sand muller, or the pattern shop. Any influence that helps to unite these functions for the solution of common problems is another step toward this second objective of increased efficiency. Accumulated control charts form a record of past performances. It is therefore possible to go back a few days, a week, or a month, and know definitely the conditions of the variables that might contribute to a run of bad castings today. The control charts are objective evidence as to what actually occurred in the past. Therefore, with a precise and accurate knowledge of the conditions that may be responsible for a run of bad castings, it becomes less difficult to locate the causes of existing trouble.

#### **Charts Beget Quality Consciousness**

When control charts are maintained on the molding floor, at coremaking benches, and at other work stations, there is a definite predisposition for the employees to be more quality conscious, or quality minded. And, instead of the word "quality" meaning something intangible, a mere expression of somebody's opinion as to whether or not a particular job is good enough to get by, it becomes something definite and real. Control charts enable employees to "see" the quality of the cores and the molds; and, hence, they take a certain amount of pride in a chart which indicates that they are doing a good job. Improvement, perhaps unconsciously, becomes an aim of the individual as well as an aim of management.

The fourth objective, that of producing a higher quality casting at a competitive, or better than competitive, price, has already been indirectly touched upon in some of the previous discussion. That is, we know that if the expense items of scrap and salvage can show a net reduction, along with the subsequent reduction in the items of labor costs, materials, and overhead, we are going to be more able to meet, or better, existing competitive prices.

When Industry is confronted with either the desire to, or the necessity for, improving quality, the challenge is too often met with the defeatist attitude that increased quality means increased costs. The writer wishes to go on record as saying that this axiom is not necessarily true. As a matter of fact, we can make a better product, and do it for less cost. But, it means work. It means that a complete study of the process must be made, that a careful analysis of all of the variables must be carried out, and that all of the accumulated data must be accurate and unbiased. When all of the pertinent information has been gathered together, assuming now that it is accurate; then, and only then, can accurate conclusions be drawn, and the correct decisions be made. The tools of Statistical Quality Control are facilities for gathering and presenting accurate information.

#### **Taking Corrective Action**

Some little time has been spent in pointing out a few of the advantages that obtain in a well-functioning quality control program. At the expense of being contradictory and perhaps, for the moment, confusing, the writer wishes to say that Statistical Quality Control, in itself, has never corrected one solitary thing either in the foundry or in the machine shops. To repeat, Statistical Quality Control never corrects anything. The improvement in quality is made by the people directly concerned with the making of that product: operators, foremen, and methods and design engineers. Quality Control, when properly used, presents a reliable picture of process quality. But all of the control charts in existence are valueless if corrective action is not taken at the time it is indicated on the control charts. The taking of corrective action is the responsibility of those people directly concerned with the making of that casting: operators, foremen, and engineers.

The techniques of Statistical Quality Control can be used to present management with manufacturing information on costs, ratio of costs, ratio of manhours to tonnage, overhead, or any other comparable items. The methods are concise, and yet comprehensive enough so that a clear picture of operating ef-

ficiency can be obtained.



Fig. 2—Gating castings

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In summarizing the first part of this paper the writer believes that foundry managers who are not using Statistical Quality Control at the present time should investigate the possibilities and advantages that accrue through the proper use of this accurate and highly adaptable function.

#### 2. How Can Such a Program Be Set Up?

When considering the adoption of a quality control program in a foundry, the manager might reasonably ask himself this question: "What am I going to let myself in for?" And, certainly, since he is responsible for maintaining an efficient and well-run organization, he has ample justification for making this, or some similar inquiry. Statistical texts, control charts, histograms, and distribution curves might spell out the everyday working tools of a quality control man, but to the uninitiated they add up into a very confusing picture. Without having a simple and clearcut plan in mind for the adoption of a statistical quality control program, the foundry manager may feel that the results are not worthy of the efforts required. Therefore, under the second section of this paper, let us discuss some of the elements of just such a simple and clear-cut method for installing this extremely useful tool.

It would be an impossible task to write up a plan, or plans, for a foundry quality control program that would cover all of the specific problems that might arise in all of the different foundries that are in operation. However, there are a few general fundamentals that form the framework for a successful plan; and from this framework, then, one can build the details that best fit his own specific needs and problems.

The writer makes no claim to being an authority on foundry operations and methods. But from his own modest experience, from having spent much time in visiting many foundries and in talking with foundry managers and foundry quality control people, there are certain ideas that have crystallized and taken form. Not all of the applications that he has witnessed have been successful, by any means. But even from the failures one can learn about some of the pitfalls to avoid. The following generalities represent the essence of these experiences and are, hence, representative of the experiences of many people under many different conditions.

One of the first steps, if not the very first, in developing a foundry statistical quality control program is to promote and carry through with an educational program for management. In using the word "management" the author wishes to divide the concept into two parts; (a) top management consisting of the executive group of administrating officers, and; (b) operating management consisting of superintendents, general foremen, foremen, and department heads who are responsible for the actual operations.

Someone may say that the first step should be the decision as to whether or not Statistical Quality Control should be used, and whoever voices such an opinion is absolutely correct. But how can top management make a valid decision unless it is at least



Fig. 3-Grinding castings

partially familiar with the basic fundamentals? Hence, in the writer's opinion, the first step should be an educational program designed for top management.

In getting together the material for such a course it need be neither as long nor as extensive as a similar course of study designed for operating management. In fact, it very probably should not be. The material and its presentation should be in the nature of an "introduction" to Statistical Quality Control, showing how it operates, illustrating the various tools, and working out one or more problems in class. Of course, there should be ample time for discussion and questions.

#### Training Operating Management

The next step should initiate a training program for operating management. These are the people who will come in direct contact with quality control, the people who will use the charts, and who should take the necessary corrective action when the control charts so indicate. The course should cover the fundamentals of Statistical Quality Control, and the training in the making and interpretation of control charts.

It is important that the departmental foremen have a good working knowledge of control charts not only because they can be used as a guide for heading off or correcting trouble, or for improving quality, but for a very good psychological reason as well. To the coremaker, to the molder, to the iron pourer, to the laborer, whoever he is and whatever he may be doing, his foreman is, in reality, the company. If he has a good foreman, it is a good company to work for. If he has a bad foreman, then it becomes a bad company. The opinions and attitudes of the foreman, regarding company policies, cannot help but influence the opinions and attitudes of the worker. If the foreman believes in quality control and realizes its advantages, then the worker will more readily accept it; and be the quicker to realize how the control charts will help him to do a better job.

The reader has probably surmised by now that the basis of a successful quality control program is a sell-

ing proposition, a cooperative endeavor. Company bulletins or company directives ordering that quality control be set up on all operations in a certain department, or departments, would probably produce a lot of nice-looking charts. And to the casual observer going through these departments it might look like a pretty good set-up. But if everyone from top management on down to the fellow on the job is not sold on the advantages to be obtained from quality control, then the charts have no value whatsoever. Statistical Quality Control is a vital and important aid to the operator, foreman, general foreman, etc. It is this group of people who stand to profit from the control charts through an improvement in quality and operating efficiency. If they are not sold on it, and if they do not take corrective measures when the control chart indicates that corrective measures should be taken, then the charts are a waste of time, and might better be dispensed with entirely. Therefore, a training program for supervisory personnel is of prime importance since it is upon the shoulders of these people that the responsibility for quality falls.

#### Training Program for Inspectors

Of course there must be a training program for the inspectors who are going to check the parts, post the charts, and carry out many of the other details for a successful program. The length of this training program for inspectors and inspection supervision may vary considerably from one plant to the next. The writer has conducted a number of schools in various plants and has had some success holding classes two hours a day for ten or twelve days. One of the advantages in holding classes for only two hours per day is that the supervisor or inspector is not taken from his job for any protracted periods. Also, by spreading the school out over a longer period of time he can do a better job of learning than in cases where the school would run all day long but for only three or four days. However, these details must be worked out to fit the conditions and needs as they exist in any particular foundry.

In order that we may continue to the next step of

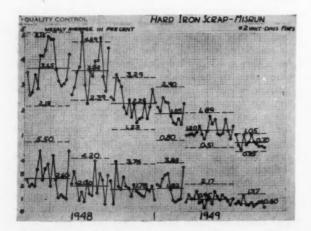


Fig. 4—Weekly average of per cent of hard iron scrap due to misruns for years 1948-1949

our plan for installing a quality control program, let us assume that we have completed this first phase of our training and educational program. As a matter of fact, the training and educational work is never really completed in its entirety. The point is never reached where the quality control director can say, "From now on, we will never have occasion to do any more training, or to hold any more schools."

It is now time in our plan to apply quality control to some process or production operation. How shall we go about it? Shall we start it on a dozen jobs at the same time? Should our beginning activities be confined to only one department, or should they start in all departments at once? These are some of the questions that will arise when a quality control program is initiated.

Sometime before installing the first application of quality control to a process, the question always arises as to how much new equipment is going to be required to carry out this function. It is easy enough to say, "We will need new tools, new measuring devices, and new this and new that." As a result, the program is either dropped or the equipment is purchased; and in three months time it is found to be inadequate, or, entirely unnecessary. Certainly, in some cases, after a thorough investigation has been made, and some experience has been gained in quality control applications, it will no doubt be necessary to obtain some additional equipment in order to do a better job, and do it faster and more efficiently. But let us not do it until a sufficient study has been made on the job, and until we have all of the facts pertaining to what is actually needed. We should begin with the tools and the equipment that we have, building our program around existing facilities, and showing management with facts that quality control can aid in doing a better job with the same tools. If this is carried out, management will be much more responsive to requests for additional tools when the needs arise.

#### Commencing a Quality Control Program

In making this first application of quality control to a process in the foundry, machine shop, or wherever it is, there is some important ground work to be laid. The first thing is to proceed slowly. The quality control director, the foreman in whose department the first installation is to be made, the general foreman and perhaps the foundry superintendent should get together and select one specific job, or one process. The details should be carefully reviewed for the operator, explaining to him the general plan, how often the job will be inspected, what the control chart is, how it functions, and other necessary information so that he becomes familiar with it. It is human nature to shy away from things that are strange, and to resist anything that might involve a change in habits or custom. If that first operator understands how the control chart operates, and how it is going to help him, a big step forward has been taken. He appreciates being taken into the confidence of management, and he can go a long way in selling quality control to the men he works with. .

The type of job that this first application is used on is not too important, except that if quality control could straighten out a job that has been consistently causing trouble, it would add a strong sales argument. After this first job is definitely on the way toward improvement, then apply quality control to another process, and another. In every case, be sure that the operator knows what this is all about, and that he understands how to use the control chart, and how the inspector uses it.

Without further elaboration or additional emphasis, the person who steps into the job as quality control director must have a good background of foundry experience, quality control training, and some knowledge of statistics and statistical applications. This point has not been touched upon previously since the writer has slanted his efforts toward the foundry manager rather than specific qualifications

for quality control personnel.

In concluding the second phase of this paper the writer would like to re-emphasize that Statistical Quality Control, when correctly used, is a valuable production function. If management is interested in attaining the goals of costs reduction, increased quality, and a more efficient structure through the application of quality control, then it must become an established policy that operating management take corrective action on a process when the control chart indicates that corrective action should be taken.

#### 3. What Are the Results?

The results of Statistical Quality Control may be measured, first, by the criterion of general acceptance, and, second, from personal experience.

Since this statistical tool, in its simplified form, was first introduced to industry about ten years ago, it has witnessed a remarkable growth not only in the quantity of applications but also in the variety. Ford, General Motors, Republic Steel, Bristol-Meyers Laboratories, Western Electric, and Bell Telephone are only several of many of the large industries in this country which are using quality control on an extensive basis; and which are finding it a necessary and important adjunct in solving problems of quality. Its expansion has been no less phenomenal in the field of smaller industries and business. In the five years of its existence, the American Society for Quality Control has grown to over 4000 members. Many engineering schools have added courses in quality control to their curriculum in order to meet the demands of industry for professionally trained personnel.

The Department of Military Defense of the United States has accepted Statistical Quality Control as a means for measuring the quality of purchased material. Many war contracts stipulate that the product shall be made using quality control applications on the various manufacturing processes.

The statistical basis for quality control is neither of recent origin nor does it involve higher mathematics. It is based upon old established principles that are easily understood and readily adaptable to foundry operations.

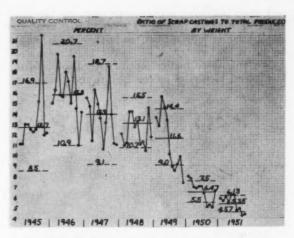


Fig. 5—Ratio of scrap castings to total produced, by weight for years 1945 through first 10 months of 1951.

From the standpoint of personal experience space will permit only a couple of illustrations. Figure 4 is a control chart showing the per cent of hard iron scrap due to misruns, for the years 1948 and 1949. Each plotting represents a weekly value. It may be observed that there are two graphs on the chart; and without going into the details of the construction of a control chart we shall refer to the top graph only. The solid lines through the center of each group of plottings represent the average per cent of misrun scrap for those particular weeks. For example, during the first 4-month period the overall average was 3.65 per cent. The average per cent scrap for the next 4-month period was 3.54. After statistical quality control was installed in the foundry the amount of scrap dropped to 2.26. And by the end of 1949, the average was down to 0.7 of one per cent. Since that time it has leveled off to a figure somewhat below this.

Figure 5 is a chart showing the ratio of scrap castings to total production in a malleable foundry for the years 1945 through the first ten months of 1951. Each plotting represents a monthly value. Again referring to the solid lines which represent the yearly average values, we note that in 1946 the average for the year was 15.8 per cent. Statistical Quality Control was installed in 1947. Since that time the ratio of scrap to total production has decreased to 5.35 per cent for this past year. And, from an observation of the last four plottings there is every indication of a further reduction for 1952. So much for the objective results.

These illustrations are only two of the many hundreds of cases with which the writer is familiar either from his own experience or from having studied them in other foundries and factories.

#### 4. Summary and Conclusions

Statistical Quality Control is becoming an important function in industry because it not only has the same goals as industry, but because it possesses a vital potential for assisting in attaining these goals. Its possible economic advantages become active and

real when operating management takes the necessary action to head off trouble, or to correct it, when the control charts indicate that corrective action should be taken.

Not only does quality control give a mathematically correct picture of process quality and process var-

iability, but it is easy to understand and simple in its application. It can be installed without disrupting current routine and schedules. Statistical Quality Control has been accepted by industry as the practical approach to the related problems of quality, and production costs.

# ZINC IN ALUMINUM CASTING ALLOYS

By

Donald L. Colwell*

#### ABSTRACT

Zinc, like silicon, was once considered an undesirable impurity in aluminum casting alloys. Data are presented showing that in nominal quantities in excess of most current specifications zinc has a beneficial effect on mechanical properties and machinability and has no adverse effects on castability, corrosion resistance, or other physical properties. Emphasis is placed on the common copper-silicon casting alloys both in the as-cast and heat-treated conditions. Mention is also made of the danger in magnesium as an impurity, particularly in alloy CS64C and also of the neutral effects of zinc on alloy S5C. Numerous examples are given in the accompanying graphs.

Prior to World War I the general purpose aluminum casting alloy used in this country was known as No. 12 aluminum (CS72A†, Table 1). It was used for sand, permanent mold, and die castings. Originally both silicon and zinc were usually limited, often to a maximum of one per cent. With the advent of the Alpax and Silumin types of aluminum alloy immediately after World War I, foundrymen soon learned that silicon was not a harmful impurity, that on the contrary it has a highly desirable effect on casting properties, particularly in permanent mold and die castings. Consequently silicon was deliberately added to the alloy, and the amounts increased gradually until today the maximum silicon limit specified by the ASTM in alloy CS72A (No. 12) is 4.0 per cent.

Zinc, however, was still considered undesirable. The early CS72A ingot was unavoidably contaminated with zinc, particularly from obsolete scrap, and largely for commercial reasons the zinc maximum in alloy CS72A was allowed to advance to the present 2.5 per cent where it stands today.

During World War II the use of aluminum castings expanded tremendously. With shortages of both aluminum and copper, the Casting Industry and the War Production Board found that alloys with about 4 per cent copper and with the silicon varying from

3 per cent in sand casting alloys to 6 per cent in permanent mold casting alloys and 9 per cent in die casting alloys not only were easier to obtain than CS72A, but also had superior casting properties and were no more expensive. This trend has continued, and the copper-silicon type of alloy is the one most widely used today. The popular general purpose alloys for all three types of castings are very close to a 3 to 4 per cent copper content with varying amounts of silicon.

Popular copper-silicon (or silicon-copper) specifications are listed in Table 1. Small amounts of manganese have been found to be beneficial and even when manganese is not present in the scrap it is usually added to help in controlling the iron. It should not exceed about 0.5 per cent maximum. Magnesium has been found to be harmful to toughness except in certain cases where castings are to be heat treated and elongation is not required. It is now removed by the specification smelters, and, with one exception, which will be discussed later, is limited as an impurity to a maximum of 0.1 per cent or less. Zinc limits in many specifications early in the war were increased from 0.1 per cent to about 0.3 per cent but remained low because zinc was not a factor in the duralumin type of scrap and some had doubts about its effect. Several investigations were undertaken in an endeavor to determine the effects of zinc on these popular casting alloys.

In one of these investigations test bars were cast of CS43A of a high purity grade and of a secondary grade both in sand and in a permanent mold. The results of mechanical tests on these bars are plotted in Fig. 1. Increases in iron and manganese and perhaps zinc had a slight effect in increasing tensile and yield strengths and Brinell hardness and in decreasing elongation in the chill cast bars (a slight increase in elongation in the sand cast bars was probably due to testing variables, as normally the elongation decreases with decreasing amounts of aluminum). In the absence of appreciable magnesium there was no difficulty in obtaining elongation in excess of 2 per cent. These results are typical of the properties in the as-cast and in the aged condition, and a similar relationship was found in cases of exposure and ac-

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^{*} Director of Laboratories, Apex Smelting Co., Chicago.

[†]In the ASTM nomenclature system, the nominal composition is indicated by the alloy number. Thus CS72A denotes a nominal composition of copper 7 per cent, silicon 2 per cent; CS43A denotes copper 4 per cent, silicon 3 per cent; SC84A denotes silicon 8 per cent, copper 4 per cent; ZG32A denotes zinc 3 per cent, magnesium 2 per cent, etc. (Table 1).

TABLE 1-COMPOSITION LIMITS OF SELECTED ASTM ALUMINUM CASTING ALLOYS1

Alloy	Cu	Fe	Si	Mn	Mg	Zn	Cr	Ti	Ni	Sn	Other Each	Elements Total
CS43A	3.5— 4.5	1.0	2.5— 3.5	0.5	0.05	1.0	_	0.2	0.3	_	_	0.5
CS72A	6.0— 8.0	1.2	1.0— 4.0	0.5	0.07	2.5		0.2	0.3	-	_	0.5
S5C (S4)	0.6	1.0	4.5— 6.0	0.3	0.1	0.3	-	-	0.5	0.1	_	0.2
S12A (S5, S9)	0.6	0.8	11.0— 13.0	0.3	0.1	0.3	_	_	0.5	0.1	_	0.2
SC51A	1.0— 1.5	0.6*	4.5— 5.5	0.5 •	0.4—	0.3	0.2	0.2	_	_	0.05	0.15
SC64B	3.3— 4.3	0.8	5.5— 7.0	0.5	0.1	1.0	_	0.2	0.3	_	_	0.5
SC64C	3.0— 4.5	1.0	5.5— 7.0	0.8	0.5	1.0	_	0.2	0.5	_	_	0.5
SC84A-B (SC6, 7)	3.0— 4.0	1.0	7.5— 9.5	0.5	0.1	0.9	_	_	0.5	0.3	_	0.5
ZC60A	0.35— 0.65	1.3	0.3	0.05	0.25— 0.45	6.0— 7.0	-	0.2	_		0.05	0.15
ZC81A-B	0.4— 1.0	0.9	0.25	0.6	0.2— 0.5	7.0— 8.0	0.3	0.2	0.1	-	0.05	0.2
2G32A	0.2	0.6	0.2	0.4— 0.6	1.4— 1.8	2.7— 3.3	0.2—	0.2	_	-	0.05	_
ZG42A	0.2	0.6	0.2	0.4— 0.6	1.8— 2.4	4.0— 4.5	0.2— 0.4	0.2	_	_	0.05	_
ZG61A	0.3	0.8	0.25	0.3	0.5— 0.65	5.2— 6.0	0.4— 0.6	0.1— 0.25	_	_	0.05	0.2
ZG61B	0.35— 0.65	0.4	0.15	0.05	0.6— 0.8	6.0— 7.0	_	0.2	_	_	0.05	0.15

^{*} If the iron content exceeds 0.4 per cent, it is desirable to have manganese present in an amount equal to one-half of the iron.

celerated corrosion.

As a result of this and numerous other similar tests made by many laboratories, the maximum limit on zinc in current specifications for most copper-silicon casting alloys has been increased to about 1 per cent. The 0.3 per cent limit for zinc in SC51A is discussed later.

There has been a decided post war trend in the aircraft industry and even, to some extent, in the foundries, to use a new type of aluminum alloy which develops unusually high strengths without heat treatment, and, when heat treated, develops strengths even higher than the earlier duralumin types. These alloys are of a zinc-magnesium-chromium type and are illustrated in the wrought condition by ZG62A with a nominal composition of 5.6 per cent zinc, 2.5 per cent magnesium, 1.6 per cent copper, and 0.3 per cent chromium. Six similar compositions used by the Foundry Industry are specified in the ZC and ZG series listed in Table 1. Scrap from such combinations is already showing up in supposedly zinc-free material, and with increasing aircraft production zinc will become even more difficult to hold to low limits in casting alloys than it is today. A further discussion of the effects of zinc in larger quantities on the copper-silicon alloys seems advisable.

Another typical test concentrated on composition CS43A. A base heat of the secondary composition previously tested was made; its analysis is shown in

Table 2—Analysis of Base Heat of CS43A Used in Testing Effects of Zinc from 0.4 to 2.0% (Figs. 2, 3, 4, and 5).

Element	Content, %
 Cu	4.20
Si	3.30
Fe	0.77
Zn	0.43
Mn	0.31
Mg	0.01
Ni	0.29
Pb	0.09
Sn	0.07
Cr	0.04
Ti	0.02
Al	Balance

Table 2. Zinc was added to this base to produce heats containing 1.0, 1.5, 1.75, and 2.0 per cent. Mechanical property tests in the as-cast and aged conditions were made and results are plotted in Figs. 2 and 3, sand cast and chill cast.

The sand cast and chill cast test bars were also given T5, T6, and T7 heat treatments and subjected to the same tests, the results of which are plotted in Figs. 4 and 5. The T5 treatment consisted of heating at 450 F for 8 hr; the T6 treatment consisted of heating at 940 F for 12 hr followed by a

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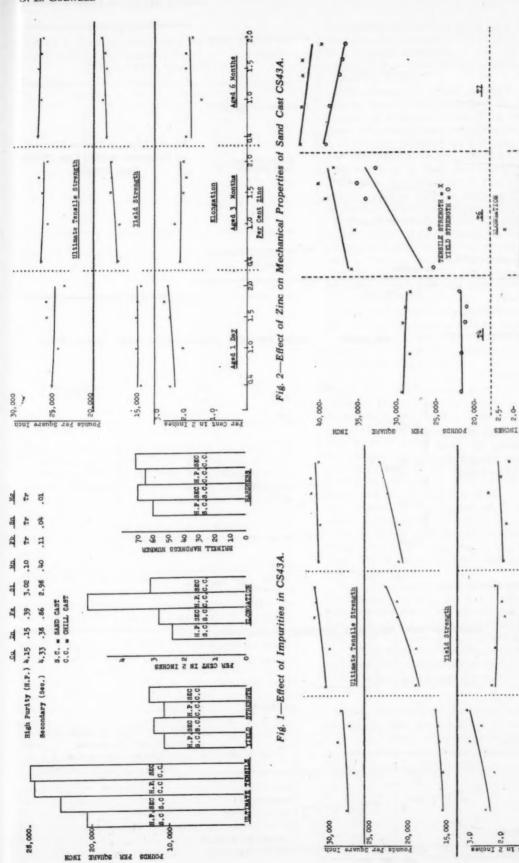


Fig. 3-Effect of Zinc on Mechanical Properties of Chill Cast CS43A.

Aged 3 Months 1.0 1.5 Per Cent 2inc

Flongation

Fig. 4-Effect of Zinc on Mechanical Properties of Sand Cast CS43A Heat Treated.

2.0 0.2

2.0 0.4 1.0 PER CENT

1.5

1.0

0.5 0.4

N 1.5-MI %

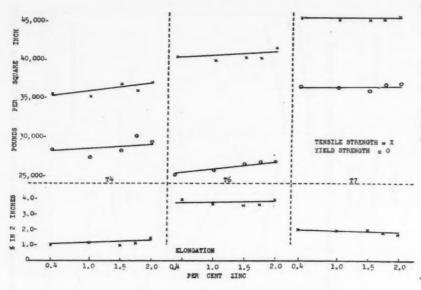
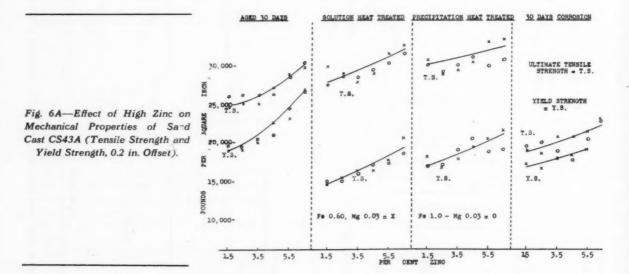


Fig. 5—Effect of Zinc on Mechanical Properties of Chill Cast CS43A
Heat Treated.



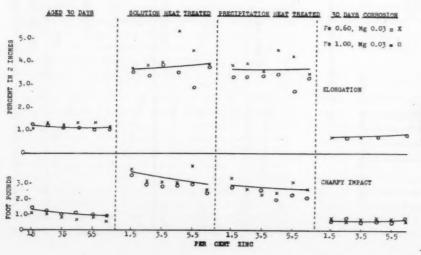


Fig. 6B—Effect of High Zinc on Mechanical Properties of Sand Cast CS43A (Elongation in 2 in. and Charpy Impact Strength).

Table 3—Composition Range of Base Heats of CS43A Used in Testing Effects of Zinc from 1.5% to 6.5% (Figs. 6A-B)

Element	Content,	Content, %	
Cu	3.95 -	4.35	
Si	2.95 -	3.15	
Mn	0.35 -	0.45	
Fe	0.6 -	1.0	
Ni	0.15 -	0.25	
Sn	0.01 ma	aximum	
Pb .	0.05	20	
Mg	0.03	0.0	
Ti	0.04	99	
Cr	0.05	99	
Al	Balance		

Table 4—Analysis of Base Heat of SC84A-B (SC6, 7)
Used in Testing Effects of Zinc on Die Castings
(Fig. 7)

3.50 8.20 0.45 0.40
0.45
0.40
0.10
0.02
0.20
0.01
0.20
Balance

hot water quench and a 3-hr aging at 300 F, and the T7 treatment consisted of the same quench followed by stabilizing 8 hr at 440 F. It is apparent from the data plotted in Figs. 2, 3, 4, and 5 that the mechanical properties, if changed at all, were changed for the better as zinc was increased up to 2 per cent; and this was true whether or not the castings were heat treated and also before and after any usual age hardening.

In this test also the various compositions were checked for electrical conductivity, thermal conductivity, specific gravity, and thermal expansion, and no significant difference could be found between the low zinc and the high zinc compositions. Even the specific gravity varied only from 2.82 to 2.84. The solidification range in all cases was between 1140 to 1145 F and 945 to 955 F which would point to no difference in "setting up" properties in casting.

A further test was made on the same alloy with varying zinc contents up to 6.5 per cent with the same results. The compositions of the base alloys used in this test are shown in Table 3. Bars had two different iron contents, one with nominally 0.6 per cent and the other 1.0 per cent. Zinc was added to produce 1.5, 2.5, 3.5, 4.5, 5.5, and 6.5 per cent alloys, and again mechanical tests were made at each zinc level. The bars were cast in sand and tested after 30 days' room temperature aging and in the T4 and T6 conditions, also after a thirty-day corrosion test. The T4 heat treatment consisting of heating 16 hr at 940 F, followed by a hot water quench and tested within 24 hr. The T6 condition consisted of the

same quench followed by aging for 3 hr at 310 F. The corrosion test consisted of total immersion for 10 sec in a solution containing 5 per cent hydrogen peroxide and 3 per cent sodium chloride, repeated every 15 min for 30 days.

The changes in tensile strength, yield strength, Charpy impact, and elongation are plotted in Figs. 6A and 6B with indications of the two iron compositions. The tensile and yield strengths increased sharply as the zinc exceeded 2.5 per cent, but the elongation and Charpy impact strength fell very slightly. This was true in both the as-cast and heat-treated conditions. It seems evident that even with a zinc content in this alloy as high as 6.5 per cent the mechanical properties do not suffer.

It has often been said that the presence of zinc in copper-silicon alloys of aluminum decreases their resistance to corrosion. In the corrosion test on these bars of varying zinc content the ultimate tensile strength dropped when corroded, but the yield strength did not, and they both were apparently improved with the higher zinc. The Charpy impact strength and elongation were lowered considerably by the severe corrosion as would be expected from such a severe test, but the bars with the higher zinc were corroded no worse and suffered no more than the bars with the lower zinc.

A further measurement of corrosion effects on CS43A with zinc contents from 1.5 to 6.5 per cent was made in the same intermittent salt and hydrogen peroxide test. The weight losses of sand-cast test bars and of chill-cast discs in the alternate immersion test are shown in Table 5, the composition of the bars being within the range shown in Table 3. The higher percentage losses of the chill-cast discs are believed to be due to the greater surface-weight ratio of that shape, and not to the chill of the casting. It is evident, therefore, that instead of decreasing the corrosion resistance of this alloy, the effect of zinc is either negligible or it slightly increases the resistance to corrosion of this type, and it throws doubt on the possibility of any significantly different reaction with other corrosive agents.

The corresponding alloy for die casting was originally developed during the war by the Ordnance Department and the War Production Board in cooperation with the Die Casting Industry Advisory Committee. It was known as AXS-679-Rev. 3, although this specification has now been superseded by Federal Specification QQ-A-591a, Composition 10. The corresponding ASTM specification is shown in Table

TABLE 5—WEIGHT LOSS AFTER ALTERNATE IMMERSION OF SC43A WITH HIGH ZINC IN HYDROGEN PEROXIDE-SALT WATER FOR 30 DAYS

Zinc Content, %	Loss on Sand Cast Test Bars, %	Loss-on-Chill Cast Discs, %
1.5	2.3	4.65
2.5	1.7	3.45
3.5	1.85	3.55
4.5	2.35	4.0
5.5	2.35	3.8
6.5	2.0	3.4

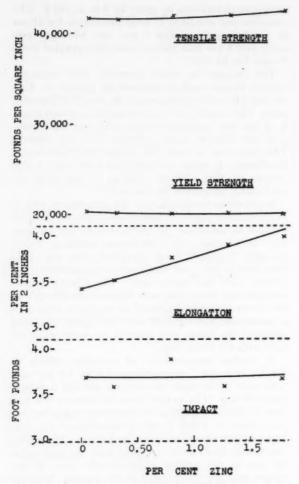


Fig. 7—Effect of Zinc on Alloy SC6(SC84A) Die Castings.²

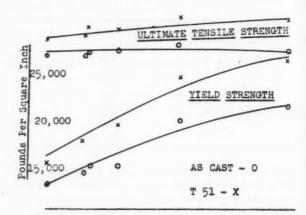
1 as alloy SC84A-B (SC6, SC7).1 The zinc allowance shown is 0.9 per cent as this is an ingot specification, but the casting limit for zinc in SC6 and SC7 is 1.0 per cent. This zinc limit was approved only after a cooperative test had been run by Committee B-6 of the ASTM, and die-cast specimens of this alloy with zinc contents of 0.25, 0.5, 1.0, 1.5, and 2.0 per cent had been tested.2 The composition of the test metal is shown in Table 4, and the results of the test are shown in Fig. 7. Each point on the curves on Fig. 7 is an average of results on five test bars made by each of three die casting companies tested by each of three laboratories, or a total of 45 individual tests. The conclusion is the same-that the only effect of the increased zinc seems to be a slight improvement in elongation.

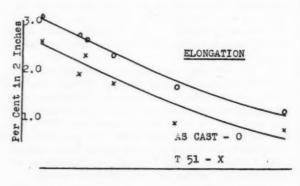
In the evaluation of the effect of zinc on these copper-silicon type alloys care must be taken to guard against magnesium contamination. In all of the above tests the magnesium contents were below 0.1 per cent, as these alloys are somewhat susceptible to age hardening and this susceptibility is tremendously magnified when the magnesium content exceeds about 0.1 per cent. This effect is graphically illustrated in Fig.

8 where the increase in tensile strength, yield strength, and hardness and the sharp decrease in elongation are demonstrated when magnesium in SC64B is increased from zero to 0.51 per cent.

A modification of SC64B is shown in Table 1 as SC64C, the only difference being the wider range of copper and the higher limits of impurities in SC64C. It is generally believed that the limits shown for SC64C are unnecessarily wide and that in the case of the maximum magnesium content particularly, this specification can result in material that tends to be brittle. Examination of the effect of zinc must be done in a manner to guard against the harmful effects of magnesium when it is allowed as high as 0.5 per cent.

The same is true of SC51A as this composition requires a magnesium content of 0.4 to 0.6 per cent to develop maximum properties when heat treated. It is usually not used in the as-cast condition, therefore,





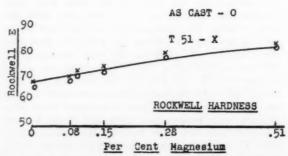


Fig. 8-Effect of Magnesium on Sand Cast SC64B.

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as the higher magnesium is likely to have the same effect as in SC64C and cause loss of elongation as the castings age. In this alloy, too, zinc is still limited to 0.3 per cent and iron to 0.6 per cent for heat treating purposes. For this reason it can be more expensive than the ordinary silicon-copper alloys and the question of higher zinc is of less importance. It seems proper to continue close specification limits on SC51A and to use it only when heat treatment is required.

Because of the purer grades of scrap necessarily used in the copper-free aluminum alloys, the need for wider zinc limits in them is not as acute as in the copper-silicon alloys. Several years ago an investigation was made by Committee B-6 of the ASTM in which die cast test bars were made of both high purity and of secondary silicon-aluminum alloy S12A-(S5) and exposed for ten years at three outdoor locations: Sandy Hook, New Jersey; New York, New York; and Altoona, Pennsylvania. Accelerated corrosion tests also were made.3 The high purity grade contained no zinc and the secondary grade 0.52 per cent zinc with higher amounts of other impurities as well. Conclusions from the several committee reports during the progress and at the conclusion of the test showed 'no significant difference in the effects of cor-

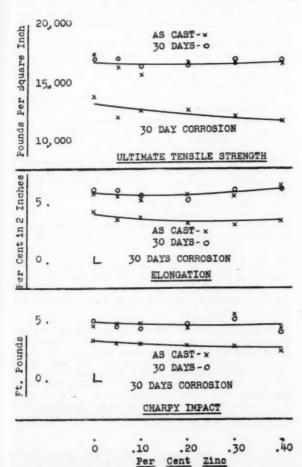


Fig. 9—Effect of Zinc on Sand Cast S5A. (Fe, 0.30; Si, 5.00; Mn, 0.02; Cu, 0.05).

rosion and exposure between the bars from the pure S12A and the bars from the secondary S12A.8

Subsequently, an individual investigation was made of the effect of a small amount of zinc on alloy S5A-(S4), and mechanical tests were made on test bars of this alloy with zinc contents ranging up to 0.40 per cent. Test results are plotted in Fig. 9. In the as-cast condition tests were made initially and after 30 days, and are so indicated in Fig. 9, although only one curve was drawn because the two sets of values were very close. These bars were also subjected to the intermittent immersion test in the hydrogen peroxide-sodium chloride mixture, and mechanical test results on the corroded test bars are also plotted in Fig. 9. All of these curves are substantially horizontal, indicating again that these nominal quantities of zinc have no effect on the material.

## Allowable Zinc Content

It can be concluded, therefore, that zinc in these copper-free silicon alloys can be safely allowed up to about one-half per cent with no harmful effects on the mechanical properties or the resistance to corrosion. Years of experience with castings of silicon-aluminum alloys lead to the same conclusion.

Mention should be made of the effect of these higher zinc contents on other properties of aluminum casting alloys. The following items were gained more from experience with hundreds of foundries and die casting companies than with specific laboratory tests, and are difficult to present in tabular form. It can be said, generally speaking, that in the copper-silicon type of alloy zinc contents up to 2 or 2.5 per cent, and in the silicon type zinc contents up to 0.5 per cent differ from their counterparts with the present lower zinc limits only as follows:

Any variation in casting properties is hardly noticeable. In certain heavy-sectioned die castings it has been found that the setup time with the higher zinc is slightly longer, but the slight increase in fluidity usually compensates production-wise. Close contact with many foundries who cast in sand and permanent molds has shown no apparent difference between castability with 1 per cent zinc and with 2 per cent zinc, other things being equal. Many foundrymen accustomed to casting No. 12 with 2 to 2.5 per cent zinc prefer the higher zinc content.

Machinability is greatly improved with higher zinc. In the author's experience with one particular permanent mold job made of an alloy containing 6 per cent copper and 4 per cent silicon the smooth facing of a flange with tools of high speed steel was practically impossible until the zinc content was raised to 2 per cent.

The cost of copper-silicon alloys and of pure silicon alloys is usually lowered when zinc limits are raised. The ZC and ZG alloys shown in Table 1, however, are necessarily more expensive because of the necessity for low limits on copper, iron, and silicon to secure the high strength; consequently their cost is not affected by their zinc content.

Further mention should be made of the ZC and ZG high strength alloys. Their increasing popularity where strength is important and heat treatment un-

desirable, is certainly evidence that the presence of zinc in aluminum castings is not in itself dangerous. Composition ZG32A, for example, has met with great success in high strength castings where high elongation is desired and where heat treatment is unnecessary. The ternary combination of aluminum with zinc and magnesium is not as readily castable as are the alloys with silicon, but the high strength together with superior corrosion resistance, greater stability, and much better machinability have in many instances more than compensated for the extra precautions required in the foundry. These combinations, like the wrought ZG62A, will no doubt become more and more popular where high strength and high elongation are desired. It should be pointed out that ZG32A, chill cast, develops a typical tensile strength of 33,000 psi with an elongation of 22 per cent without heat treatment, and the other combinations develop properties corresponding.

#### Conclusion

In conclusion, it is felt that a definite statement can be made that most of the common copper-silicon casting alloys will be improved with a zinc content approaching double the present allowable amounts. It has even been thought that like silicon, which was once considered an undesirable impurity, zinc may prove to be a highly desirable addition to these alloys when present in controlled quantities to definite spe-

The author wishes to acknowledge gratefully the assistance of the following members of his organization. Mr. Robert Raisig in preparation of the graphs. and Mr. O. Tichy, Mr. Howard R. Youngkrantz, and Mr. John J. Stobie in the experimental work.

Recognition should also be given to the work of Mr. Walter Bonsack, some of which is represented in this paper, and which is so well summarized in his Dudley Award paper in 1943.4

#### References

1. Selected from Table 1 of ASTM Specification B179-51T, "Aluminum-Base Alloys in Ingot Form for Sand Castings, Die Castings, and Permanent Mold Castings;" 1951 Supplement to Book of ASTM Standards, Part 2, Non-Ferrous Metals.

2. Minutes of the meeting of Sub-committee I, Committee B-6 on Light Metals and Alloys, American Society Testing Materials, March 1, 1950.

3. Donald L. Colwell, "Investigation of Purity of Aluminum-Silicon Die Casting Alloys," ASTM Bulletin No. 163, American

Society Testing Materials, p. 51, January, 1950.
4. Walter Bonsack, "Effects of Minor Alloying Elements in Aluminum Casting Alloys," ASTM Bulletin, American Society for Testing Materials, August, 1942 and October, 1943.

#### DISCUSSION

Chairman: W. J. KLAYER, Aluminum Industries, Inc., Cincin-

HIRAM BROWN (Written Discussion): 1 For years it was the impression that alloys of aluminum and zinc were very bad from the corrosion standpoint. Work in recent years, particularly that done on the aluminum-zinc-magnesium type of alloys, has very thoroughly disproved this point. I am happy to see that some one is also working on the effects of zinc on castability, fluidity.

I note that the author mentions that in heavy section die castings, set up time with higher zinc content is slightly longer. Is this believed to be due to the solubility of zinc and the tendency of zinc alloys to have a long freezing range of the solid solution type? If so, this brings to my mind the question of what effect the increased zinc would have in reducing the tendency of the normal copper and copper-silicon alloys to have pin hole porosity due to the presence of hydrogen in view of the relative freedom from pin hole porosity of the aluminum-zinc-magnesium alloys.

In view of the fact that many of the aluminum-zinc-magne-sium alloys are quite sensitive to the presence of silicon, copper, and manganese, I wonder what effect increased zinc would have on the normal copper or copper-silicon alloys if the copper, manganese, and silicon are on the high side of the residual range or alloying range, as the case may be. If any heats were made up approximating these conditions, was there any tendency toward embrittlement as zinc increased?

WALTER BONSACK (Written Discussion): 2 The author of this paper should be congratulated for his fine presentation of further evidence that zinc is a useful alloying element in aluminum alloys. It is especially commendable that he had enough perseverance to bring up a subject which seems to be tabu to talk about, and further that he has proven again that this element is harmless or even beneficial in aluminum alloys. I have observed this struggle of pro and con for the last 20 years. It is still going on. I will not say that it is a losing battle, but it is certainly a war of nerves, i.e., the "cold zinc war."

About 20 years ago I attended a meeting of a specification writing body in which the question of zinc was discussed at length. Poor zinc lost because not enough figures or facts were presented to prove the good effect of zinc.

After this meeting I set out to prove to myself and the metallurgical world what zinc actually would do in aluminum alloys. All possible tests were used to study its behavior. Because one of the contentions stated that zinc makes aluminum alloys hotshort, tests at elevated temperatures also were included during

The zinc content of standard foundry alloys were increased to many times their top limits. Still no harmful effect was found. The results of these tests and the analysis of test results of other people were published by me in 1942.4

In addition many very interesting facts were found. For example, adding zinc to CN42A alloy increased not only the strength and yield strength and hardness without diminishing the elongation noticeably, but this study also led to a series of alloys ZG32A and 42A and others.*

Additions of zinc to silicon alloys not only improves castability but also increases many times their machineability. These facts were used during World War II to great advantages in our emergency. In fact, alloys of 7 Cu, 3 Si with additions of zinc up to 5 per cent performed very useful purposes without any report to the contrary

Shortly after the war when a conglomerate of aluminum alloys came back in the form of war scrap, compositions of aluminum allovs sprang up with such wide limits in all elements-including the harmful ones-that even the proponents of the harmless element zinc were flabbergasted.

One of the alloys coming back is the strongest wrought alumi-num alloy which contains 5-6 per cent zinc. This alloy was used extensively in aircraft construction at home and abroad. It is of the same family discussed in the paper by me.* Nevertheless, the "cold war" against zinc generally continued.

In summary, the findings of the author and myself show that additions of zinc to present aluminum alloys do increase castability, machineability, and color in anodic oxidation. They do not harm physical or mechanical properties or corrosion resistance. They do not increase hotshortness.

Still the objection against zinc addition in sensible amounts to aluminum alloys continues. The critics have not produced facts to prove their case, nor have they been able to disprove the facts brought forth by the proponents. What then is the reason for this antagonism? Is it to be concluded that some kind of business politics is blocking the way of progress? I do not want to believe that such a thing is possible in an age where we are speaking of "pure" science.

Science has proven so far that zinc up to 3 per cent or so in aluminum alloys is either harmless or beneficial. Practice and use has borne this out.

^{*}W. Bonsack, "High Strength Non Heat-Treated Aluminum Casting Alloys," pp. 453-461 in this volume.

¹ Company Chief Metallurgist, Solar Aircraft Co., Des Moines, Iowa.

² Director of Research, Christiansen Corp., Chicago.

# EFFECT OF SLAG ON FURNACE LININGS

By

L. B. Wyckoff*

#### ABSTRACT

The following paper deals primarily with the chemical reactions of the slag on furnace linings. The course of the slag forming materials is followed from the charge to the slag tap hole showing calcining, preheating, random distribution and finally imperfect blending that may leave active slags in contact with the lining in spite of an overall balanced slag-lining combination.

Constitutional diagrams of some of the more common slagforming combinations are shown as a visual aid and a reasonably simple lesson on the use of such diagrams is offered. Four binary systems, two ternary systems and one diagram showing how to develop a cross-section of a ternary system are shown.

The saturation of the slag with respect to the lining material and the maximum furnace temperature is given as the basis of preventing lining erosion.

The erosion of furnace linings is the result of friction, heat and chemical activity. The effect of a solid charge sliding past a lining creates friction which is easy to visualize and needs no further discussion. The effect of heat is a little more complicated. Heat softens certain combinations within the refractory material at temperatures often much lower than the final melting point of the refractory. This softening may loosen the bond between the higher melting particles and allow friction to scour them off. Heat may also melt enough of the refractory to make it actually flow, or sag and drop like a silica brick in the roof of an open hearth furnace when it becomes too hot.

The chemical activity, however, is probably the greatest of all because it contributes to the other two actions. In considering slags and refractories we will be dealing mostly with metallic oxides, flourides or sulphides. In general, silica, SiO₂, is the principal acid oxide, lime (CaO), magnesia (MgO), manganese oxide (MnO), iron oxide (FeO), are the principal basic oxides. Alumina (Al₂O₃) may act as an acid, a base or a neutral, depending on the balance or lack of balance of other acids and bases. Calcium fluoride (CaF₂), fluorspar, is very active in combining or reacting with both acids and bases. It acts rapidly to make slags more fluid and requires only about 5

per cent. Unfortunately one of its compounds is volatile at furnace temperatures and rapidly escapes from some slags, so its effect is only temporary.

Slags and refractories are usually composed of some combination of the above oxides with some minor constituents which are carried in or on the pig, scrap, coke and fluxes. Since slag and refractories are often only different proportions of the same ingredients, it is possible for the two to react to give some different combination which would produce undesirable results. For instance, the ordinary fire brick is an aluminum silicate. The higher the Al₂O₃ the higher the melting point.

#### Cupola Slags

Cupola slags are aluminum silicates with some bases such as lime, magnesia, or iron oxides in addition. The limestone is calcined in the upper part of the cupola, driving off carbon and oxygen (CO2) and converting it to lime in lump form. The ash of the coke is partly in the form of chips of slate or shale and partly dispersed finer particles deeply imbedded in the coke. These materials individually may have higher melting points than the final slags and do not readily blend until much of the coke has been burned away. Thus they reach the reaction zone not far above the tuyeres and only then do they begin to blend to form the slag which is tapped out. Since the various components of the final slag reach the hot zone uncombined, it is inescapable that some of these ingredients will contact the lining and react with it before they are neutralized by other components of the charge.

The stability of lining from a chemical standpoint depends on the slag being combined in such proportion that reaction with the lining material would require higher temperatures than is obtained in the furnace.

In practice slags are seldom entirely uniform. They form in a bed of coke which has not given up all of its ash yet. Free blending is restricted by the physical difficulty of excesses of one element moving to a point where excesses of the balancing element exist. Slags which will no longer dissolve lime at the furnace temperatures may still carry it along as solids

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^{*} Electro Metallurgical Co., Retired, Lewiston, N.Y.

when it taps from the furnace, while some other portion of the slag may be lacking in lime. Slags may carry a considerable proportion of solids without greatly reducing their fluidity. A small excess of lime above the saturation point is permissible and should be a good deterrent against corrosion of basic linings. Lime being a stronger base than MgO would tie up the acid oxides and thus protect magnesia linings.

Magnesia (MgO) has considerable fluxing action on acid slags but relatively small fluxing action on basic slags. Additions of MgO in small amounts, possibly 5 to 10 per cent, could replace CaO for the sake of producing a saturation with respect to the MgO lining. From a straight fluxing standpoint Al₂O₃, CaF₂, FeO o_i MnO would be more effective. Korber and Oelsen* have discussed fluxing values of the various oxides along these lines.

#### **Acid Linings**

In the case of acid linings with basic slags, the acids from the lining would tend to lower the melting point of the basic slag and the bases from the slag would blend with the brick to give the brick a lower melting point. This, of course, would lead to heavy scouring of the brick.

Acid linings with acid slags may be considered as the normal conditions which are well known to all operators. In this case slags with less than 15 per cent Al₂O₃ would be fluxed to a slightly lower melting point by absorbing some of the brick. However, acid slags have lower heat conductivity and reactions are more sluggish than basic slags, so it is likely that the random distribution of the lime bringing some particles in contact with the brick is the major chemical factor leading to corrosion of acid linings with acid slags.

Scientists have studied the interaction of many of

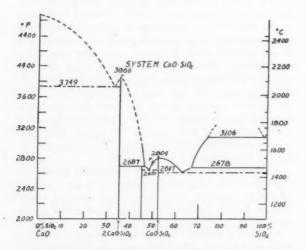


Fig. 1—CaO-SiO₂ System. (Adapted from G. A. Rankin and F. E. Wright, American Journal of Science, 4th Series, vol 39, p. 5, 1915, modified by J. W. Grieg, ibid. 5th series, vol. 13, pp. 1-44, 1927).

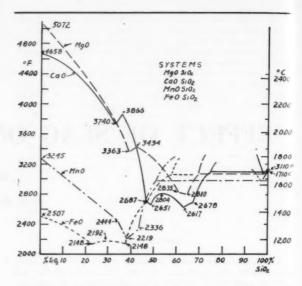


Fig. 2—Composite diagram of four systems adapted from the following works: MgO-SiO₂, N. L. Bowen and O. Anderson, American Journal of Science, 4th Series, vol. 37, p. 488, 1914; CaO-SiO₂, G. A. Rankin and F. E. Wright, American Journal of Science, 4th Series, vol. 39, p. 5, 1950; MnO-SiO₂, J. White, D. Howat and R. Hay, Journal, Royal Technical College, Glasgow, vol. 3, p. 239, 1933-36; FeO-SiO₂, N. L. Bowen and J. F. Schairer, American Journal of Science, 5th Series, vol. 24, p. 200, 1932).

the slag-forming elements systematically in groups of two or three components. Such studies have been set off in diagrams which give a visual impression of the general trends and for the advanced student give considerable detail.

First consider the system lime silica (CaO-SiO₂), Fig. 1. This diagram indicates a series of melts composed of lime and silica in all proportions from 100 per cent lime to 100 per cent silica. The percentage of SiO2 is set off on the base line. The balance in each case is CaO. For a composition represented by any point on the base line the point on the top line in the graph above that point represents the liquidus temperature, which is the temperature required to make that particular composition entirely liquid. The first horizontal line below the liquidus represents the solidus, which is the temperature below which no liquid is found when the melts have had time to reach equilibrium. Between these two lines there will be a mixture of liquid and solids. For example, assume point P at 50 per cent  $SiO_2$ . The liquidus would be 2740 F while the solidus would be 2651 F. Likewise, for any other ratio of CaO to SiO₂ corresponding values may be determined from the curve. From this diagram we can see that only the central section of the diagram shows melting ranges within the scope of the cupola practice. It is also apparent that it is not safe to generalize in saying either of the components raises or lowers the melting point without tying it in with some specific range of the diagram.

Figure 2 shows a composite diagram comparing

F. Korber and W. Oelsen, Stahl and Eisen, vol. 60, pp. 924-9 (1940).

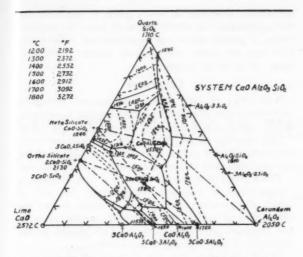


Fig. 3—CaO-Al₂O₃-SiO₂ System. (G. A. Rankin and F. E. Wright, American Journal of Science, 4th Series, vol. 39, p. 52, 1915; J. W. Greig, ibid. 5th Series, vol. 13, p. 41, 1923).

four systems which have significance in the studies of slag properties. These superimposed diagrams show that in binary composition with SiO₂, MgO gives no melting at all below 2810 F which temperature occurs at about 65 per cent SiO₂. CaO gives low melting points of 2617 F at about 63 per cent SiO₂ and 2651 F at about 49 per cent SiO₂. On the other hand MnO and FeO suppress the melting points down around the 2200 F range with SiO₂ as low as 38 per cent.

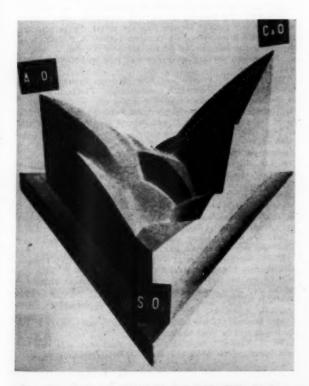


Fig. 3A—Photo of model of CaO-Al₂O₃-SiO₂ System.

These are all in respect to two component systems only and should be considered only as to the comparative fluxing values, not as final values. In slags there are practically no two component systems. Even in steel slags, which are often considered as lime silicates, there is always iron oxide present as well as phosphates and added fluxes which make the systems more complex.

Figure 3 shows the Ternary System (three components) CaO, Al₂O₃, SiO₂. This is given to show the effect of alumina on lime silica slags. Figure 3A is a photograph of a solid model of this system. In this triangular system each corner represents 100 per cent of the indicated component and the side opposite equals zero. Each side is laid off in ten equal parts by V marks (Fig. 3). Lines drawn through these marks parallel to each side will divide the diagram into 10 per cent lines, reading from the side line toward the opposite corner. The temperatures cannot be shown conveniently as vertical lines in this type of diagram so instead they are shown as dotted lines connecting all the points of an equal temperature. These lines are called isothermal lines or isotherms.

To see the effect of alumina (Al₂O₃) on the lime silica slags, take the point CaO-SiO₂ at 52 per cent SiO₂ on the ternary diagram CaO₂-Al₂O₃-SiO₂ as a starting point. Draw a line from this point to the Al₂O₃ corner. By laying the edge of a paper along this line the location of each isotherm and each boundary line can be marked on the paper. This then becomes the base for a sectional plane through the diagram. From each point so marked draw a perpendicular line and mark a distance to scale corresponding to the temperature indicated at this point by the isotherms. Connect these points to produce a series of curves intersecting at the boundary lines.

Figure 4 shows two sections thus constructed, one for the line CaO-SiO₂+Al₂O₃, and the other for a shorter section on the line CaO-SiO₂+Al₂O₃-SiO₂. The base line is calibrated in units equal to 10 per cent Al₂O₃ which in the case of the section going to the Al₂O₃ corner gives 10 divisions or 100 per cent, while the shorter section only covers a little over 60 per cent Al₂O₃.

Any line across the ternary field could be treated in the same way to produce a visual image of the

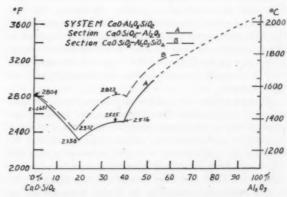


Fig. 4—Sections derived from Rankin and Wright ternary diagram CaO, Al₂O₃, SiO₂.

changes that would take place by the blending of two different combinations. One such combination might be the slag, the other the lining. By studying the curve along such a section it would be easy to visualize the changes in melting point that would occur to the lining if it united with some of the slag, or the changes in the melting point of the slag by uniting with some of the lining. Another combination might be the slag forming materials in the coke or other raw materials, the other the flux, such as lime.

. With a little practice these trends can be estimated by the pattern of the isothermal lines alone without actually drawing a cross-section. Isothermal lines close together indicate a steep slope to the curve, while lines far apart indicate a more gentle slope.

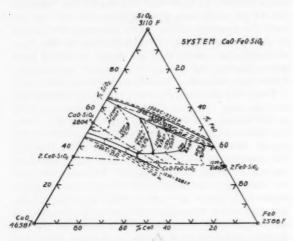


Fig. 5 — CaO-FeO-SiO₂ System. (N. L. Bowen, J. F. Schairer and E. Posnjak, American Journal of Science, 5th Series, vol 26, p. 204, 1933).

Figure 5 shows the system CaO, FeO, SiO₂. This shows that FeO like Al₂O₃ causes a decided lowering of the melting point as FeO is added to the CaO-SiO₂ compound. Looking at the two diagrams further though it would indicate that the iron had less effect than the Al₂O₃ in the more basic slags.

When a fourth or fifth component is added to the slag there are no adequate diagrams to indicate the behavior as a complete system. However, in the absence of other information it can be assumed that the first additions of a new element has a depressing effect on the melting point. With elements which only occur in small quantities it can be assumed that each additional element lowers the melting point to some extent.

Melting point is a rather indefinite term. It has different implications for different uses. In slags the implication is the temperature at which free flowing properties are produced, which approaches the liquidus. In a refractory it is the point at which the refractory loses its strength and sags out of place or collapses. That is somewhere above the solidus. At a slag refractory interface it is the temperature at which the lining will be dissolved in the slag. In using scientific diagrams we should select the liquidus

or solidus according to which point is significant in the application to which the data is to be applied. Only at points where the liquidus and solidus are the same are there true melting points. Otherwise there are melting ranges.

#### Conclusions

The effect of slag erosion on furnace linings is principally due to chemical action leading to solution of the refractory elements in the slag.

Slag forming elements combine to form low melting compositions called eutectics and high melting compositions and various intermediate combinations.

For a given temperature there is a line of limiting compositions in a ternary diagram on one side of which all particles may be put in solution and on the other side of which the excess of one or more of the elements will remain solid. This would be the isotherm of the liquidus temperature.

When the lining material, if dissolved, would raise the melting point of the slag above the maximum furnace temperature, the corrosion of the lining would be reduced to a minimum.

Slags are never completely uniform and some portions of slag may be of such a composition as to dissolve part of the lining even though the average or the bulk of the slag would be neutral.

A slight excess in the slag of elements akin to the lining material will help to eliminate the random scouring compositions.

#### DISCUSSION

Chairman: R. H. STONE, Vesuvius Crucible Co., Swissvale, Pittsburgh.

Co-Chairman: W. R. JAESCHKE, The Whiting Corp., Chicago.

Recorders: R. H. STONE and W. R. JAESCHKE.

MEMBER: The problem of the metallurgist is

MEMBER: The problem of the metallurgist is more complicated. Factors present are the slag, the metal, the air and the lining; with from five or six elements present it is difficult to determine the eutectic. In electric furnaces the slag is reduced or decomposed and metalloids present pass to both the metal and the slag.

MR. WYCKOFF: Fourth and fifth elements present tend to lower the liquidus point. It is more complex. The melting point of the last element may be lower but slag may be more viscous. Anything left in the slag will lower the melting point. Will Al₂O₂ lower the melting point? It will depend on what point of the

tri-axial diagram is being considered.

MEMBER: An acid lining in an electric furnace will pick up silicon from the metal. If basic elements are added in the stack the slag will become too fluid and corrode the lining. Adding silica will raise the melting point but slag will become viscous. Viscous or "long" slags will lower the electrical and thermal conductivity. It is difficult to select a lining which does not affect the alloy. With 7 per cent graphite in the lining there is no change in the metal (presumably carbon pick-up). You will get corrosion of the lining.

Mr. Wyckoff: There never is equilibrium in the furnace until the metal is tapped out.

H. W. LOWNIE: ¹ Typical acid slag will melt at 2200 F; basic slag at 2500 F. Tapping from the cupola at 2700 F, basic slag will be less fluid. Is acid slag necessarily more fluid at 3000 F in the stack than when tapped at 2700 F?

Mr. Wyckoff: Basic slag is in equilibrium with metal. Basic slag will appear less fluid.

Mr. Lownie: When basic or acid slag is up to 3000 F, the basic slag will be fluid like water.

W. M. Kerlin: ² Basic slag is more fluid than acid slag but will freeze quickly while acid slag will retain liquidity longer.

Battelle Memorial Institute, Columbus, Ohio.
 Meehanite Metal Corp., Cleveland.

# NEW METHODS IN METAL PATTERNMAKING

By

H. J. Jacobson*

We have just passed the halfway mark in the present century and it would be safe to say that man has made more progress in the last 50 years than in all the years since the patternmaker was born, some 350 years ago.

When we study the past history of man, we find that there are three basic developments in our entire history that have brought us where we are today.

The first was the discovery and development of sand molding with the use of a split pattern, along with the development of the common use of iron, and for 300 years man literally dug our present day civilization out of the earth with his bare hands and the muscles of his body.

The start of the second basic development took place in the 19th century, approximately 200 years later, and this was the discovery and development of artificial power. Here man began to use his brain instead of his muscles, harnessing the fruit of 300 years of hard labor.

We are all somewhat familiar with the third basic development which has taken place in the last 50 years; we have become the greatest nation on earth because of the development of machines and tools with which to use this power.

This machine and tool development has had its effect on patternmaking methods. We have changed from a wood working industry to a metal working industry. We have changed from wood shrink rules to steel shrink rules. We have changed from steel shrink rules to shrinkage charts which measure shrinkage in thousandths of an inch, so that micrometers can be used. We have changed from a pattern shop to a die shop in the true sense of the word.

Why has this change been necessary? There are two basic reasons: First, because of the machine and tool development in the various sequences of operation in the foundry to produce the casting.

A pattern equipment has two types of dimensions, "The Product Dimensions" and the "Foundry Functional Dimensions" and the more advanced the foundry becomes in machine and tool development, the

more important these functional dimensions become.

Today it has become general practice in the Pattern Shop to check these functional dimensions with "Checking or Proving Plaster" to see that all of the functional dimensions are correct within decimal tol-

erances before sending the equipment to the foundry.

The second reason these changes in Pattern Shop methods have been necessary is because of the machine and tool development on succeeding sequences of operation that casting must pass through to become a finished part or product.

On automotive and aircraft work it has become the practice to make casting drawings showing the amount of finish allowances for the pattern maker to provide and also predetermine locating pads on the casting which are first spot-faced in a qualifying fixture so that castings will machine in true relationship to the as-cast dimensions. These fixtures are called qualifying fixtures because of the significance of the operation being performed, such as "The casting is being qualified for the next sequence of the operation."

On small castings, however, where large quantities are made, this method is not practical, and is not necessary if proper patternshop and foundry procedure is followed. Here the casting must be a qualified casting in its as-cast condition and no qualifying operation is used.

#### Automatic Chucking of Castings

Automatic chucking of castings started in the valve and fitting industry 40 years ago and has now advanced to completely automatic loading exactly like feeding rolled strip steel to a punch press or feeding bar stock to a screw machine. A hardened steel qualifying gage at the top of the loading chute prevents shifted castings from entering the machine.

When man discovered how to make iron castings which were economically within his means, his first requirement was a stove with which to heat his home. His next requirement, developed in the 1800's, was plumbing and thus was born the valve and fitting industry. When an industry as basic and important to our national economy as the valve and fitting industry, universally accepts a means by which to gain a given result, it would be well for us to consider the

^{*} Industrial Pattern Works, Chicago.

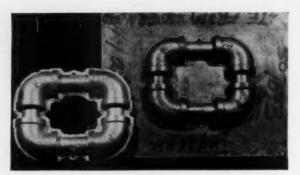


Fig. 1

reason for this practice.

Metal patterns or impression dies, as they should be called, were made in the valve and fitting pattern shops 40 years ago to extremely close tolerances, but machining a large number of split patterns all over before mounting on plates is costly and the result depended entirely on the skill of the mechanic or patternmaker doing the job. With a shortage of skilled patternmakers at that time, this problem led to a new development which was universally accepted by this industry and is still used today, and that is the electroplate copper method of producing exact reproductions of the master supplied, whether made of wood or metal.

From the dimensional standpoint these reproductions were almost to zero tolerance of the master, not only in the first and second dimension, but in the third dimension as well, exactly what the pattern-maker was looking for. The only fault that could be found with these pattern reproductions was in the material itself. Copper being a comparatively soft material, nicked easily, but for want of something better, this method of producing patterns was universally accepted in the valve and fitting industry.

This would lead us to believe that some basic value other than wear resistance was the answer and when we realize that this development took place at approximately the same time as automatic chucking, it ties in perfectly with the dimensional accuracy requirements.

#### Aluminum Metal Patterns

In the last 20 years aluminum has become a widely accepted metal for metal patterns and in the early 1940's one of the leaders in the valve and fitting field experimented with pressure cast aluminum pattern castings to replace the electroplate method. The basic difference here was that the electroplate method was a cold working method where no shrinkage was involved which meant that new master patterns had to be made for any replacements on equipment previously made.

After working with one of the leaders in the pressure cast field for some time, satisfactory pattern castings were made and this procedure has been followed since that time.

Figure 1 shows the master pattern and one pattern casting of the present method used. The master is made of brass machined all over to tolerances of

 $\pm 0.001$  in. The only machining allowance for the pattern castings is placed on the parting. This finish allowance is added by using  $\frac{1}{32}$ -in, sheet brass which is cut to a size which is approximately  $\frac{1}{8}$  in. larger than the contour shape of the pattern at the parting line.

The brass master pattern is then mounted on an aluminum plate held down by screws from the back side. The  $\frac{1}{32}$ -in. brass finish stock is placed between the master pattern and plate and held down by the same screws that hold the pattern. The reason for allowing this finish stock to extend over the edge of the pattern around the entire perimeter is so that when the pattern castings are machined on the parting face a sharp corner will result. This machining is usually done in a lathe by placing the casting against a jig block of 3-point or 4-point contact on core prints, depending on the shape of the pattern casting, and holding them in place by the tail stock center of the lathe or hold down screws.

The half pattern castings are made by the pressure cast method the same as pressure cast matchplates without the plate being cast integrally. The plaster sections are slushed together into one mold and poured at one time and gating is done on the back side of the casting through a round gate opening. In this way the pattern area is not distorted by gate grinding.

No internal chills are used thus eliminating the porosity prevalent in pressure cast plates.

An important advantage of this method is that accurate duplication can be made quite inexpensively if patterns should wear or if additional pattern equipment should be needed. This means then that you have a pattern set-up of somewhat of an expendable nature wherein it would be cheaper to replace these patterns than to do extensive repair work if required.

In some applications the patterns are jig drilled for complete interchangeability and replacements are made in the foundry proper.

#### 1-in. Valve Body Pattern

Figure 2 shows the results of metal pattern equipment made for a 1-in. valve body using this same method of producing the pattern castings. This valve body is machined in an automatic machine with form fitting jaws which hold only on the body contour of the casting. The complete machining cycle is completed without the casting being removed from the jaws.

The finish allowance on the inside of the casting is held to 0.35 in, on a side so that a minimum amount of stock must be removed in the machining operation. The general wall section on this casting was specified at 0.135 in, and in checking a representative number



Fig. 2

of castings the high was 0.140 in. and the low 0.130 in. or a  $\pm 0.005$  in. on wall thickness.

With this kind of close tolerances being held in sand casting dimensions the question will immediately be raised as to what kind of core box and drier equipment was used.

Figure 3 shows the core box equipment which is a 4-cavity blower core box run with one upper half box and two lower half boxes, the third half is not shown in Fig. 3 but it is an extra pin half which would be the same as the half shown on the lower part of this illustration. These boxes are also of an expendable type wherein it is cheaper to make new boxes than to make extensive repairs to the old or wornout boxes.

This is accomplished by a method similar to the die sinking method wherein a steel or permanent hob is made of the complete cavity cluster. This hob is referred to as a mandrel which becomes a permanent dimensional master for making any number of core boxes.

The castings made from this mandrel are accurate within  $\pm 0.002$  in, within a distance of 12 in,

The cavity surfaces are of mirror-like finish without hand or machine work of any kind and exact match on cavities can be held within 0.001 in. on a box of this kind with commonplace consistency.

Whenever possible cavity locations are laid out in such a way that the upper half box and the lower half box can be made from the same mandrel. (Note in Fig. 3 the two center cavities are mating halves and two outside are mating halves).

The material is an aluminum alloy with a very high abrasive resistance quality and these boxes are run in the blower without armor or facing of any kind. This makes it possible to keep the replacement cost at a comparatively low figure which is much less than extensive repair work on worn boxes.

#### **Driers**

The importance of driers in a production equipment of this kind must be given as much consideration as the rest of the pattern equipment.

Figure 4 shows a plaster core which was not removed from the lower half of the box after being made. These plasters were made with the box closed and plaster was poured and vibrated into the cavi-

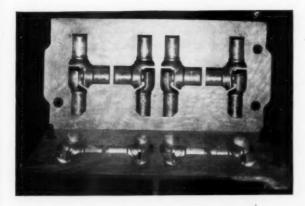


Fig. 3

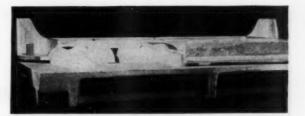


Fig. 4

ties through the blower holes; thus the plaster is an exact duplicate of the cavity from which it was made.

The drier which is in place on top of the lower. half of the box has been cut away to show the exact fit to the cavity contour. These driers are also cast by the permanent mandrel or impression die method. You will note that the conventional clearance used on driers of this kind is lacking in Fig. 4. The mandrel for the drier is made to 0.010 in. on the diameter larger than the core box and the height is made to give 0.002-in. to 0.004-in. crush on the core when the drier is in place over the core. These driers could be cast to the exact size of the core but it has been found necessary to give a clearance of 0.005 in. on the side so that the core can be removed readily from the drier.

It has been found possible to eliminate core wires in cores of this kind because of the extreme accuracy of the driers made by this method. The bridge wall through the valve body and port opening are not supported by the drier. This is shown by the slight opening at the top of the cavity on the left in Fig. 4. This is possible only because the balance of core is well supported by the drier contour.

The lower surface of the drier or the surface nearest the parting when the drier is in place is made  $\frac{1}{8}$  in. to  $\frac{5}{32}$  in. above the parting so that the bearing surface of the drier does not go all the way to the parting of the box. This is done so the drier will not shave when lowered into position. Pads shown in center hold drier to proper level.

Straight or near straight close-over surfaces are given  $\frac{1}{32}$ -in, clearance. This also can be seen just to the right of the bridge wall opening on the last cavity to the left.

These driers are normalized at 700 F for 2 hr which eliminates 90 per cent of the growth and removes all strains.

On present day core blowers designed to handle boxes of this type production speed of 500 blows per hour have become possible. This means that the operator is placing a drier on the box half every 8 seconds.

When rolling the box over at this speed if the core is not supported properly by the drier a great amount of core scrap would result.

#### Pin Corebox and Driers

Figure 5 shows this same practice used on a simple pin core of approximately 3/4-in. diameter. Here again only enough clearance is given on the side support of the cavity to allow the core to be removed

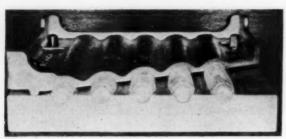


Fig. 5

readily from the drier without sticking.

Figure 5 also shows how the supporting surface of

the drier stops approximately 1/8 in. away from the parting of the box and the drier is supported by padof the same height which rest on the parting.

#### Summary

- 1. A casting can be made no more accurate than the pattern or impression die from which it is made.
- 2. Pattern accuracy must progress in direct proportion to machine and tool development.
- 3. Qualified castings can only be made with qualified pattern equipment.
- 4. To avoid confusion in pattern buying, this type of equipment should be called "impression dies."

# INFLUENCE OF SPECIMEN TUBE LOADING AND RIDDLE SIZE ON MOLDING SAND TEST RESULTS

By

D. I. Huizenga* and K. E. Spray**

In an effort to eliminate discrepancies in testing procedures between foundry laboratories throughout the country, the authors investigated various methods of test tube loading and riddling sands through various sized riddles to determine their influence upon molding sand test results. Thirteen foundry sands were used in the collection of data and every effort was made to achieve consistency in the sand testing procedure.

# Preparation of Sand Samples

The sand mixtures, six natural and seven synthetic, were taken from previous experimentation by Sanders and Sigerfoos.† Each mixture was carefully duplicated in composition, clay, and moisture content. A 36-in. muller-type mixer was used throughout the experiment. The muller was carefully scraped and cleaned of old sand before a new mixture was mulled. In each case the new mixture was mulled dry for 3 min or 135 revolutions, moisture added, and the mixture mulled an additional 5 min or 225 revolutions. The batch of muller sand was emptied into a large container and then screened directly into three small containers. Each container was sealed and labeled according to the mesh size of the riddle used. The containers remained sealed until actual testing time. All tests were conducted in the Michigan State College sand laboratory and standard A.F.S. sand testing procedures were used.

#### Moisture Determination

Since error in moisture computation could not be tolerated, care was exercised in making the moisture tests. The containers were systematically sampled and 50-gram weights of sand were oven dried for 1 hr at 225 F.

The percentage of moisture in the sand as listed in Table 1 is an average of three moisture checks of each mixture.

In compiling test data, alternate test specimens were followed completely through the testing procedure by the individuals involved. This provided a check on laboratory technique and gave each individual a thorough understanding of the problems involved in making the series of tests.

## Specimen Tube Loading

In an effort to determine the effect of the method used in loading the specimen tube various techniques were employed. Sand was gently introduced, thrown in, and put in with the aid of a copper funnel. These various methods produced no apparent discrepancies and it was assumed the actual ramming evened out any differences which might have occurred during the varoius types of loading.

## Errors Due to Test Equipment

Considerable difficulty and delay was experienced in getting consistent results while using the equipment in the laboratory. As a result of careful exam-

TABLE 1-MIXTURES TESTED

1.	(DY-1)	2.	50 lb Non-Ferrous Natura Bond
	5.8% Water		7.3% Water
3.	50 lb Gallia Red No. 3	4.	
	7.8% Water		2.6% Water
5.	50 lb Albany No. 2	6.	
	4.0% Water		6.9% Water
7.	50 lb Albany No. 2	Phy.	
	8.2% Water		
	Synthe	etic Sa	ands *
8.	48 lb Michigan City	9.	221/2 lb Michigan City
	2 lb Western Bentonite		221/2 lb Juanita Bank
	4.5% Water		5 lb Fireclay
			2.2% Water
10.	221/2 lb Michigan City	11.	221/2 lb Michigan City
	221/2 lb Juanita Bank		221/2 lb Juanita Bank
	5 lb Fireclay		5 lb Fireclay
	2.7% Water		3.7% Water
12.	24 lb Michigan City	13.	25 lb Michigan City
	24 lb Juanita Bank		25 lb Juanita Bank
	2 lb Southern Bentonite		2 lb Southern Bentonite
	2.0% Water		2.4% Water
		s give	en is actual percentage used.

^{*} Albion Malleable Iron Co., Albion, Mich.

^{**} U. S. Navy Dept., Bureau of Ships, Washington, D. C. The opinions expressed in the paper are those of the authors and do not necessarily reflect the opinions of the Navy Dept.

⁺ C. A. Sanders and C. C. Sigerfoos, "Gray Iron Shrinkage Related to Molding Sand Conditions," American Foundryman, Feb. 1951, pp. 49-55.

ination the specimen tube in use was found to be defective and was replaced by a new tube. After further examination, the stripping pedestal was found to be slightly worn and immediate corrections were made. This eliminated the main source of variables in the testing equipment. Thereafter, test data were more consistent. Considerable data were discarded at this point to eliminate any deviations due to new equipment. With the exception of the moisture determination, each sand property recorded in Table 2 represents an average of ten tests.

After all testing was completed the results were tabulated and the variations were noted in the averages for the three screen sizes (Table 2). By visual examination, differences were noted in the averages of the tests run for each of the riddle sizes. Further proof of these differences was obtained by statistical examination and in most cases significant differences were found. In this connection the Michigan State College Mathematics Department* furnished the

TABLE 2-AVERAGE PROPERTIES OF SANDS TESTED

	Size, penings	Weight, grams/ stand. specimen	Permea- bility	Mold Hard- ness	Green Comp. Str., psi	Mois ture,
•	2 4	160 158	20.3 21.5	88.3 87.2	26.6 29.1	5.8
	8	156	23.7	87.0	26.0	
	2	175	8.5	79.8	9.63	
2	4	170.7	9.5	80.6	9.84	7.3
	8 .	167	9.5	82.1	10.26	
	2	172	42.9	89.5	29.3	
3	4	170	44.6	88.9	35.3	7.8
	8	167	45.1	88.2	33.95	
	2	170	1.05	89.2	14.9	
4	4	169	1.	89.2	14.7	2.6
	8	168	_	88.8	14.2	
	2	170	1.7	86.1	14.5	
5	4	169	1.45	85.7	14.3	4.0
	8	168	1.55	85.7	14.1	
	2	170.9	10.2	84.8	14.6	
6	4	171	10.5	84.7	14.2	6.9
	8	168	11.05	84.2	13.5	
	2	178.2	2.5	83.9	11.13	
7	4	178	2.6	81.9	10.95	8.2
	8	177	2.75	80.6	10.27	
	2	164.7	246.	74.6	4.70	
8	4	162.8	251.5	74.95	4.69	4.5
	8	163.2	251.5	74.45	4.29	
	2	176	60.4	89.7	13.57	
9	4	175.2	59.8	89.9	13.48	2.2
	3	175	57.9	90.1	13.73	
	2	173.1	78.	86.8	11.2	
10	4	174	81.7	86.4	11.74	2.7
	8	174	81.0	87.	11.87	
	2	174	88.2	80.4	7.43	
11	4	173.2	90.7	79.3	6.96	3.7
	8	173	91.7	78.8	6.89	0.1
	2	159.6	120.	84.	9.26	
12	4	158.1	120.	83.2	9.28	2.0
	8	158	120.	84.6	9.32	410
	2	160.4	110.3	84.4	10.16	
13	4	160.4	110.3	84.1	9.97	2.4
	8	160	110.5	83.5	9.64	A. I
	Mixture		- 10.0	00.0	5.01	

STATISTICAL FORMULA USED

	Screen	Openings/in.	
2		4	8
$X_i$		$X_{11}$	$X_{21}$
-		-	
_		-	_
-			
X ₁₀		X ₂₀	X ₃₀
-1	Σ X ² —	$\frac{(\Sigma X)^2}{30} = B$	
	$S_1^2 + S_2^2 + S_3^2 + S_3^$	$S_3^2 \qquad (\Sigma X)^2$	С
	10	30	

ANALYSIS OF VARIANCE OF PROPERTY MEASUREMENTS

Degrees of Freedom	Sum of Squares	Mean Square	F Ratio
29	В	C	
2	C		$\frac{C^1}{-} = 3.4$
27	В-С=Е	$\frac{E}{-}=E^{1}$	E1
	Freedom 29	Freedom Squares  29 B  2 C	Freedom Squares Square  29 B  2 C $\frac{C}{2} = C^{1}$ 27 B-C=E

 $\frac{C^t}{If}$  is equal to or greater than 3.4 a significant difference is proven. This factor proves that in the present investigation the differences were due to the riddling of the sands.

authors with statistical formulas and verified the results noted.

## Conclusions

 The method employed in loading the test tube had a negligible effect on test results.

2. It was noted that differences in test results of the sands were a direct result of the riddling of the sands through No. 2, No. 4, and No. 8 riddles. All properties of the sands showed some degree of variation. The weight of a standard sample tends to decrease and the permeability tends to increase as the screen openings per linear inch increase. No definite trends were observed for mold hardness and green compression strength. By statistical formulas it was proven that these differences were not due

[•] The authors wish to express sincere appreciation to Dr. W. D. Baten of the M.S.C. Mathematics Department and the Experimental Station for hit assistance and aid in statistical procedure.

to chance, but were a direct result of riddling the sands through screens prior to testing.

#### Recommendations

1. Sand testing equipment should be checked periodically for wear. Equipment should be repaired or discarded if wear is apparent.

2. For consistency and accuracy, identical procedures should be followed for each test. These precautions should start with the collection of the sand sample and follow through on all testing procedures.

3. For standardization of test results between various laboratories, it is recommended that the sample be mulled and riddled according to a standard operating procedure.

4. For shop tests, the laboratory sample should, in most cases, be tested by using the same handling techniques as those used in the shop procedure.

#### DISCUSSION

Chairman: H. J. WILLIAMS, New Jersey Silica Sand Co., Millville, N. J.

Co-Chairman: STANTON WALKER, National Industrial Sand

Assn., Washington, D. C.

Recorder: D. C. WILLIAMS, Ohio State University, Columbus. J. E. Huss: 1 Did you turn the specimen tube after filling and before ramming?

MR. Spray: No, we did not turn the tube.

E. ZUPPANN: When conducting some similar tests we found the mixture lost moisture content during riddling in the open. This was particularly evident when the bond and/or water content was high. We found the best results were obtained when screening was executed through a funnel into a closed system.

Mr. Spray: Our practice was to screen sand from one container to another without the use of a closed system. All our tests were performed the same way so that whatever change would take place would always be the same.

Metallurgical Engineer, Lauhoff Grain Co., Danville, Ill.
 Foundry Engineer, Oliver Corp., South Bend, Ind.

# QUALITY CONTROL PROGRAM FOR A STEEL JOBBING FOUNDRY

By

H. H. Fairfield*

• Statistical quality control is a technique of management which has been used successfully in many different types of industries. The Society of Quality Control Engineers has active chapters in most industrial centers. The purpose of this paper is to describe the application of quality control in a jobbing foundry. Our steel foundry produces castings from one pound to five tons in weight. Carbon, alloy, and manganese steels are poured. There are no long production runs; most casting orders range from one to six pieces.

#### **Metal Quality**

The first step in the program was to post all chemical tests, physical tests, melting times, and down times on charts (Fig. 4). This enabled the melters to study the results of their efforts. The second step was to initiate standard statistical quality control charts for physical test results and some chemical tests (Fig. 2 and 3). This enabled the metallurgist to detect "assignable causes" of variation.

(An explanation of this term seems necessary at this point. If a graph is drawn of a test value, carbon content of mild steel for example, there will be shown a

variation from heat to heat. A certain amount of variation is unpreventable; it is characteristic of the operation. The melting foreman is anxious to detect any change in procedure. He may make the mistake of looking for trouble when the carbon is within normal limits, or he may fail to notice when carbon has exceeded the normal variation. The function of the quality control chart is to determine mathematically the normal range of variation. When a test value exceeds the normal range, it is assumed that an assignable cause for variation exists. If a process is to be kept under control and steadily improved, it is necessary to investigate each time an assignable cause is detected. In the rest of this report the term "A.C." will be used to indicate this type of investigation.)

The third step was to post a scoreboard at the electric furnaces (Fig. 1). Each melter was rated by the percentage of perfect heats made. Naturally, when the melters were charged with an A.C., they wanted to know the reason why the heat was off standard. Therefore the metallurgical lab had to investigate each A.C. and make a report on it. The quality control program greatly increased the amount of metallurgical work. However, the principle adopted was that we could not mark up an off heat against a melter if we could not

Chief Metallurgist, William Kennedy & Sons Ltd., Owen Sound, Ontario, Canada.

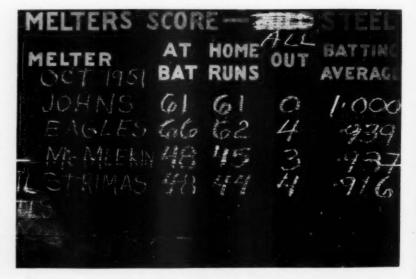


Fig. 1 . . . Scoreboard at the electric furnaces. Each melter is rated according to the percentage of perfect heats made. Naturally, he is anxious to maintain his record at as high a level as possible. Relative standings lend a competitive air to routine work.

explain how or why it was off. An A.C. summary is shown in Table 1.

#### **Assignable Causes**

I he most frequent A.C. was high or low carbon. The melters pointed out that the lab was too long in completing the carbon test, and that the metal changed while they were waiting for results. In order to overcome this difficulty, a 30-second test was developed which the melter could carry out. This enabled the melters to control carbons more accurately.

The second most frequent A.C. was high or low manganese. As the melters acquired better control over the deoxidizing phase of melting, manganese control improved. Since lack of manganese control resulted in erratic tensile strength, a rapid tensile strength test was developed. This test is carried out by the melter.

In 1948 and 1949, it was evident that the melters could not hold sulfurs within the range 0.10 to 0.15 per cent consistently. This steel was made in a basic furnace. One of the melters developed a method of handling the furnace slag in such a way that sulfurs were brought under control.

Many heats having lowered ductility were found to have Type II inclusions.² A decrease in the amount of aluminum from 3 lb to 2 lb per ton seemed to eliminate this trouble. Seven low-ductility heats tested by the Canadian Bureau of Mines showed Type II inclusions and 0.25 to 0.35 per cent copper. It was assumed that non-ferrous metals in the scrap were responsible.

Only a few cases of pinholes were satisfactorily solved. New ladles, green furnace spouts, and wet sand

Table 1 . . . Frequency of assignable causes are below.

Assignable Causes	June '48-May '49, per cent	June '49-Oct '51, per cent
High carbon	5.05	1.9
Low carbon	4.00	1.12
High manganese	5,30	0.85
Low manganese	3.50	0.76
Sulfur in free- cutting steel	4.00	0.32
Type 2 inclusions	1.70	0.06
Pinholes	1.65	0.17
Chromium	1.25	0.53
Unknown	1.1	0.085
Hydrogen flake	0.93	0.13
Sulfur	0.71	0.233
Dirty steel	0.60	0.42
Nickel	0.38	0.042
Phosphorus	0.33	0.042
Woody fracture	0.165	0.042

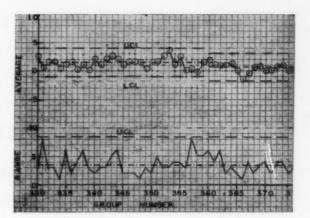


Fig. 2... Quality control chart for elongation of mild steel.

Deviations from Table 2 are plotted here.

were blamed. The incidence of pinholes has dropped from 1.65 to 0.17 per cent. This is probably due to the present practise of boiling every heat from 0.40-0.45 per cent carbon down to 0.08-0.13 per cent carbon.

#### **Metal Quality Improvement**

The frequency of assignable causes has dropped from 30 per cent to less than 7 per cent. It is hoped that further improvement can be made. The melting process is not under control in the statistical sense; however, it is tending in that direction. In order to clearly define the role played by quality control in this project it is necessary to analyze the reasons for the

Table 2 . . . Mild steel elongation and tensile strength.

Tensile strength, psi	Elongation, per cent
65,000	32.3
66,000	32.0
67,000	31.6
68,000	31.3
69,000	31.0
70,000	30.6
71,000	30.2
72,000	29.9
73,000	29.6
74,000	29.2
75,000	28.9
76,000	28.4
77,000	28.2
78,000	27.8
79,000	27.4
80,000	27.1

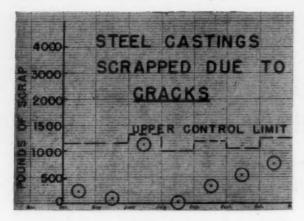


Fig. 3... Quality control chart for pounds of steel castings found defective due to cracks in them.

improvement in quality. These reasons are as follows:

 All test data were displayed so that melters and lab staff were fully informed on melting operations.

2. Each melter was rated monthly according to the per cent of perfect heats made. This rating was posted on the wall.

3. Melters did their best to get a good score. In order to improve their score, they called upon the metallurgical department for information.

4. Melting foremen obtained information pertaining to melting operations from a number of sources.

5. Under the incentive supplied by the quality control program, all of the people involved in the melting

Table 3 . . . Shows fraction defective steel castings.

Defect	Fraction Defective	Normal limit pound of defective casting
	201001110	per month
Planning errors	.00920	7700
Pinholes	. 00490	4640
Cracks	. 00260	1748
Dirt	.00200	2268
Misrun	.00170	1652
Runout	.00115	1490
Gas holes	.00089	1222
Molding	. 00076	1110
Shrinks	. 00062	843
Bad metal	. 00058	896
Scabs	. 00046	776
Short pour	.00040	700
Wrong metal	.00010	320
Burned cores	. 00008	.246

Note: Normal limits based on 300 tons per month.



Fig. 4... Charts of chemical and physical tests are shown here. Similar charts give melting and down times.

operation learned more about the work and did their jobs more efficiently.

Since the production runs here are short, it is not possible to set up charts for individual castings. A monthly summary of scrap castings is compiled. Data are recorded as in Table 3 and Fig. 2 and 3. When normal ranges are exceeded, corrections are applied.

Quality control methods have not yet been applied to sand control. Some 12 different sand mixtures are prepared daily. This does not mean that 12 separate charts would have to be plotted for quality control purposes. Deviation charts could be used. On the moisture chart, for example, the moisture tests for all mixes would be recorded in terms of deviation from standard. This would show the degree of control exercised by the mixer.

As the castings are poured, the weights are recorded. Since the selling value of each casting is known, it is possible to determine the hours of labor required per \$1,000 worth of castings. Charts are maintained and a bulletin posted, showing the man-hours worked per unit of product in each foundry department. The cleaning room is rated according to the value of product shipped. Each foreman therefore knows from week to week his production and efficiency score.

Quality control charts are an excellent medium by means of which workers and management can be made acquainted with factual information concerning manufacturing operations. When a worker finds his own personal performance charted on a graph or listed in a bulletin, he is nearly always motivated to improve his work. Most men are proud of doing a good job, and the control chart enables a man to check his work.

Posting up quality control charts throws extra work on the management. Foremen are called upon to correct conditions which prevent the worker from attaining optimum efficiency. Foremen in turn call upon higher management to improve working facilities.

#### Bibliography

1. H. H. Fairfield, H. F. Graham, and A. E. McMeekin, "Metal Composition Tests for the Steel Melter," TRANSACTIONS, American Foundrymen's Society, vol. 58, pp. 530-536 (1950).

2. C. W. Briggs, Metallurgy of Steel Castings, (1946), p. 149-150. McGraw-Hill Book Co., New York.

# CUTTING WITH ULTRA-HIGH PURITY OXYGEN

By

E. H. Roper* and J. F. Kiernan**

#### ABSTRACT

Experiments were made to determine effects of oxygen purities between 99.0 and 99.83 per cent on cutting and starting of cuts on steel plate, billets, and cast steel sections. Contrary to earlier investigations, some of which showed rapidly increasing efficiencies and others which showed a leveling off of the increase of efficiency as the purity increased, these studies show a straight line relation between oxygen purity and cutting efficiency for the region explored. In general, the rate of gain between 99.0 and 99.83 per cent is about 2 per cent increase in speed for each 0.1 per cent increase in purity. Starting times were independent of purity in the ranges tested. As between rolled and cast material there was no significant difference except as regards difficulty of measurement. Almost 1400 cuts and starts were made.

A commercial size oxygen plant was run to determine the trend of expense of producing higher than 99.5 per cent pure oxygen and a study was made of the reliability of oxygen

The conclusion was reached that the small gains, increased cost, and analytical difficulties attendant to the production and use of oxygen approaching 100 per cent purity prevent its being a feasible commercial product for oxygen cutting in the present state of development.

Although the study of the effects of oxygen purity on the oxygen cutting process and its economics has been going on for some time, the region of purity between 99.5 and 100 per cent has been largely one of speculation. The work reported on in this paper was undertaken to provide some data in that region concerning the possibilities and costs of providing and using oxygen with a guaranteed purity higher than the present standard 99.5 per cent pure product.

#### Historical Background

It is commonly accepted that the first man to use oxygen for the cutting of steel was Thomas Fletcher of Warrington, England.¹ At the time, 1888, there was much interest in the production and use of oxygen. Fletcher was an indefatigable researcher in this field, but nowhere in his papers or in similar papers of that period have the authors been able to find any allusion to the purity of the oxygen then available.

In 1890, Dr. L. T. Thorne in a paper² read before the Society of Chemical Industry in Liverpool, England, reviewed the methods and the already extensive patent literature on oxygen production. He mentioned over fifty patents covering four different basic methods of getting this useful gas. It is significant that only a miniscule percentage of the oxygen produced in the world today is by any of the methods mentioned by Dr. Thorne.

For his work in combustion, Fletcher used oxygen made by Brin's process and it is reasonable to assume that the same process was used to make the oxygen used in his oxygen lance. Briefly, the Brin process consisted of maintaining barium oxide in a retort at about 700 F and subjecting it to air at about 10 psi gauge. Under such circumstances a peroxide was formed, which decomposed to the oxide and oxygen when the pressure was reduced to 13 to 14 psi below the atmosphere. A system of pumps, cocks, and purge valves reversed the cycle every 15 min. Depending on nitrogen entrainment and leakage, oxygen of around 93 to 95 per cent purity could be obtained and this is probably what Fletcher used.

In 1911 Tucker delivered a paper showing the results³ of oxygen cutting experiments using oxygen of different purities. He was prompted to undertake this work because the oxygen then available varied in purity from 92.0 to 99.7 per cent and there was noticed some variation in cutting efficiency as the purity of the oxygen varied. So well did Tucker do his work that, in general, subsequent experiments have merely proved the soundness of his overall conclusions.

In the 41 years since Tucker delivered his paper many investigators have studied the effect of oxygen purity on various phases of the cutting process. Since their techniques varied, it is hard to make exact comparison of the results. However, if the results are reduced to as near a common basis as possible, you get a relationship similar to that in Fig. 1. This chart teaches that the greatest gains are to be made in the region from 98 per cent purity upwards. That area has been intensively explored by many, with some investigators showing that the low of diminish-

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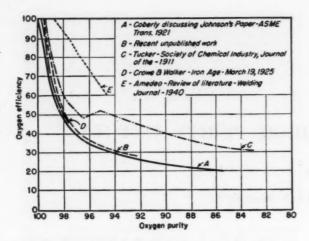


Fig. 1—Approximate comparison of investigations of the effects of oxygen purity on cutting. (From "Oxygen Cutting," Slottman and Roper).

ing returns sets in as we approach 100 per cent purity, while others felt that small increases in purity give increasing large gains. Up to now, there has been little careful investigation beyond 99.5 per cent purity.

#### Analysis of the Problem

At first glance it appears odd that, with oxygen generally available with purities guaranteed within close tolerances to be 99.5 per cent there has not been exploration in the region from 99.5 to 100 per cent. There are however, several things which are responsible, all of which result in certain current practices of oxygen production and use.

First, there is the lack of a ready means suitable for plant use to determine quickly and with the desired accuracy the purity of a given sample of very pure oxygen. The standard instrument used today is the Hempel absorption apparatus in which the oxygen in the sample is absorbed in a flask containing copper wire in an ammoniacal solution of cuprous chloride. That portion of the sample not absorbed is transferred to a burette, measured, and called the impurity. Among analysts those methods of gas analysis employing aqueous solutions are not rated in accuracy closer than  $\pm 0.05$  per cent. If the percentage of oxygen in our samples were small, this would be a satisfactory accuracy, but as we approach 100 per cent oxygen in the sample, it is inadequate. An accurate method of testing purity another way, namely, by absorbing the small amount of impurity would be ideal, but there does not seem to be any such method which is sufficiently fast and simple. Research into this problem is continuous, and we hope ultimately to have a better method.

Meanwhile, a producer of oxygen who wishes to sell a 99.5 per cent pure product gets along with the Hempel type equipment. Since the testing instrument may be in error by  $\pm 0.05$  per cent and since no oxygen producing equipment runs at a constant purity hour after hour, the average user will get industrial oxygen varying in purity from 99.5 to 99.8

per cent. The 0.5 per cent between 99.5 and 100 per cent is a range in which the producer of the gas can operate without delivering a product below specification. However, if an effort is made to supply 99.8 per cent pure gas this margin for operating variations is reduced to 0.2 per cent, only two-fifths of that available when working at the lower purity. In order for a user to build his cutting practices around oxygen 99.8 per cent pure, he would have to be assured of getting that grade day in and day out, otherwise his operators would reduce their cutting techniques to a point where they would not lose their cuts or have other troubles. Their practice would soon be based on the lowest purity they were likely to get. Should the purity of the oxygen coming into a shop drop and there be spoiled work due to lost or ragged cuts. the operators would quickly reduce their cutting speeds or increase their oxygen flows to insure against further difficulty.

A second factor that has held back exploitation of oxygen over 99.5 per cent pure is the difficulty of determining with significant accuracy the practical effects of purities higher than this. A host of variables haunt the experimenter. If he is to achieve reproducible results, he must hold all factors but one constant while that remaining one is varied according to a plan. Some of items to be controlled are:

The purity of the oxygen The uniformity of the fuel gas Temperature of the oxygen and fuel gas Flows of all gases (cutting oxygen, preheat oxygen and fuel gas) Diameter of cutting oxygen stream Cleanliness of all orifices in the tip Temperature of the workpiece Barometric pressure Atmospheric temperature Temperature of the workpiece Design of the cutting tip Height of tip from work Angularity of torch to the work Composition of the workpiece Air movement in the cutting area Speed of cutting Surface condition of workpiece (top and bottom)

Some of the items to be considered in evaluating the results are:

Quality of the cut surface Depth and regularity of drag lines Slope of drag lines Squareness of cut face Flatness of cut face Meltdown of top edge Slag adherance on bottom edge.

In view of the above, it is small wonder that the cutting procedures recommended by makers of oxyacetylene cutting equipment and machines carry a notation to the effect that the procedures are approximate only; or that the effects of small differences in purity have been so much debated in the literature.

Finally to be considered are the costs of producing

TABLE 1-TABULATION OF DUPLICATE ANALYSES OF OXYGEN, %

Cylinder No.	Orsat	Modified Hempel Pipette	Plant Hempel	Develop- ment Division Hempel
693794	99.72-99.71	99.75-99.75-99.75	99.82-99.82	99.80-99.81
693811	99.80-99.82	99.77-99.78	99.80-99.82	99.79-99.78
693774	99.78-99.80	99.80-99.80	99.80-99.80	99.79-99.80
693776	99.81-99.83	99.80-99.79	99.82-99.82	99.81-99.80
693773	99.78-99.78	99.79-99.80	99.80-99.80	99.81-99.83
693777	99.80-99.85	99.78-99.80	99.82-99.82	99.82-99.80
693782	99.85-99.81	99.80-99.80	99.82-99.82	99.80-99.81
693793	99.83-99.84	99.79-99.78	99.82-99.82	99.80-99.82
693775	99.85-99.85	99.79-99.80	99.82-99.83	99.76-99.78
693827	99.69-99.65	99.79-99.78	99.82-99.82	99.79-99.80
693798	99.78-99.79	99.79-99.78	99.80-99.80	99.79-99.81
693816	99.80-99.83	99.79-99.80	99.81-99.82	99.81-99.83
693778	99.80-99.78	99.80-99.80	99.82-99.82	99.82-99.83
693771	99.81-99.79	99.75-99.78	99.82-99.82	99.75-99.79
693792	99.78-99.78	99.78-99.77	99.81-99.82	99.84-99.88
693788	99.79-99.79	99.74-99.73	99.82-99.82	99.78-99.82
693810	99.82-99.80	99.76-99.75	99.80-99.81	99.81-99.83
693808	99.85-99.84	99.75-99.75	99.81-99.82	99.83-99.84
693797	99.79-99.82	99.78-99.77	99.82-99.82	99.80-99.82
693791 -	99.78-99.79	99.76-99.76	99.80-99.82	99.81-99.82

a gas of a guaranteed purity higher than 99.5 per cent. Production rates would be somewhat lowered and gas analysis would be critical as well as debatable, so, since the anticipated gains to the user at that time seemed probably small, the oxygen industry settled on the 99.5 per cent pure gas as their standard. Higher purities are available in small quantities for research laboratory use and large quantities of lower purity (about 90 to 95 per cent so called, tonnage oxygen) are made, but in this paper we shall consider the 99.5 per cent grade normally used in the United States for welding and cutting purposes.

#### **Experimental Procedure**

Oxygen Testing and Supply—Basic to any cutting study of this sort are supplies of oxygen of reliable known purities. As can be realized from the discussion of purity measurements, this is not an easy task. Securing a supply of gas of definite purity was solved in one case by asking a plant to make up to 20 cylinders as near 99.8 per cent as possible. The gas in each cylinder was then analyzed in two different locations, by three separate groups of men using a total of four methods. The methods and range of results on each cylinder are listed in Table 1. In addition, one cylinder was tested by an outside laboratory. If we take the average of all the results on this cylinder as correct, we get a figure of 99.82 per cent and the deviation of each of the methods is as follows:

Plant Hempel	0.00
Research Division Hempel	-0.04
Development Division Hempel	$\pm 0.04$
Research Division Orsat	-0.04
Outside Laboratory	$\pm 0.01$

To obtain the lower purities of gases for the cutting tests small amounts of argon were added, and the cylinders rolled on the floor and tested until ad-

TABLE 2-OXYGEN ANALYSIS

Cylinder No.	% Oxygen
693827	99.01, 99.02
693771	99.15, 99.11
693808	99.30, 99.32
693788	99.34, 99.34
693798	99.42, 99.41
693816	99.53, 99.51
693777	99.53, 99.55
693794	99.73, 99.71
693776	99.73, —

ditional rolling showed no change in the results. The analysis of these cylinders are given in Table 2. Oxygen for the cutting of the keel blocks mentioned later in the paper was obtained by filling a trailer with the highest purity oxygen a new and modern column would produce and then testing it as it was discharged. At the same time, cylinders were filled and later diluted by the methods just described. While commercial oxygen is normally guaranteed to meet a specified purity within ±0.1 per cent, it is felt that the care used when analyzing the gas used to conduct the experiments described in this paper warrants considering the purities as being within ±0.05 per cent. If there are other errors, they are constant ones and their effects on the results would be uniform.

Cutting 3/4-in. Plate—Approximately 380 cuts 12 in. long were made in 3/4-in. plate to determine the relationships between orifice diameter, cutting speed, oxygen purity and cutting oxygen flows. In these tests, tips with cutting oxygen orifices of 0.0465 in. (Size No. 1), 0.055 in. (Size No. 2), and 0.0635 in. (Size No. 3) diameter were employed. For all tips and cutting oxygen flows, preheat acetylene was maintained at 14 cfh with neutral flames. The same

TABLE 3-ANALYSIS OF 3/4-IN. PLATE USED FOR TEST

Element	Fire Box	Quality Steel	Hot Rolled Steel
Carbon		0.186	0.22
Sulphur		0.026	0.052
Manganese		0.589	0.446
Phosphorus		0.014	0.043
Silicon		0.16	
Aluminum		0.02	_
Iron		Balance	Balance

purity of gas was used for both cutting and preheat with all flows measured by calibrated direct reading rotameters.

All cutting was done by mounting the torch in a machine to insure constant speed, straight line cutting, and a constant distance from tip to work (13/32 in. in all tests). Torch and workpiece were checked for squareness in two planes. The workpiece was supported on both sides of the cut so that the dropping of the cut piece would not effect the determination. Speed of cutting was calculated by measuring electrically (to within 1/100 ths of a second) the time for the machine to travel a mechanically fixed distance. All steel was sand blasted or wire brushed to remove surface effects and the "end of cut" edge carefully squared. Analyses are in the Appendix. Barometric pressures and gas flowing temperatures were measured and used to correct readings.

Actual test procedure was to set all conditions as desired except for speed and then to make cuts at varying speeds until a maximum was found at which complete severance of the final corner of the cut

could just be made.

4-in. Plate-Sixty-four cuts, 4 in. long, were made in 4-in. square steel bars of the analysis shown in Table 4 using largely the same procedure as for 3/4in. material. However, determinations were made using only 99.5 and 99.8 per cent pure oxygen, whereas 99.0, 99.32, 99.49, and 99.80 per cent were used on 3/4-in. material. Prior to cutting, all steel was washed with carbon tetrachloride and rubbed down with steel wool.

An 0.082-in. cutting oxygen orifice tip was employed during the study. The cutting oxygen flow was maintained constant at 360 cfh; the acetylene flow was held constant at 23 cfh with the flames adjusted to neutral. The cutting oxygen orifice was cleaned prior to each cut. The material on both sides of the cut was fully supported in order that the weight of the dropping piece would not influence the drop cut speed determination. The torch and workpiece was frequently checked by means of a square and a bubble level.

The general method for isolating the drop cut

TABLE 4-ANALYSIS OF 4-IN. SQUARE BAR USED FOR TEST

Element	Content, %	
Carbon	0.177	_
Manganese	0.89	
Phosphorus	0.016	
Sulphur	0.023	
Iron	Balance	

speed determination was to make cuts in a selected speed bracket, and by successive increments of speed increases and decreases narrow the speed range to where a drop cut could just be made. In order to minimize any effects upon the cutting speed that may have resulted from non-uniformity of the steel, the final determinations for both purities were made by alternate switching of the two purities and making several cuts with each purity. During the purity switching operations, the gas systems were subjected to a triple purging operation to insure against possible dilution.

Starts-Although it was felt desirable to do some hand cutting using oxygen of different purities, the practical impossibility of reducing the variables so as to get meaningful results was recognized. In any hand cutting of bars, rounds, risers, etc., the start is important. Hence, 716 tests were made using a mechanized arrangement as shown in Fig. 2 to determine

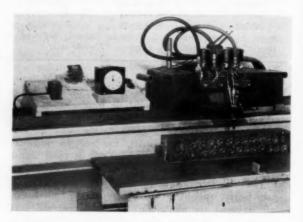


Fig. 2—Arrangement of apparatus for oxygen purity starting tests.

the effects of oxygen purity on the time needed to make a piercing start. In this work, oxygen of 99.3, 99.5, and 99.8 per cent purities was employed. The investigation was conducted on a number of 2-ft long bars of 4 in. square cold rolled steel (for steel analysis, see Table 4), the surfaces of which were cleaned by rubbing with steel wool and carbon tetrachloride. Starts were attempted at a minimum distance of 1 in. from all edges in order to insure that massive steel was being acted upon at all times. The steel specimens were mounted on a fixed cutting base, and the special torch and torch holder fixture were attended above it. The mounting base and torch holder fixture were permanently located with respect to each other, thus insuring a fixed relationship between the cutting tip and work specimen throughout the investigation.

An 0.082-in, orifice cutting tip was employed with constant operating conditions of a cutting oxygen flow of 360 cfh, an acetylene flow of 44 cfh (flames adjusted to neutral), and an operating height of 13/32 in. (tip face to work). The conditions were selected as representative of those that would normally be employed in the cutting of 4-in. cast steel. All gas

flows were measured by means of direct reading rota-

Control of all gases was exerted by use of electrically operated solenoid valves located in close proximity of the torch head. These valves were automatically controlled by an electrically operated starting time measurer which will measure intervals as small

The operating sequence was as follows: Closure of the main control switch energized the timers and the preheat gas solenoid valves; a pilot flame at the tip immediately ignited the preheat flames. At the end of the preset duration, the timers energized the cutting oxygen solenoid valve; the action of the timers at this phase having no effect on the solenoids controlling the preheat flames which continued burning while the cutting oxygen was flowing.

After the cutting oxygen had flowed for a sufficient time to ascertain the ability to develop a start, the main control switch was opened and all circuits were de-energized, thus resulting in the closure of all gas lines.

The action of the starting time device was such that preheating time interval and the application of the cutting oxygen were automatically controlled and that preheating time intervals were precisely recorded. It was observed that as the preheating gases were ignited, a short period of time elapsed during which the flame characteristics were slightly carburizing prior to changing to neutral; but inasmuch as the physical operating conditions of the system remained constant throughout the investigations, this short time interval also remained constant, thus having no effect upon the relativity between tests.

Cast Steel-Since it was realized that any significant results on cast steel would depend largely on the use of uniform materials, special efforts were made in that direction. We were fortunate in obtaining a number of keel blocks of Grade "B" steel (carbon 0.27, manganese 0.70, silicon 0.45, phosphorus 0.028, and sulphur 0.042). These were prepared in a leading foundry under the direction of personnel who

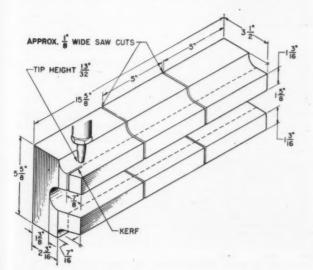


Fig. 3-Preparation of cast keel blocks for cutting.

appreciated the problem and were delivered in the as-cast condition after a thorough sand blasting. In order to end each cut on a surface that would be uniform, each block was prepared as in Fig. 3, permitting about a 5-in. cut ending on a machined surface and six cuts from each block.

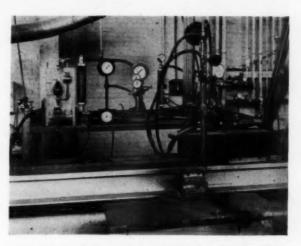


Fig. 4—Arrangement of apparatus employed during cutting studies.

Much the same set-up as used for the 3/4-in. plate was used for cutting the cast material. Figure 4 shows the general arrangement of apparatus. A heavy preheat tip using No. 2 (0.055-in. diam) cutting orifices was employed with the following conditions remaining constant throughout: 145 cfh cutting oxygen flow, 14 cfh acetylene flow (preheat flames neutral), and 13/32-in. operating height. Cognizance was taken of the other factors that might affect the results and the usual frequent checks made in alignment of the apparatus, support of the work, etc.

While it was hoped that the same criteria could be used as when cutting the wrought material, it soon became apparent that it would be impossible to "pin-point" with the same accuracy the maximum drop-cut speeds when using the different oxygen purities of 99.83 per cent, 99.58 per cent and 99.47 per cent which were available for the cast steel series of tests. When cutting, variations in results developed

due to any or several of the following:

1. Irregularities in the as-cast skin of the castings. 2. Presence of non-oxidizable inclusions, porosity,

and segregations.

3. Variations in the cross-sectional dimensions.

Under normal conditions of production, cutting these factors would not have been noticed, since their magnitudes were small, but for this precise test program they were very significant.

Hence, the testing procedure had to be modified so as to make the basis for comparison of various oxygen purities the maximum speed ranges at which a given consistency (65 per cent) of drop cuts could be obtained.

#### Cutting and Starting Tests on Rolled Steel

Since our recommended procedure for cutting 3/4-

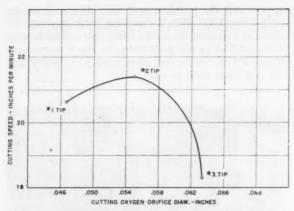
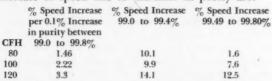


Fig. 5—Determination of best size of tip, using 99.8 pct purity oxygen. 99.80% Purity—100 cfh; Acet. = 14 cfh; Flame neutral; ¾-in. steel plate.

in. plate using 99.5 per cent oxygen calls for a No. 2 tip using 100 cfh, one set of determinations was used to establish the correct tip size for 99.8 per cent pure gas. Figure 5 shows the No. 2 size tip best.

In Figure 6 are shown the plotted results on 3/4-in. plate. Point A represents normal recommended operating procedure for this material. It is only necessary to accept a 21/2 per cent variation in the results to draw a straight line through the purity-speed reading for any gas tested. Non-homogeneity of the material, variations in the speed of the torch carriage and the expected errors in oxygen analysis can easily account for a 21/2 per cent variation in results. Increase of speed due to increases in purity are:



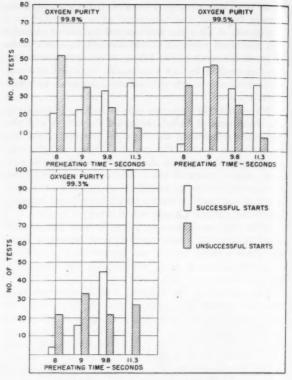


FIG. 7 EFFECT OF OXYGEN PURITY ON STARTING TIME

Fig. 7—Effect of oxygen purity on starting time.

For the cutting in 4-in. material results were as follows:

Purity 99.5 8.2 99.8 Cutting Speed In./Min % Increase in Speed  $6 \pm 2$ 

The  $\pm 2$  per cent tolerance is necessary to accommodate the spread in readings on the successful cuts

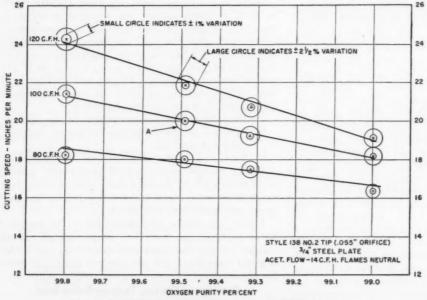


Fig. 6—Effect of oxygen purity on speed of cutting ¾-in. steel plate.

FIG. 6 EFFECT OF OXYGEN PURITY ON SPEED OF CUTTING 3/4" STEEL PLATE

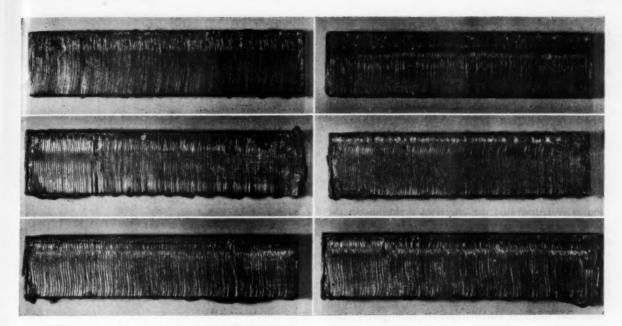


Fig. 8—Specimens of cast steel cut quality. Top, using 99.47 pct oxygen purity; Center, using 99.58 pct oxygen purity; Bottom, using 99.83 pct oxygen purity.

and we feel represents realistic accuracy.

The results of the starting tests are shown in Fig. 7. Numerically the results are:

Preheating Time, sec	Percentage	of Successful Gas Of	Starts Using
	99.3% purity	99.5% purity	99.8% purity
$8.0 \pm .2$	15	17	17
$9.0 \pm .2$	34	54	41
$9.8 \pm .2$	67	55	57
$11.3 \pm .2$	81	89	88

In other words, within the range of purities tested, preheating time is the dominant factor and variations in purity have no effect.

#### Results of Cutting Cast Steel

Of the 239 cuts made on the cast steel keel blocks, 31 were lost as the result of too high a speed or unexpected non-oxidizable inclusions. In addition, evidences of cut variation caused by irregularities of the

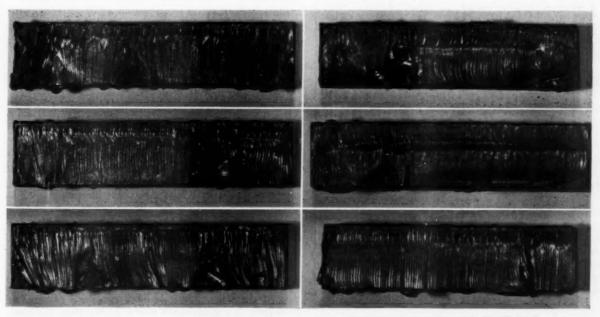


Fig. 9—Effect upon cut quality of inclusions in cast steel. Top, using 99.47 pct oxygen purity; Center, using 99.58 pct. oxygen purity; Bottom, using 99.83 pct oxygen purity.

as-cast skin were noted in 58 other cuts. Hence almost one-third of the cuts might be excluded from any discussion of the results. However, we have included them, since they are about evenly distributed among the three oxygen purities used. It was observed during this cutting program that the degree of preheating was very important. Those working on the project felt that higher preheats and reduced speed employing any purity of oxygen used in these tests would have eliminated all lost cuts.

Some typical cut surfaces when the steel was clean are shown in Fig. 8. All were cut at speeds within the maximum speed range for the purity involved and in general the quality is uniform. How the skin condition affects quality is readily noticeable. In Fig. 9 are illustrated the effects of inclusions in the steel and it will be noted that the gouges are attributable to conditions in the skin, either at the lower

or upper edges of the cut.

The results of all cuts on cast steel including those lost regardless of the reason are shown in Fig. 10. The points on the left represent the first tests made. As the experimenters became more familiar with how the material would react at each purity, they were able to narrow the speed ranges. Even so, the final selection plotted in Fig. 11 represents a consistency of 65 per cent drop cuts for each speed range. The manner of plotting the data is given in the Appendix.

For Fig. 11 there might be justification for drawing a curve between the three points, which would

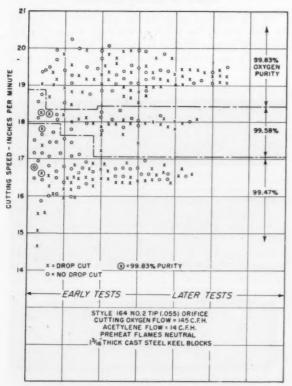


FIG. 10 RESULTS OF ALL CUTTING TESTS ON CAST STEEL

Fig. 10—Results of all cutting tests on cast steel.

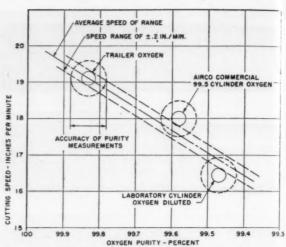


Fig. 11—Effect of oxygen purity on speed of cutting cast steel. Style 164 No. 2 tip (.055) orifice; cutting oxygen flow, 145 cfh; acetylene flow, 14 cfh; preheat flames neutral; 13/16-in. thick cast steel keel blocks.

tend to show a diminishing rate of improvement with higher purities of oxygen. However, in view of the straight line relation developed on wrought material and the large experimental variations that resulted from the work on cast steel we believe a similar straight line is more realistic. The average improvement in speed when using 99.80 per cent oxygen over 99.50 per cent is then 10.8 per cent with a probable variation of  $\pm 2.5$  per cent.

#### Effect of Purity on Cost of Oxygen

As is the case in the production of many other industrial materials, higher oxygen purity means greater cost and decreased output of the producing machinery. Each increase in purity becomes increasingly hard to get as we approach a 100 per cent pure product. While there may be variations between plants, our experience indicates that invariably a product of a guaranteed purity higher than 99.5 per cent will be more expensive than one meeting this standard.

#### Conclusions

Using the best techniques and maximum care it has been established that increases in oxygen purity from 99.0 up to 99.83 per cent result in linear increases in cutting speed. For  $\frac{3}{4}$ -in. plate at normal flows this represents a  $7.6 \pm 2\frac{1}{2}$  per cent increase in speed as the oxygen purity goes from 99.49 to 99.80 per cent. For 4-in. rolled steel a similar increase in purity results in a  $6 \pm 2$  per cent increase in speed. For the cast steel an increase in purity from 99.5 to 99.8 per cent results in a  $10.8 \pm 2.5$  per cent increase in speed. While the gain in cast steel appears greatest, it is the most difficult one to "pin-point."

Work done by others and reported in private communications at still higher purities also confirms the linear relations formulated in this paper.

At the same time it is more expensive to provide oxygen at a guaranteed purity higher than the present 99.5 per cent standard, all conditions of production being equal.

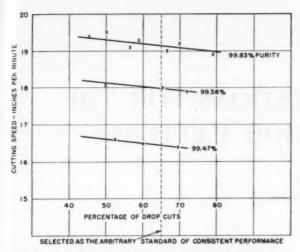


Fig. 12—Determination of drop cut percentages. Style 164 No. 2 tip (.055) orifice; cutting oxygen flow, 145 cfh; acetylene flow, 14 cfh; preheat flames neutral;  $1\sqrt[3]{16}$ -in. thick cast steel keel blocks.

In view of the difficulties of measuring the purity and expense of production, the small economies of using oxygen over 99.5 per cent pure do not seem to justify producing such a grade of oxygen in the present state of the art. Most of the cutting done in a steel foundry is by hand and on risers. For the purpose of this paper it would have been desirable to do the test work by hand cutting and on typical steel risers. However, the operator is such a variable factor that conscientious researchers have long ago ceased to give credance to any tests involving hand cutting unless only the most approximate results were acceptable. Hence, all the work reported in this paper is on machine cutting. As for the possibilities of using commercial cast steel risers as test materials, one has only to examine the way one riser will vary from another as regards surface conditions to realize the difficulties of getting consistent results. For the purpose of this investigation we were fortunate in being able to procure cast steel specimens found under conditions especially controlled to produce pieces as uniform as possible. Even so, the results are not as uniform as on rolled materials.

#### Acknowledgment

The experimental data reported in this paper were accumulated by the Air Reduction Research and Operating Departments. Their help, particularly that of Messrs, J. J. Rendos and G. A. Benz, as well as that of the Technical Sales Department in preparing this paper is gratefully acknowledged.

Special credit is due to Mr. C. W. Briggs of the Steel Founder's Society for encouraging both this research project and the writing of this paper and to Mr. G. A. Lillieqvist of the American Steel Foundries for his advice and the furnishing of the cast steel material.

References

1. The Engineer, London, March 23, 1888, p. 235.

2. The Journal, of the Society of Chemical Industry, March 31, 1890, p. 246-253.

3. The Journal of the Society of Chemical Industry, July 15, 1911, p. 779-782.

 Oxygen Cutting, Slottman and Roper, p. 51, McGraw-Hill, 1951.

#### Appendix

Analysis of Cylinders Used in Plate Cutting and Starting Tests

Methods used:

 The plant analysis involved an absorbing solution of ammoniacal cuprous chloride and a simplified construction using tubing clamps instead of stopcocks.

2. The Research Division's Hempel Pipette was essentially a more complicated version of the plant device. The added stopcocks make manipulation easier but are potential leaks and bubble catchers.

3. The Orsat analyzer involves both ammoniacal cuprous chloride and pyrogallol absorption. The added apparatus gives more possibilities for leaks and bubble losses.

4. The Development Division's Hempel Pipette was a carefully calibrated copy of the plant equipment. It might be added that the manipulation of the rubber tubing in this plant type apparatus requires practice.

### CONSTRUCTION OF THE SPEED-PURITY GRAPH FOR CAST STEEL (Fig. 11)

The values shown in Fig. 10 were used in the following manner:

1. For each of the three oxygen purities a specific base speed was selected with a range of  $\pm 0.1$  in./min, and the ratio of drop cuts to total cuts was calculated and expressed as a percentage. The initial base speeds and ranges were: for 99.83 per cent purity,  $19.0\pm0.1$  in./min; 99.58 per cent purity,  $18.0\pm0.1$  in./min; and for 99.47 per cent purity,  $16.5\pm0.1$  in./min.

2. This procedure was followed by varying each base speed 0.1 in./min both above and below the specific speeds initially chosen.

3. The percentages of drop cuts versus base speeds were plotted as shown in Fig. 12.

By inspection of Fig. 12, the 65 per cent drop cut line was chosen as the standard consistency and the points of Fig. 11 plotted on that basis.

Some other percentage line could have been selected but the end result would have been changed only a negligible amount because:

1. The slopes of the three purity lines of Fig. 12 are almost identical.

2. The change in speed with increasing drop cut percentage is small within the region plotted.

3. If the speeds plotted in Fig. 11 are given a ±0.2 in./min range, the performance consistencies for all purities tested fall within the range of 60 to 65 per cent. In addition, this seems a reasonable range to take care of variations in the apparatus and the precision of data recording.

## PRACTICAL SUGGESTIONS FOR THE BUILDING OF WOOD PATTERNS

W. H. Siebert*

· Machine tools are much needed today and the demand for machine tool patterns is great. The large number of castings required for many types of machines coupled with the complexities of modern design challenge patternmakers and foundrymen to strive for better pattern equipment that helps the molder produce more and better castings. Large patterns are quite often thought of in terms of cheap patterns; never can enough emphasis be placed on the fact that a pattern must be built substantially, and to foundry requirements.

Therefore it seems only common sense to use the ideas which follow. Even if a pattern is to be made for only one casting, they still hold true.

#### Not Just Guesswork

To produce a satisfactory pattern we must teach young men that there is more to modern patternmaking than merely haphazard guesswork as to how a good pattern should be built. The patternmaker of today must be able to visualize everything that goes into the making of a large piece of equipment. In order to do this he must have a working knowledge of a designer, a machinist, and above all a molder and coremaker, so that with the close cooperation of these men plus skilled know how, he may achieve the final result of a usable pattern.

In this report we would like to confine ourselves to the construction of a pattern that will make a casting for a specific large machine tool casting. This casting is a typical example of the production requirements stated above. The large number of castings required makes it mandatory to engineer this equipment to provide the foundry with a pattern that enables mold after mold to be made with a minimum of effort, and to construct the pattern in such a way as to avoid delays for repairs as much as possible.

With this in mind, we suggest a few details of making such a pattern. The first step should be to become thoroughly familiar with what the designer is trying to achieve, determining the finished surfaces most critical, dimensions most critical, oil or water tight compartments, etc. These facts may help in determining the best possible surface to have in the drag, and perhaps in arranging the cores in such a manner as to do away with the need of chaplets where

a leak proof casting surface is required.

This accomplished, the next step should be to consult the foundryman who is to make the casting. It would indeed be a costly error to neglect this allimportant step. Assuming that he has to produce 50 good castings, he will be of utmost value in helping design the pattern so that he will be able to make these castings to the best satisfaction of the customer and meet a reasonable time schedule. His advice will be an important factor in determining molding methods, flask sizes, sufficient core supports, core vents, risers, and gates. Most of all he will aid in designing the cores to enable him to set them without costly delay. His core oven capacity may be a factor in determining the size of the cores. In places where two cores are separated by a wall section, it is sometimes possible to make these two cores into one simply by one or two permissible holes through this wall section. The saving is quite obvious. There is one less core box to make, one less core, and one less set for the molder. Perhaps one of the more common applications for adding holes is on large cope surfaces where ordinary chaplets are either objectionable or impractical. The molder very often insists on them. In cases where the cope side of the mold is also the bottom of a machine that has tanks and other compartments incorporated in the design, openings, if held to a proper diameter, can be welch plugged in the machine shop and will prove quite satisfactory. It is very bene ficial to the molder to have as many holes as possible in the cope surfaces, for they afford an excellent place for venting out core gases and prevent any possibility of cores raising to the cope.

After proper planning, and after all necessary layouts have been made (normally a typical cross section) begin construction of the pattern. The first concern must be the frame work, or heads. These must be set up on an absolutely flat surface. If it is desired to have a pattern that is accurate, all precautions must be observed from the very beginning.

Use this hard and fast rule in selecting lumber: do not skimp on thickness. Patternmakers must bear

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in mind that the modern foundry employs slingers for ramming, which are quite capable of bowing a 11/8 in. pattern lagging enough to render a draw impossible. Therefore heads must be sturdy. Sixteen in. should be the maximum spacing of heads.

After all the heads are erected, tie rails are placed. This is best accomplished by screwing them directly to the outside of the heads. It is the author's opinion that the common practice of recessing the rails into the heads tends only to weaken the head structure. This is a frequent cause of serious pattern damage. The rails should be at least  $1\frac{3}{4} \times 6$  in., so placed that the outer lagging can be fastened to them with the grain of the wood following the draw of the pattern.

However, before the outer lagging can be applied, enough inner cross bracing must be installed to insure a rigid pattern. Good judgment must be employed in this so as not to increase the weight of the pattern excessively. To facilitate bolting or screwing of rapping or vibrator plates to the pattern, sufficient stock must be placed at proper locations between the heads.

Next in line is the matter of draw straps, which also should be installed before any lagging is applied. These straps should consist of at least 2½ x ¼ in. cold rolled steel. They should be hooked to the heads and bolted for maximum strength. These straps should be kept flush with the face of the pattern, and not project above as is so often the case. A 1-in. hole drilled in this steel strap will serve for handling purposes. If the pattern is to be rolled with the flash, handling holes must also be provided both on the cope and drag side of the pattern. Hooks should be flush with the pattern surface.

#### Now Enclose the Pattern

Now proceed to finish enclosing the pattern. If possible use 13/8 in. stock for lagging, as it will help strengthen the job and avoid bowing by the slinger. These laggings can either be nailed or screwed to the rails. Consider at this point all external parts of the pattern, such as core prints, bosses, large round corners, fillets, etc. These may well be incorporated in plans for lagging.

Avoid loose pieces where possible. If this is impossible, it is best to fasten them to a dovetail. Pieces attached with loose pins often are rammed off their particular location.

It is very difficult to remove large, loose pieces. In order to aid the molder in producing a better mold, quite often loose pieces can be made in ram-up cores. This will aid greatly in finishing the mold, as the hardto-get-to places are already blackened on the ram-up core in the core room.

Coreprints should be sufficiently large to carry the weight of the core and permit solid anchorage. In the event the core must be tied down in the mold, enough space must be given on the face of the coreprints to permit the coremaker to provide the proper tie rods.

The patternmaker must also be mindful of print sizes to prevent the possibility of the molder placing a core in the wrong print or of placing a core wrongly in its own print. He uses core markers where necessary, and makes them large and distinct. A patternmaker should work on the assumption that if a molder can err in locating a core, it is the fault of the pattern design.

Pattern draft is of course of prime importance; lack of it may cause needless patching. Usually 3/16-in. draft for every two feet of pattern surface is sufficient on any side surface. However, a good practice is to have all the draft that the part design will permit. If it becomes necessary to have recesses in the cope surface of the pattern, use as much draft as possible. The lack of sufficient draft is perhaps the most common error a patternmaker will make.

The construction of coreboxes requires as much consideration as the pattern itself. It should likewise be considered from the foundry angle. An important factor to the coremaker is the outer shape of the corebox. No matter what shapes or angles the core itself may take, the box should always be started as a rectangular frame of sturdy construction. That is the easiest shape to roll over and draw. The sides should be of 2-in. stock with a ½-in. gaining in of the ends. The bottom can be made of thinner stock, depending on the size of the box.

Since the core may take many shapes, and incorporate large bosses, ribs, etc., the patternmaker must try to picture in his mind what the coremaker has to do to make the core. Good judgment may save needless work. It is often impossible to ram under large, loose pieces. In such instances, make it possible for the coremaker to sweep up a flat surface under the loose piece. This can often be accomplished by making the box one inch wider per side and adding 1-in. stock up to the level of the loose piece. This will serve a twofold purpose. First it will provide a seat for the loose piece; second, it will furnish a surface on which a sweep will ride to establish the lower level of the loose piece. After the coremaker has sweep this sur-

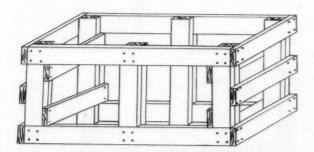


Fig 1 . . . Illustrating proper construction method for heads. Frame must be set on an absolute flat surface.

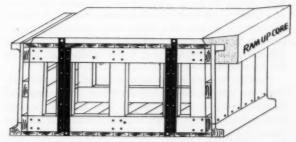


Fig. 2 . . . Section of completed pattern, showing proper use of ram-up cores. Notice use of draw straps.

face, he has only to set the loose piece on it and finish the ramming operation.

It is also wise to remember that large cores may settle, thus making the side larger and subsequently giving the molder thinner metal walls. This can be avoided by making the box smaller than the actual measurements. It is good practice to make the inside dimensions of the corebox with the standard rule. As in many foundry practices, no hard and fast rules are applicable. It is good to discuss with the foundry such cases where cores are exceptionally high or large. In case a patternmaker must use his own judgment on such cores, 1/16 in. per foot of width and length may be deducted from the actual size where cores are more than 2 feet high, using the shrink rule. A standard rule may make core slump allowances too great on larger cores. It is, of course, the height factor that more than anything else causes core growth.

Too little attention is often given to the outer shell of the box. A good battening job helps preserve it. Remember the box plus sand, core rods, and core plate is a lot of weight to roll over. This in itself is a good reason to make the core box as strong as possible. Never glue and nail battens; screw them on.

Now that the pattern and boxes are completed, there remains one more thing to do. After proper application of a pattern coating, mark up the equipment in such a way that the foundry has no trouble in setting up the job. This identification can best be accomplished by using embossed zinc tape; painted-on setups are soon rendered unreadable by service.

Successful completion of a pattern often can only be accomplished by cooperation among all involved. Design, patternshop, foundry, and coreroom all play an important part in pattern engineering. By utilizing all these resources, the patternmaker can produce a pattern that will stand up under the roughest treatment and yet turn out casting after casting which meet the highest quality standards.

It would be embarrassing to have a pattern come out of a mold in pieces. Equal embarrassment can result if the foundryman declares that the already constructed pattern can't be used unless a costly rebuilding job is done. Several factors can contribute to a faulty pattern. First, there is a lack of foundry knowledge on the part of some manufacturers of patterns. Second, in cases where the purchaser specifies a well-made pattern, the producer may deliberately slight the construction in order to save expenses. Third, purchasers of patterns may allocate an insufficient amount of money for the cost of a pattern, causing the same bad results. It seems false economy to make the price of a pattern the most important factor and then build the pattern to suit this price. The resulting cheaper pattern would have a decidedly shorter life. Such a pattern could quickly cost many additional dollars for rebuilding and repair costs and down time.

Pattern manufacturers should strive to build patterns to meet foundry requirements. Following the factors outlined in this paper will lead to strong, serviceable patterns which will cost more but will return this extra expense in castings at less cost per pound. Patternmaking is one of the few highly skilled trades. It should be applied to the fullest extent. This and nothing else leads to successful patternmaking.

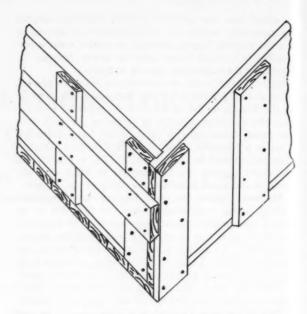


Fig. 3 . . . Detail of outer construction of core boxes.

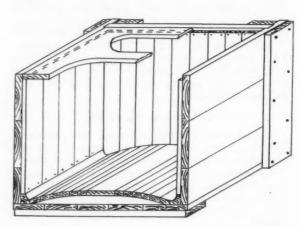


Fig. 4... Cutaway section of core box showing inner lining. This lining has two purposes; it provides a seat for the loose piece, and it furnishes a surface on which a sweep will ride to establish lower level of piece.

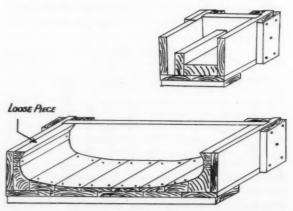


Fig. 5 . . . Shows construction of ram-up and drag cores.

#### DISCUSSION

Chairman: V. J. SEDLON, Master Pattern Co., Cleveland. Co-Chairman: H. C. Swanson, Arrow Pattern and Foundry. Chicago.

Recorder: F. J. OKLESSEN, Motor Patterns Co., Cleveland.

RECORDER OKLESSEN: I merely want to re-emphasize some of the important points brought out by the author. The first step in making pattern equipment is the analysis of the drawing. The drawing must be practical where close tolerances are to be held. The patternmaker must show what faces are critical in regard to sound metal.

The second step is to confer with the proper foundry personnel on how the mold will be gated, how the cores are to be supported in the mold and what the size of coreprints should be.

The third step in making the pattern is to layout the pattern on the board. When constructing the headers, proper thickness and cross bracing is necessary. Draw bars on large patterns are a must. These straps should be bolted down not screwed. The pattern facing lumber joints should be in the direction of the draw. Avoid loose pieces on patterns. When a projection occurs on a pattern a core should be placed under the loose piece, said loose piece should be dove-tailed in the pattern. Core box construction should be started with a straight box frame. Patterns should not be glued on but screwed on. Feather edge core fillets should be avoided. All corners should be of solid construction. The insides of the coreboxes can then be readily dropped into place and held securely by means of screws. This type of pattern equipment will give good service, little maintenance, and a good acceptable casting for the foundry. Here are three reasons why bad patterns are made: (1) lack of knowledge of pattern by patternmakers, (2) some pattern shops deliberately short cut on jobs, and (3) pattern buying by the purchasing agents strictly on price comparison only.

MEMBER: What pattern finishes does the author recommend? MR. SIEBERT: We use iron oxide with shellac. It is a rougher

finish, dries slower but gives a hard final finish. JOHN GIANO: 1 Our practice on machine tool patterns is to apply one coat of shellac followed by two coats of lacquer.

MEMBER: Is the hardwood frame necessary?

MR. SIEBERT: No, unless wear surfaces get much abuse.

F. C. CECH: 2 What lumber is used in this work?

MR. SIEBERT: We use Idaho pine, cherry and ash. Do not use sugar pine because it is too soft.

E. J. McAfee: Do molders use the rapping plates provided? MR. SIEBERT: The molders should use the rapping plates provided because that prolongs the life of the pattern.

Royal Pattern Works, Cleveland.
 Cleveland Trade School, Cleveland.
 Puget Sound Naval Shipyard, Bremerton, Wash.

# USE OF ALUMINUM BRONZE IN HIGH PRESSURE CASTINGS

By

Thornton C. Bunch* and George E. Dalbey**

#### ABSTRACT

Consideration is given to the possibility of substituting aluminum bronze castings for brass forgings and for gun metal in certain high pressure heavy castings.

Gun metal has the inherent property of solidifying with a dispersed pipe. Aluminum bronze solidifies with a concentrated pipe.

#### Introduction

The present day accelerated industrial economy requires a dynamic foundry industry prepared to meet ever increasing engineering demands.

The large number of non-ferrous alloys in industrial use indicates there is no one best alloy for all purposes.

The selection of an alloy for a specific use usually results in a compromise of mechanical properties, chemical compositions, ease of processing, availability and cost.

Consultation between the designing engineer, foundryman, patternmaker and metallurgist usually results in a product having the highest average properties, the lowest cost and a minimum of production difficulties.

#### Metallurgical Considerations

Gun metal having a nominal composition of copper 88.0 per cent, tin 8.0 per cent and zinc 4.0 per cent and minimum mechanical properties as follows: yield strength 20,000 psi, tensile strength 40,000 psi, and elongation 20 per cent, is an alloy that has been extensively used and mis-used. This alloy is easily handled in the foundry. The gating is not too critical and some turbulence can be tolerated in pouring. This alloy will generally make a satisfactory pressure-tight casting if the casting walls are relatively thin and the walls are left unmachined. This same alloy when used in a heavy chunky casting, a casting on which considerable machining is to be performed, and

in the center of which several holes may be drilled close together, leaving only a thin shell of metal between the holes, may produce an unsatisfactory casting (Fig. 1). A less massive casting, but having considerable machining on it, is a border line casting when cast of gun metal. This type of casting is shown in Fig. 2.

Gun metal is a wide range solidifying alloy. It solidifies by the formation of the six branched primary dendritic skeletons in the building of the octahedral crystalline grains.¹ During solidification, the growing grains are composed of skeletons of primary, secondary and tertiary branches surrounded by liquid metal, and there is but little thickening of these branches until late in the solidifying process. The inherent result of this method of solidification is the formation of fine intradendritic spherical porosity and fine scattered tortuous interdendritic porosity.^{3,5} This type of porosity is difficult to eliminate completely by feeding. The presence of gas in the metal aggravates this condition.²

Aluminum bronze having a nominal composition of copper 87.5 per cent, aluminum 9.0 per cent, iron 3.5 per cent, and minimum mechanical properties as follows: yield strength 25,000 psi, tensile strength 65,000 psi, and elongation 20 per cent, is an alloy that might be substituted for brass forgings, for gun metal in massive castings, and for gun metal in castings where considerable external and internal machining is to be performed.

The gating of castings produced from aluminum bronze is critical as little if any turbulence can be tolerated in pouring, because this alloy readily forms large amounts of dross. This alloy will generally make a pressure tight casting if properly handled in the foundry.

Aluminum bronze is a narrow range solidifying alloy. During solidification the grains are composed of skeletons of primary and secondary branches surrounded by liquid metal. Thickening of the primary and secondary branches occurs early in the process of solidification, resulting in the formation of a solid compact skeleton rather than a thin extended skeleton growth. The inherent result of this method of solidification is the formation of a solid wall with

^{*} Master Molder and ** Metallurgist, Mare Island Naval Shipyard, Vallejo, Calif.

The opinions or assertions contained herein are the private ones of the writers and are not to be construed as reflecting the views of the Navy Department,



Fig. 1-A finished high pressure manifold casting.



Fig. 2-A finished high pressure cylinder head casting.



Fig. 4—High pressure cylinder head castings showing risers and gating. Left, gun metal; right, aluminum bronze.



Fig. 3—A high pressure manifold casting showing riser and gating.

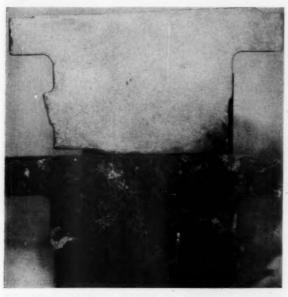
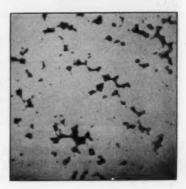


Fig. 6—Photograph of 1/4 inch slabs taken from mid-section of high pressure manifold castings and etched with strong nitrohydrochloric acid to show the difference in structure and density between gun metal and aluminum bronze.

Both were etched in the acid for same length of time.

Bottom, gun metal; top, aluminum bronze.

Fig. 7—Photomicrograph of a portion of the specific gravity specimen from the high pressure gun metal manifold, at X25 unetched, showing considerable interdendritic porosity. This metal leaked.



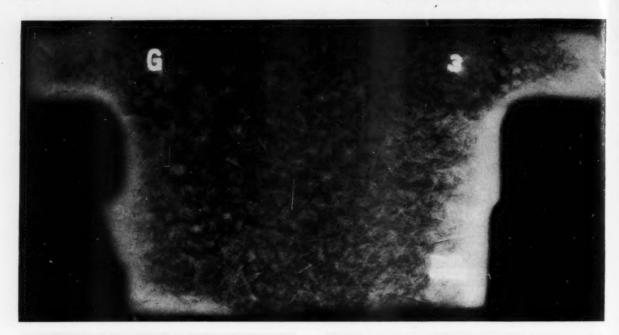


Fig. 5-Radiograph of 1/4 inch thick slab, a mid-section of a gun metal high pressure manifold casting.

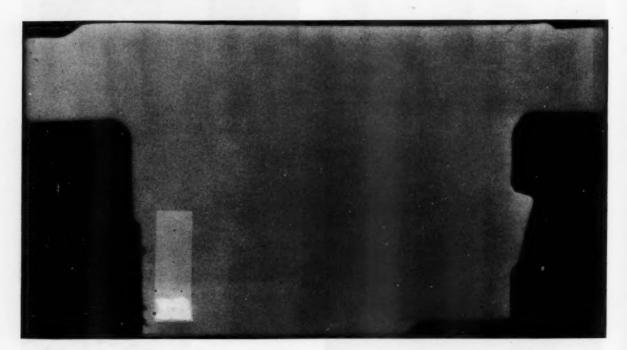


Fig. 8-Radiograph of 1/4 inch thick slab, a mid-section of an aluminum bronze high pressure manifold casting.

little if any interdendritic porosity. This alloy can be expected to withstand high air and hydrostatic pressures in the as-cast state.⁴

#### Corrosion

Corrosion is another factor which must be taken into consideration when selecting an alloy. The following quotation from Frank L. LaQue's memorandum indicates that G-bronze and aluminum-bronze are not too far apart in general corrosion properties. "You will note that in quiet water there is no substantial difference between the two compositions of bronze. However, where erosion is involved, the available data indicates that the aluminum bronze will be somewhat inferior to the G-bronze and that the erosion resistance of the aluminum bronze, as is the

case with other copper alloys improves as the iron content of the alloy is raised. Along the same line is the fact that one of the most erosion and corrosion resistant of the common bronzes is the nickel aluminum bronze which contains 10 per cent aluminum, 5 per cent nickel and 5 per cent iron."

"In erosion tests in another type of machine at Harbor Island involving velocities of 27 ft per sec this aluminum-nickel-iron bronze suffered a weight loss of only 93 mg/sq dm/day as compared with a representative value of 218 mg/sq dm/day for composition G-bronze under the same exposure condition."

In specific applications where resistance to erosion is required, the high nickel aluminum bronze may be substituted for G-bronze, and aluminum bronze.

#### **Experimental Procedure**

Castings—To get an insight into how casting size and/or a change in alloy will affect the properties of a casting, the following type castings were cast of gun metal and aluminum bronze.

(a) A heavy, chunky, high pressure manifold casting as illustrated in Fig. 1.

(b) A smaller, lighter, high pressure cylinder head as illustrated in Fig. 2.

Gating—The gating on the high pressure manifold was similar for both alloys. An inverted horn gate was used, the small end acting as a choke and the large end attached to casting. The area of the small end was 0.25 sq in., and the area of the large end

was 5.05 in. It was expected this gating arrangement would minimize the turbulence. The riser was adequate, probably larger than necessary (Fig. 3).

The gating method used for the high pressure cylinder head casting is shown in Fig. 4. The casting on the left is gun metal, that on the right, aluminum bronze. Both castings have generous risers.

Laboratory Tests-Cross-section slabs 1/4 in. thick were removed from the mid-section of each casting. The slabs were radiographed and pressure tests made with compressed air at 90 lb. The slabs were deep etched with nitrohydrochloric acid as a check on the metal soundness.

Specific gravity was determined on specimens taken adjacent to the center of the castings. Microscopic examinations were made on specimens taken from the specific gravity specimens.

#### **Experimental Results**

(1). Heavy high pressure manifold casting.

(a) Gun Metal—The radiograph of the 1/4-in. thick section disclosed wide spread interdendritic porosity except for a shell of metal that was in contact with the mold surface (Fig. 5).

The same section, when put under a 90-lb com-

pressed air test, leaked badly.

The same section when etched with nitrohydrochloric acid showed much the same as the radiograph.6 (Fig. 6, bottom.)

The specific gravity of a specimen adjacent to the

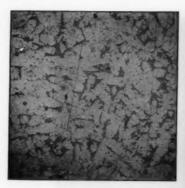


Fig. 9-Photomicrograph of a portion of the specific gravity specimen from the high pressure aluminum bronze manifold at X25 unetched, showing a few scattered gas holes. The very fine iron compound is well distributed throughout the alpha matrix. The beta constituent is largely resolved into duplex eutectoid structure. This metal did not leak.

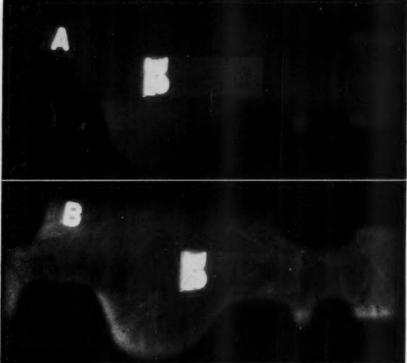


Fig. 10-Radiograph of 1/4 inch thick slabs taken from the mid-section of high pressure cylinder head castings. Note the difference in structure and density between gun metal and aluminum bronze.

center of the 1/4-in. slab was 8.40.

The photomicrograph of a portion of the specific gravity specimen at X25 unetched, shows considerable interdendritic porosity (Fig. 7).

(b) Aluminum Bronze—The radiograph of the 1/4-in. thick section shows a uniformly dense metal with no evidence of a denser metal shell. There are some small scattered spherical holes (Fig. 8).

The same section when put under a 90-lb compressed air test did not leak. The section was then reduced to ½-in. thickness and again tested with 90-lb compressed air. The section did not leak.

The same section when etched with nitrohydrochloric acid showed much the same pattern as the radiograph (Fig. 6, top).

The specific gravity of a specimen adjacent to the

center of the 1/4-in. slab was 7.83.

The photomicrograph of a portion of the specific gravity specimen at X25 unetched, shows a few scattered gas holes. The very fine iron compound is well distributed throughout the alpha matrix. The beta constituent is largely resolved into duplex eutectoid structure (Fig. 9).

(2). High pressure cylinder head casting.

(a) Gun Metal—The radiograph of the ½-in. thick section discloses a large crystalline pattern with some scattered low density areas. There is a shell of high density metal that was in contact with the mold surface (Fig. 10, top).

The same section when put under a 90-lb compressed air test did not leak.

The same section when etched with nitrohydrochloric acid showed much the same pattern as the radiograph (Fig. 11, bottom).

The specific gravity of a specimen adjacent to the

center of the 1/4-in. slab was 8.68.

The photomicrograph of a portion of the specific gravity specimen at X25 unetched shows considerable interdendritic porosity, but less than in the heavy casting (Fig. 12).

(b) Aluminum Bronze—The radiograph of ½-in. thick section shows a uniformly dense metal with no evidence of a denser metal shell. There are a few small scattered spherical holes (Fig. 10, bottom).

The same section, when put under a 90-lb com-

pressed air test, did not leak.

The same section, when etched with nitrohydrochloric acid, showed much the same pattern as the radiograph (Fig. 11, top).

The specific gravity of a specimen adjacent to the

center of the 1/4-in. slab was 7.85.

The photomicrograph of a portion of the specific gravity specimen at X25 unetched was similar to that shown in Fig. 9.

#### Summary and Conclusions

Gun metal is sensitive to section size. Thin walled castings are more apt to have denser, tighter metal than heavier castings.

The type of gas and shrinkage voids produced in gun metal castings are more apt to produce leaky



Fig. 11—Photograph of mid-sections of high pressure cylinder head castings, etched with strong nitrohydrochloric acid, showing difference in structure and density between gun metal and aluminum bronze.

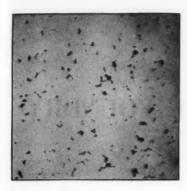


Fig. 12 - Photomicrograph of a portion of the specific gravity specimen from the high pressure gun metal cylinder head at X25 unetched, showing considerable interdendritic porosity.

castings than are the voids produced in a similar aluminum bronze casting.

The heavy chunky castings which are difficult to produce pressure tight in gun metal can more often be made successfully in aluminum bronze.

There is a marked density difference in heavy gun metal castings between the surface and center. Removal of the surface by machining removes the strongest and soundest metal.

The aluminum bronze castings do not show the marked density difference between surface and center of casting, and metal can be removed by machining without sacrificing quality.

The aluminum bronze must be gated in a manner to have non-turbulent flow, in order to produce drossfree, pressure-tight, sound castings.

The uniformly dense structure of the aluminum bronze casting indicates a possible substitution of aluminum bronze castings for brass forgings.

#### Acknowledgment

The writers wish to express their appreciation to the metallurgical and x-ray sections of the Industrial Laboratory, and to the Photographic laboratory, for their assistance in the preparation of the photomicrographs, the x-ray films and the photographs. Photographs contained in this paper are official Armed Services photographs released by Mare Island Naval Shipyard, Vallejo, Calif.

#### **Bibliography**

1. S. L. Finch, "Some Notes on Feeding," Foundry Trade

Journal, July 31, 1947, vol. 82.
2. Clyde L. Frear, "The Effect of Gases on Tin Bronze," The Foundry, Nov. 1948, vol. 76, p. 70.

3. Clyde L. Frear, "Solidification Mechanism of Tin Bronzes."

The Foundry, April 1948, vol. 76, p. 68.
4. Harold J. Roast, "Why Not Use More High Shrinkage Bronze?", The Foundry, November 1949, vol. 77, p. 93.

5. B. N. Ames and N. A. Kahn, "Effect of Superheating and Casting Temperatures on Physical Properties and Solidification Characteristics of Tin Bronzes," Transactions, American Found-rymen's Society, vol. 58, p. 229 (1950).

6. W. F. Graham and L. A. Meisse, "Deep-Etch Test of Brass," Transactions, American Foundrymens' Society, vol. 38, p. 810 (1930).

#### DISCUSSION

Chairman: W. A. MADER, Oberdorfer Foundries, Inc., Syracuse, N. Y.

Co-Chairman: B. W. Schaffer, Kuhlman Electric Co., Bay City, Mich.

Recorder: C. P. KOTOWICZ, Ampco Metal, Inc., Milwaukee.

R. A. COLTON: 1 Mr. Dalbey demonstrated you can make sounder castings in aluminum bronze than in tin bronze but in his opinion he felt gating should not have been the same for both metals.

MR. DALBEY: The only reason gating was identical in both aluminum bronze and tin bronze is because we wanted conditions to be identical in both metals.

S. W. CHAPPELL, JR.: 2 We are making the same manifold as shown in the paper with a heavy chill in aluminum bronze. However, a runner box is used in pouring operations which is filled and then the plug is pulled. We are handling aluminum bronze the same as gun metal.

MR. BUNCH: During the experiments conducted, the one thought we had in mind was of obtaining sound, dense material, because of time involved in machining, pressure checking, etc. Aluminum bronze is a much better alloy because it can be readily welded and fed with exothermic materials. Many castings in metals other than aluminum bronze had been made previously as forgings because of high pressure requirements.

MEMBER: We are interested in aluminum bronzes particularly in the aluminum nickel bronzes which are used successfully for gears on submarines. I am surprised that the authors did not use the Durvill system of gating (Fig. 3). Attention must be paid not only to the turbulence and dross usually found in aluminum bronze, but care must be taken so that melting conditions are

W. B. Scott: 2 The Navy should be cheered for changing over to aluminum bronze in pressure castings. Leakers usually occur after improper heat treatment.

MR. CHAPPELL: The top gating I mentioned previously was on dry sand only. A plug 3 in. in diameter is used with the runner box whenever we can work green sand molding. Naturally, we must be more careful in gating somewhat differently because of the turbulence created in aluminum bronze and the possibility of sand washing.

MR. BUNCH: Whenever aluminum bronze is poured at the Mare Island Naval Yards it is poured as slowly as possible or under controlled pouring conditions.

Conn.

³ Metallurgist, American Brake Shoe Co., Meadville, Pa.

¹ Research Metallurgist, American Smelting & Refining Co., Barber, N.
² Supt. of Foundries, Electric Boat Div., General Dynamics Corp., Groto

### OLIVINE APPLICATION IN THE FOUNDRY

By

Gilbert S. Schaller*

The present research program on olivine application in the foundry has been undertaken by the foundry division of the Mechanical Engineering Department, University of Washington, for several reasons. A compelling motive is the present national preparedness situation since west coast foundries are dependent largely upon Ottawa silica or that imported from the Belgium beaches. A curtailment of either of these supply sources would develop a genuine, if not an insurmountable, emergency to the production of castings in the Far West.

A second reason for the research program is the abundance of olivine in Washington State where large known deposits of high quality olivine occur, indicating that no economic obstruction to the use

of olivine in the foundry exists.

Finally, it was desired to discover any properties of olivine sand aggregate that would be of special interest. No attempt has been made to compare olivine with silica on the basis of superiority of one over the other. Investigation has been slowed because accepted testing and evaluation procedures have been developed for silica primarily whereas olivine has different physical, chemical, and mechanical properties which do not, in every case, lend themselves to established testing routine. It has been found necessary, to develop distinctive procedures in order to evaluate olivine molding sand aggregate. It seems safe to predict that the introduction of olivine in the foundry will be characterized by misgivings and failures because olivine will be expected to conform in its behavior to long-established foundry practice. Information received from the Olivine Committee of M. V. L. proves this situation in a report, "The Use of Olivine Sand in Steel and Iron Foundries."1

#### Properties of Olivine

Olivine is magnesium-iron-ortho-silicate that is composed of forsterite (Mg₂SiO₄) and fayalite (Fe₂SiO₄). It does not contain free silica. Olivine in Washington occurs as rock or boulders in its natural state and must be crushed prior to any foundry application. It occurs in the Twin Sisters area in the foothills of the



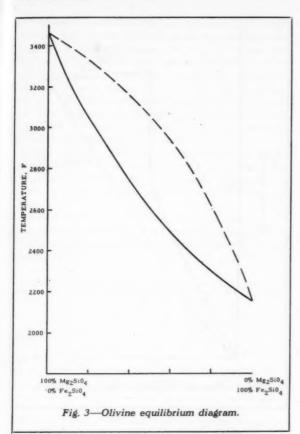
Fig. 1—General view of the Twin Sisters area, Cascade Mountains.

Cascade range. The deposits are extensive² and capable of long-time production. Another deposit is located on Cypress Island in the San Juans. This occurence is not of the same quality as the Twin Sisters deposit because of a somewhat higher fayalite content. However, its geographical location is more favorable since it is accessible to tidewater and barge transportation.

Thermal Properties: The fusion temperature of olivine is dependent upon the proportions of its constituents. The constitution diagram of the olivine system is shown in Fig. 3. It can be seen that pure forsterite has a fusion temperature above 3400 F. Olivine used in our work contains approximately 7 per cent fayalite and has a fusion point approximating 3300 F. Obviously, its fusion point is sufficiently high to meet foundry casting temperatures. No difficulties were encountered in producing non-ferrous as well as gray iron castings. Some steel castings were made, but the scope of that operation was limited due to a lack of adequate steel melting equipment in our foundry. Olivine sand is reported to be producing steel castings successfully in Norway.3 Manganese steel castings of excellent quality are reported from Norwegian foundries, as well as the Electric Steel Foundry Co., Portland, Ore.

Expansion Characteristics: Olivine has a lesser thermal expansion rate than quartz as reported by

Professor of Mechanical Engineering, University of Washington, Seattle.



Bowin and Schainer (Fig. 4). A lower thermal expansion rate is a desirable characteristic since it indicates that molds made from olivine foundry sand aggregate should exhibit low buckling and scabbing tendencies. Our experience indicates such a condition, although buckling was not extensively investigated. It is inaccurate to connect these qualities to the sand aggregate alone since there are other variables that influence scabbing and buckling tendencies. Olivine does not have the inversion factor that occurs in quartz at 1090 F. Inversion is accompanied by a sudden change from rapid expansion to slow contraction which is a major contributing cause to silica expansion defects.

Thermal Conductivity: The tendency for a casting to scab or burn-on can be associated, in a degree at least, with the rapidity of skin formation at the metal mold interface. Lock and Ashbrook⁴ report that olivine has a thermal conductivity about midway between silicon carbide and silica. A high rate of thermal conductivity is a desirable mold property, especially where controlled solidification as in casting zinc base-forming dies⁵ is desired.

Physical Properties: The specific gravity of olivine lies in the range of 3.27 to 3.37. By comparison, quartz has a specific gravity of 2.65. In consequence, olivine is approximately 25 per cent heavier than quartz. Such a condition appears to be unfavorable to olivine since more weight for equal volumes of present molding sand is indicated.

A heavier mold aggregate has its favorable aspects. Jolt ramming is aided due to increased sand weight. There is good reason to believe that larger molds will require fewer bars and less gaggering. In fact, it appears entirely logical to re-examine flask size where heavier olivine is molded since it seems possible to use less sand volume to equal the weight of established molding sand. In the general run of hand molding, olivine shows to disadvantage due to its greater density.

Chemical Properties: Olivine is considered to be a basic refractory. Many clays common to the foundry tend to flux olivine. Both aluminous materials as well as silica react with olivine. The nature and severity of these reactions are dependent upon temperature conditions. In some Norwegian iron foundries, olivine is introduced as heap additions. A continuation of this program results in a gradual change over to a complete olivine sand aggregate.

In our work we employed western bentonite for bonding the ferrous sands and southern bentonite for the non-ferrous heaps. No evidence of fluxing action was noted. Insufficient work was done to determine what effect acid steel slag might have on olivine sand aggregate. This is a situation that is deserving of close investigation where an olivine molding sand aggregate is contemplated for steel foundry application.

Hygienic Considerations: Replacement of established foundry sand aggregates with olivine has some technical advantages. For the most part, though, a change to olivine has been considered primarily because of the potential prophylactic advantages to be

Fig. 2—Formation showing source of olivine used in the research program.



gained thereby. H. J. Weber⁶ states that silicosis is caused by the inhalation of free silica dust in the mutual presence of three factors:

1. Silica particles must be smaller than five mi-

crons.

Silica dust concentration must exceed five million particles per cu ft per 8-hr day.

3. Exposure time under the above listed conditions must be extensive.

Olivine does not contain free silica and, for that reason, it offers one possible solution to silicosis difficulties in the foundry.

There are many references in both industrial as well as medical literature indicating a favorable position for olivine as a preventative for silicosis. Examples of such publication are "Olivine Synthetic Molding Sand Controls Silicosis" and "The Properties of Olivine and Its Use for Refractories and Molding Sands." Among extensive medical references, "The Effects of Olivine on the Lungs of Rats" exemplifies the type of research being pursued in evaluating olivine's toxic potential. The increasing interest in this subject is not confined to any one community nor to this country; Norwegian foundries are especially concerned.

The present research program in the application of olivine in the foundry was inaugurated Aug. 15, 1950. All molding and core sands were removed and foundry floors vacuum cleaned. It was desired to avoid all possible contamination from any previously used materials. The program anticipated a complete change-over to olivine in every applicable category. It was further determined that olivine-based sand would be used for the regularly scheduled instruction program. In consequence, the foundry has served in the dual capacity of a research base as well as for student instruction.

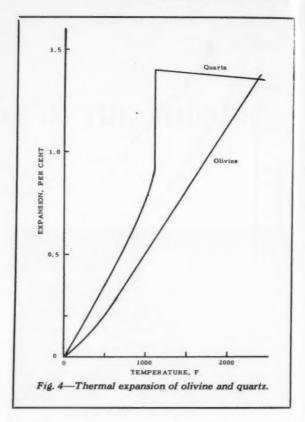
#### Grain Size Classification

Olivine Preparation: The initial supply of olivine was received from the mine crushed to pass 1/4-in. mesh. Further crushing at the School of Mineral Engineering produced an aggregate that would all pass a 20-mesh screen. Air classification was used to remove the fines prior to introduction into the foundry. Where research requirements dictated a narrow spread of grain size, water-classified material was employed. Sand for the iron foundry was compounded from 3 per cent western bentonite and the de-dusted olivine aggregate. The mixture was mulled dry for about 1 min, when 3 per cent water was added and mulling continued for an additional 3 min. A production-type muller was used in this and all subsequent mulling operations.

The screen analysis of the original olivine aggregate is given in Table 1. Bond and moisture per-

TABLE 1-IRON HEAP SAND SCREEN ANALYSIS

	Retained on U. S. Screen, %										
	20	30	40	50	70	100	140	200	270	pan	
Before De-dust After	0.2	12.7	25.5	17.3	11.3	9.2	7.6	6.8	2.9	6.5	
De-dust	0.4	23.2	36.4	18.1	8.5	4.6	3.2	2.0	0.9	2.7	



centages were chosen arbitrarily since there were no authentic data available as a precedent. The molding sand had a permeability range from 100 to 130, with

green strength averaging 7 psi.

Gray iron castings poured during the initial heats were not of ideal quality. Considerable surface roughness occurred and blows were in evidence. Castings were difficult to clean of adherent sand. Surfaces were closely examined in order to study their nature. A surface test procedure was developed that is described in detail under the title "Olivine Molding Sand." Adherent surface scale resulted from mechanically-locked sand rather than from sintering. The absence of sintering was confirmed by sectioning specimen castings having sand inclusions. Atmosphere agents proved beneficial to the elimination of sand encrustation. Shake-on graphite was used, and in some molds a facing containing oilless core binder was introduced with encouraging results.

Sand workability was not of the best. Low moisture content to provide easy ramming resulted in a brittle sand with dry and crumbly mold edges. Increased moisture improved these difficulties, at a sacrifice of casting surface condition. Hard ramming improved casting surface but caused difficult pattern draw.

Dry strength proved to be abnormally high causing difficult shakeout. This difficulty became less noticeable after several heats had been poured. Loss in dry strength was accompanied by increasing green strength. The sand was mulled one or more times between each foundry heat.

After a year of operation no bond has been added

Table 2.

to the heap sand even though 24 heats have been poured. All of these were by students, and in some instances the entire volume of heap sand was not molded, although all sand was re-mulled prior to subsequent usage. The green strength has increased from 5 to 8 psi to about 12 psi at 3 per cent moisture. Dry strength has decreased to the point of easy shakeout. One addition of olivine aggregate bonded with 3 per cent bentonite was made in order to maintain the original volume of heap sand.

Usage Effects: Property changes previously noted can be ascribed to the combination of bond burn-out and change in grain size. Prior to the aggregate addition, a screen analysis gave the results shown in

Workability remained about the same even though green strength increased. Less ramming was required to produce a smooth, dense surface than with the original sand.

TABLE 2-IRON HEAP SAND SCREEN ANALYSIS

	Retained on U. S. Screen, %									
	20	30	40	50	70	100	140	200	270	pan
Before										
De-dust After	0.2	12.7	25.5	17.3	11.3	9.2	7.6	6.8	2.9	6.5
De-dust	0.4	23.2	36.4	18.1	8.5	4.6	3.2	2.0	0.9	2.7
	1		Iron H	leap B	reakd	own				
	0.6	9.3	23.6	16.9	10.9	7.3	6.1	4.7	2.0	7.1

Mulling Effects: The information gained from observing usage results indicated the desirability for a study of mulling effects. A random 10-lb sample of olivine was selected for processing in a laboratory muller. The mulled sand was tested at timed intervals during both dry and wet mulling. In the latter instance 3 per cent moisture was used. The extent of mulling breakdown is given in Tables 3 and 4. Test results indicate that dry mulling causes rapid grain breakdown, whereas wet mulling results in slower pulverizing action. The knowledge gained from the mulling tests has proved of profound significance for further olivine study.

Grain Size Control: Behavior of the olivine sand

TABLE 3-DRY MULLING BREAKDOWN TEST SCREEN ANALYSIS

Time,			S. S	creer	1, %					
min	20	30	40	50	70	100	140	200	270	pan
0	12.8	27.4	27.4	17.6	9.6	4.6	0.2	0.1	0.1	0.2
1/2	11.5	25.3	27.5	18.9	10.4	4.6	1.0	0.1	0.1	0.6
1/2	6.3	17.6	25.2	20.9	14.3	8.5	3.1	1.5	0.5	2.1
12	2.9	13.1	23.5	20.1	13.6	9.1	5.5	3.5	1.5	7.2

Table 4-Wet Mulling Breakdown Test Screen Analysis

Time,			Ret	ained	on U.	S. S	creer	1, %		
min	20	30	40	50	70	100	140	200	270	pan
0	12.8	27.4	27.4	17.6	9.6	4.6	0.2	0.1	0.1	0.2
1/2	9.8	24.3	28.0	18.6	11.0	4.7	1.2	0.8	0.4	1.2
1/2	9.2	22.5	26.7	18.4	10.8	5.8	2.4	1.6	0.6	2.0
12	5.2	18.0	37.5	20.5	12.6	7.4	3.4	1.8	0.8	2.8

aggregate on continued usage indicated that a change in grain size was occurring. Apparently this change was slowed following a noticeable effect when the sand was first used. Grain size distribution was out of control. It appeared as though an insurmountable problem was at hand because of the necessity for separating the aggregate in order to obtain grain distributions suited to casting requirements. The situation was intensified because of the desirability for economically separating sand finer than 50 mesh.

A separator was built to meet conditions in the University foundry. It was designed around the principle that surface to volume ratios of particles increase as their size decreases. The pilot model now in use employs the gravity principle for sand movement. A dust collector equipped with an exhaust fan is the only moving element in the separator.

Sand to be separated is dropped from the feed bin to a spreader that distributes the sand over the slide. Drop action is important since it breaks up the sand stream as it strikes the slide. Further, the smaller the sand particle, the greater the air resistance, re-

TABLE 5-SAND SEPARATOR SCREEN ANALYSIS

U.S.	Retained on Screen, %							
Screen No.	(a) Pull	(b) Cut-off	(c) Oversize					
12	0	0	0					
20	0	T	5.4					
30	T	0.6	27.7					
40	T 0.8	5.6	41.0					
50	T	21.2	20.5					
70	3.4	34.2	4.8					
100	6.3	28.9	0.6					
140	21.0	8.1	T					
200	21.4	0.6	T					
270	8.6	T	T					
Pan	32.2	0.8	T					

sulting in a velocity differential as the sand starts down the slide. This differential increases as the sand continues down the slide, and as it emerges velocity is a function of particle size. A cut-off plate is located at the end and below the slide. A vacuum is applied whereby its air stream moves normal to the sand flow as a means of carrying fines to the expansion chamber. The latter has a double compartment which serves to separate fines. The extremely fine material is carried into the vacuum collection bag. It is possible to separate sand into four size gradations with the present equipment. However, greater or lesser separation can be accomplished through adjusting and adding cut-off plates. An example of olivine sand separation is given in Table 5. This separation was made on new dry olivine sand free of bond.

The resulting separations give an ideal sand grain distribution in the selected ranges. The selection of the sizing terms used in the table arise from operating features of the separator equipment. Size and grain distribution obtained in the pull category have been placed on the non-ferrous floor with gratifying results. The sand molds readily and casting quality is excellent. Casting surfaces of degree of excellence that can be measured in micro-inches are now being produced regularly in aluminum alloys by the students.

The satisfactory results obtained with the separator equipment have led to processing all heap sand. The fines have been worked into the non-ferrous heaps and a classification for gray iron has been established. With this new classification of grain size and distribution of the iron sand in use, superior results are now at hand. These results strongly indicate that pr vious difficulties encountered with rough casting surfaces and tough adherent scale resulted from an incorrect grain size and distribution. Olivine molding sand aggregates, because of their grain geometry, are apparently more sensitive in the choice of size and distribution than is the case with spherical grains. Special attention in this matter will be conducive t splendid casting results.

#### Correlary Applications for Olivine

C-Molds: The results of the investigations indicate that olivine grains can be made available in an extremely wide range of sizes. The smaller sizes have proved to be splendid base sand for shell or C-process molds. This is a development of significance since olivine in fine grain sizes will cause little concern from a hygienic standpoint. Quality of casting surfaces from shell molds produced in this foundry is competitive to surfaces obtained by some of the precision casting methods. There are advantages of shell over precision castings in some categories. The relative cost of making a casting by these processes is a matter for management to decide in relation to individual foundry operations.

Cores: Olivine as a sand for baked cores has not been examined closely in the research program. Olivine cores have been used exclusively since it is desirable to eliminate all possible sources of heap contamination. The cores have a high green strength, a condition that is traceable to grain shape, size, and distribution. It was necessary to increase the amount of binder over usual practice in order to develop satisfactory dry strength. No examination has been made to find a remedy. It may be necessary to approach olivine dry sand core production from a fresh viewpoint since available knowledge applies primarily to silica-based-core mixtures. The cores used have performed satisfactorily in both gray iron and non-ferrous operations. Some of these were placed in the "as-baked" conditions. In other instances a core wash of graphite or olivine base was employed. No study has been made of fusion problems although some difficulties of this nature have been reported in steel castings in a Norwegian foundry. There is no information at hand to indicate whether casting design or olivine core processing caused this difficulty.

Related Applications: With all silica sand removed from the foundry, olivine was used in ladle linings. This change-over has worked out satisfactorily, although there is, as yet, no experience with steel operation. Cupola spout lining has also been made with olivine aggregate.

#### Summary

The research reported in this paper has all been of an experimental nature. There is, as yet, no commercial olivine producer in this area. Olivine was delivered in ¼-mesh size and required further crushing at the University. The resultant aggregate constituted a wide grain size and distribution. A commercial source capable of supplying olivine in grain size and distribution suited to a specific foundry requirement would be important to olivine adoption.

A vast amount of testing has been done by A. M. Nelson and reported in his graduate thesis, "Olivine for Foundry Sand." In the research program more than 600 mechanical tests have been made on 66 different olivine sand aggregate mixtures. Some of the results of the experimental work have previously been published by Schaller and Snyder¹¹ and by Schaller and Nelson. 10 A further investigation, "Experiments with Surface Tension Bonds," by Prof. W. A. Snyder will be published in an early issue of AMERICAN FOUNDRYMAN. The knowledge gained during this experimental program indicates the potential of olivine sand aggregate as satisfactory for gray iron and nonferrous operation. While some steel castings have been poured, there is insufficient information to warrant any definite conclusions.

#### Acknowledgments

The author gratefully acknowledges the great assistance of his colleague, Prof. W. A. Snyder, and his co-worker, A. M. Nelson, research fellow, both of the Mechanical Engineering Department, University of Washington.

#### References

- 1. Bjorn Langum, Foundry Engineer, Drammen, Norway, 1951.
- Olivine—Information Circular, I. C. 7239, U. S. Bureau of Mines, March 1943.
- Correspondence with A/S Strommers Varksted, Oslo, Norway, 1950.
  - 4. Illinois Institute of Technology.
  - 5. Western Metals, vol. 9, no. 12, Dec. 1951.
- "Industrial Hygiene in the Foundry," A.F.S. Transactions, vol. 57, pp. 567-575 (1949).
- 7. Geo. Allen, Canadian Metals Met. Industry, 1949.
- 8. K. J. Stenvik, Journal, Iron & Steel Institute, 1949. 9. King et al., Journal of Pathology and Bacteriology, 1945.
- 10. G. S. Schaller and A. M. Nelson, *The Foundry*, Oct. 1951. 11. "Olivine Sand Research," *The Foundry*, May 1951; and "Olivine—A Foundry Sand Aggregate," *The Trend*, July 1951.

#### DISCUSSION

Chairman: C. C. Sigerfoos, Michigan State College, East Lansing, Mich.

Co-Chairman: V. M. Rowell, Archer-Daniels-Midland Co., Cleveland.

Recorder: F. P. GOETTMAN, Standard Sand Co., Grand Haven, Mich.

Co-CHAIRMAN ROWELL: We did some laboratory research on olivine (GFN 41) for cores and found it to result in weaker

Mr. Schaller: It is quite possible that more encouraging results would be obtained with a different grain size. We have found a great improvement in casting finish by changing grain size.

## VENTILATION AT NON-FERROUS MELTING AND POURING OPERATIONS

By

Herbert J. Weber*

Some reasons why ventilation of furnace and pouring opera-tions is essential are presented. These are based on experiences with litigation cases, with occupational disease such as lead poisoning, and with labor relations.

The incidence of lead poisoning founded on experimental studies of the solubility of lead oxide in body fluids is briefly treated in its relationship to melting and pouring operations as compared to others.

Atmospheric concentrations of lead characteristic of these operations are reported on the basis of actual analyses of numerous air samples collected at the breathing zone of foundry workers.

Costs, practical applications, and deficiencies of typical exhaust ventilation for melting and pouring operations are described and illustrated by photographs of equipment in actual operation.

The results obtained so far indicate that a favorable solution

of the problem has been found.

The examples given herewith of methods of exhausting typical non-ferrous pouring and melting operations have been found effective and practical. They are relatively cheap measured in terms of the factors pointed out previously in this paper.

In treating a topic of this type, if the author were a foundry manager, he would first want to know why it was necessary to ventilate melting and pouring operations, whether the plant can afford ventilation, and what proposed methods were practical for his particular operation.

There are several reasons why these operations should be exhausted. First of all, there is the selfish one of protecting oneself against financial losses due to awards made in cases of compensation. A single \$5,000 award plus the cost of litigation can pay for a lot of ventilation. During the many times the author has testified in such cases, he has found that the chances of winning are far greater if a process is ventilated. If one has no ventilation or exhaust hoods, it is very difficult to prove that the claimant was not severely exposed to toxic fumes even though, for example, his blood and urinary lead level were within normal limits.

#### **A Typical Case**

Not long ago a certain man died of uremic poisoning. When he first became ill his family physcian was called in and he diagnosed the case as lead intoxication because the man worked in a brass foundry. He was taken to a local hospital and the physician's diagnosis was entered in the record. Later the patient became comatose because of his kidney condition and he was transferred to the veterans' hospital where he died.

The death certificate stated that the primary cause of death was uremic poisoning and the contributory cause lead poisoning.

The company lost this case before the referee and an award was granted for \$5,000.00. On appeal the decision was reversed because it was possible to demonstrate that atmospheric exposure to lead was safe as proved by air samples and by blood and urine analyses for lead over a period of many years.

Then too, there is another cogent reason for ventilating. State agencies are requiring it in their health and safety codes and one will be forced to exhaust melting and pouring operations anyway, perhaps at a time when the budget would be very strained to do so.

It is far better for industry to take the initiative in this matter and to do the job gradually as funds permit rather than to procrastinate until intolerable laws put the smaller operator, at least, out of business.

If this seems to be exaggerated, the author merely states in passing that smog control ordinances are becoming so strict that he doubts if the small operator can stay in business. To get a permit to build a foundry in Los Angeles County, you would need to install a collector on the cupola. One guaranteed to meet the Smog Control Ordinance will cost between \$60,000 and \$80,000, and the anticipated annual maintenance cost for filter replacement is \$12,000 for a single cupola

It may be of further interest to mention that claims for occupational loss of hearing are springing up all around us much like the old silicosis cases, and the average claim paid in New York State amounts to \$4,000 per case.

Thus we can look for more stringent health and safety codes, not a relaxation of the present ones, and it behooves the industry to take care of its own problems of ventilation in a practical manner now so as to avert passage of intolerable codes later.

There is also the intangible reason for installing ventilation-the personnel one. It is hard to measure this in dollars but it is becoming increasingly difficult to attract good men to the foundry.

After we installed exhuast hoods and ventilating systems in our foundries, we received many expres-

^{*} Chief Industrial Hygienist, American Brake Shoe Co., Chicago.



Fig. 1... Crucible is shown here without any cover.

sions of appreciation from our men and some of them told the author it was the best thing the company ever did for them. A satisfied employee will usually give a dollar's worth of work for a dollar. The very fact that one installs an exhaust hood is proof to him that management has his interest and well being at heart.

Finally there is the most important reason for installing exhaust ventilation—the moral one. When a man comes to us free of occupational disease, it is our moral responsibility to him to see to it that when he leaves our employment he does so just as symptom-free as when he came to us.

It is economically unsound to lose a skilled man or to have him off sick because of occupational disease even though he does not file a claim; and it is unjust to him since he and his family suffer an economic loss and physical pain. We are in fact punishing a good loyal employee by not protecting his health.

If we had the money, we could replace every furnace, cupola, or machine we have; but we cannot so readily replace skilled talent because it is not as available as

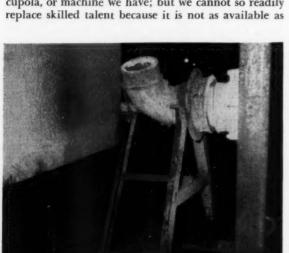


Fig. 3... Fume-diverting ell is bolted to tripod.



Fig. 2... Crucible with cover. Note absence of fume.

machines. The industry has a tremendous investment in its people. It takes years to make a first class molder or melter and we are foolish indeed if we do not protect his health and keep him on the job.

These men cannot be kept free of occupational disease unless melting and pouring operations are ventilated, for the simple reason that the fumes generated by these processes are toxic.

Most non-ferrous work involves the melting of copper, tin, manganese, zinc, aluminum and lead. There are certainly others such as silver, cadmium, and so forth. In the author's experience, lead is the principal and most toxic one. It is capable of producing both the acute and chronic forms of lead intoxication; it is cumulative, especially where there is pre-existing kidney impairment; and it can cause permanent damage to the body.

In our experience, concentrations of lead below 1.5 milligrams per 10 cubic meters of air are safe for all persons during a lifetime of exposure. Atmospheric concentrations up to 3.0 milligrams per 10 cubic meters of air are safe for all but susceptible persons.

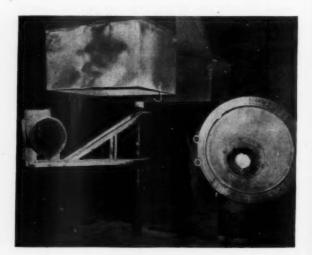


Fig. 4... Same ell hung on pivot in pour-off position.



Fig. 5... Ell here is shown in the melting position.

Concentrations much in excess of 3.0 milligrams may produce episodes of lead poisoning in any person. When one considers that a milligram is equal to 1/2500th of an ounce, it can be understood that we are dealing here with minute amounts of lead.

Melting and pouring operations are far more hazardous than machining or cleaning jobs for two reasons. First the particle size of lead fume is one-half micron and less. Secondly, the fume is evolved as lead oxide which is very soluble in body fluids. Thus the particle is so small that it is not strained out by the filtering action of the respiratory tract, and the size and solubility product are conducive to rapid absorption into the body.

The particle size in machining and cleaning operations ranges from ½ to about 100 and averages about 12 microns. Further, the lead is present as elemental lead which is not as soluble as lead oxide.

In a study which we made over a five-year period some time ago we found some interesting facts. The men in the machine shop were exposed to an average atmospheric concentration of lead dust of 4.90 milli-

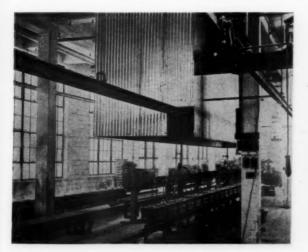


Fig. 7... Hood over pouring area used with monorail.

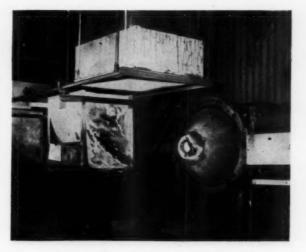


Fig. 6... Fume baffle suspended on overhead rail.

grams per 10 cubic meters of air, and the men in the furnace bay were exposed during the same period to 4.95 milligrams. Here, intensity and duration of exposure were practically identical. The differences in chemical composition and particle size were as previously mentioned. However, the percentage of men in the furnace bay who suffered episodes of lead intoxication was 57 while in the machine shop only 2.8 per cent of the men had episodes.

These facts indicate that a program of hygienic control should begin with the ventilation of melting and pouring operations, and answer the question why it is necessary to control the fume here.

Can the plant afford to ventilate? The author does not think it can afford not to, if management's thinking is long range and if it intends to stay in business. If the program of control is done gradually by starting with the worst offender, the cost is not too burdensome. Then there are minor things which the plant can do with little or no cost.

If we consider that the ventilating of melting and pouring operations is an air problem and not a fume



Fig. 8... Note position of hood as ladle grows empty.



Fig. 9... Hood over cupola spout. There are no fumes.

problem we have the clue to some simple remedies. How does lead fume get from the crucible, for example, to a point some 20 ft away? Obviously some kinetic energy is imparted to it by the vapor pressure above the molten metal, but this would not transport it very far in a vacuum. However, air currents caused by rising heated air and cross drafts act as convectors for the fume. Now if these convectors can be stopped or controlled we can prevent dispersal of fume.

In transporting a crucible of molten metal from one end of a bay to the other the man at the tail end of the bale is in constant exposure to high concentrations of lead fume. This is true also of a lift truck operator who transports crucibles to pouring stations.

By simply covering the crucible with a garbage can type of cover, we reduced the concentration of lead from 44.6 mg per 10 cubic meters of air to 2.5 mg. These covers can be made by the maintenance man.

Figure 1 shows a crucible without the cover and Fig. 2 shows one with the cover secured in position. By placing a steel lid over the pouring spout of a



Fig. 11... Hood over a furnace in melting position.



Fig. 10 ... Effect when exhaust fan is turned off.

reverberatory furnace during melting, contamination at the breathing level from this source can be almost entirely eliminated. The cost of this protective lid is negligible.

If bottom pouring is practical in your operation, a crucible or ladle can be kept covered and the amount of fume given off is only a fraction of that produced by the conventional method.

Mention was made previously of controlling air currents. Barrel furnaces discharge fume directly into the breathing zone. To baffle this fume above the breathing zone where it could be removed by overhead ventilators we formerly used a diverting ell as shown in Fig. 3. The ell was bolted to a tripod on casters and moved into position during melting. It was so cumbersome, however, that frequently the men failed to use it. We then installed it on a pivot which made it possible to swing the ell in and out of position very easily. This is shown in Figs. 4 and 5. This type of baffle clogs with slag and unless kept clear, it does little good. As a result we designed the baffle shown in Fig. 6 which can be hung on an overhead

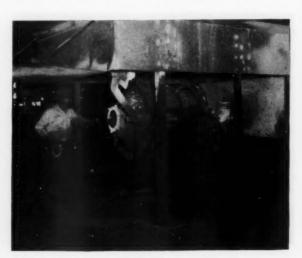


Fig. 12... Hood over furnace with an ungapped stack.

monorail as shown or on a pivot. The total cost of one gravity hood and baffle shown in the figure was \$1,000.

In production foundries where a monorail can be used a simple hood (Fig. 7) over the pouring area is effective. But in jobbing work or where overhead cranes are used, hoods are not always practical. For this type of operation we have developed a simple hood fitted to the bale and connected by flexible tubing to an overhead exhaust fan which is attached to the crane bridge. This installation blows the fume to overhead ventilators which discharge it to the outside; it reduced the exposure of the pouring crew 65 per cent; and it keeps contamination from reaching the molding areas. The total cost was \$300. Figure 8 shows the degree of fume reduction produced by a ladle exhausted in this way.

#### **Cupola Can Also Contaminate**

Another source of great contamination is the cupola, especially during tapping. We have found concentrations of atmospheric lead well in excess of 20 milligrams per 10 cubic meters of air in the tapping areas of cupolas. This is a very severe exposure. By means of a simple hood over the spout shown in Fig. 9, we were able to reduce the atmospheric lead to a concentration in the region of 1.4 milligrams per 10 cubic meters of air-a safe exposure. The cost of this type of hood was \$1200. One case of lead poisoning could easily cost as much. The hood is effective, inexpensive, and does not interfere with operation. Furthermore it does not remove a large volume of warm air, which becomes a factor with all exhaust equipment moving large volumes of air especially in the winter time. Figure 10 shows the effect produced when the exhaust fan is turned off.

General roof ventilation in the long run is the most expensive and least effective kind. It generally removes fume after it has been breathed by the workmen. It does serve a good purpose however in that it keeps the general atmosphere diluted, but it does not guarantee that worker A or B is in a safe environmnt. It is more applicable where the contaminant is principally smoke or some other physiologically inert substance. In non-

Fig. 13... Note cover on pouring spout, lack of fume.

ferrous foundries where the fume is apt to be toxic, it should be used as an adjunct to local exhaust ventilation, not as a substitute.

Some codes specify that a foundry should have 20 air changes per hour. This is misleading for non-ferrous melting and pouring bays since the air-change cannot easily be distributed in a uniform manner.

In one of our melting bays, we had eighteen 18,000-cfm ventilators in the roof. This is equal to exhausting 19,440,000 cu ft of air per hour and is equivalent to 56 air changes per hour. Yet the average concentration of lead was in the order of 12 milligrams per 10 cubic meters of air. By means of the hoods with gravity stacks shown in Figs. 11, 12 and 13 we reduced the concentration to 2.5 milligrams of lead per 10 cubic meters of air. The costs of these hoods were \$1,600, \$1,000 and \$900 respectively.

One problem we have had with power ventilators is in trying to convince workmen that if they open the windows in the monitor they merely short circuit the exhaust so that little contaminated air is moved from the breathing zone. In one test we collected air samples at various work stations with all the monitor windows closed. We then collected air samples at the identical stations with all of the monitor windows open. In both cases, the power ventilators in the roof were operating. We found a reduction of 31 per cent in the atmospheric contamination at the breathing zone when the windows were closed. This is a very important fact to bear in mind if the furnace and pouring bays are exhausted by power ventilators.

#### **Use of Floor Gratings**

A very effective way of exhausting a furnace bay is shown in Figs. 14, 15 and 16. For this particular bay, fume from the furnaces is directed upward by means of baffles into the gravity hoods (Fig. 14). Fresh outside air is introduced through the floor gratings between each furnace (Fig. 15) at the rate of 60,000 cfm, and overhead ventilators exhausting at the rate of 100,400 cfm keep the bay under vacuum and remove any traces of fume escaping from hoods, ladles, etc. In addition, the charging floor is completely walled off to protect the charging crew, and furnaces are charged by



Fig. 14... Gravity stack and baffles in furnace bay.



Fig. 15... The floor grating admits fresh air here.

means of chutes (Fig. 16) projecting in the furnace doors proper.

The concentration of lead in air in the bay itself ranges from 0.3 to 1.9 and averages 1.0 milligrams per 10 cubic meters. On the charging floor, atmospheric lead ranges from 0.4 to 1.1 and averages 0.7 milligrams per 10 cubic meters of air. These are safe values, and we have no cases of lead intoxication in this very hazardous department.

Referring to a previous statement that fume control is really a problem in air control, we followed that principle in designing this particular bay.

1. We prevent fume from being carried to the charging floor by walling it off from the bay. This stops the convecting air currents from reaching that area.

2. We control the hot convecting gases issuing from the furnaces by baffling them upward above the breathing level into gravity stacks.

3. We introduce fresh outside air to serve as a convector and diluent for those fumes escaping the hoods or rising from ladles during skimming and pour off.

4. We protect areas outside of the bay by keeping it under vacuum so that the relatively clean air from the molding areas travels toward the bay.

The writer does not believe a furnace bay can be effectively exhausted by overhead ventilators alone unless the furnaces are raised to a height by hydraulic elevators so that the pouring spouts during melting will be about 10 ft above the floor in a low bay—say 25 ft at the peak.

Even the reverberatory type of furnace and the kettle type (examples of which are shown in Fig. 17 and 18) which discharge fume well above the breathing level, in the experience of our company, require local exhaust hoods.

Pit furnaces or crucible furnaces, as they are often called, are a source of severe exposure to toxic fumes in the melting department. Somtimes it is impractical to use monorail cranes for removing the crucibles from the furnaces so that the more flexible conventional overhead crane may be required which may preclude the use of local exhaust hoods. We were confronted



Fig. 16... Charging floor of bay shown in Fig. 14.

with this problem and solved it in the following way (see Fig. 19 for details).

The floor directly in front of the furnaces was removed and an air tunnel installed. The tunnel was covered with a steel grating as shown in the figure. Baffle plates were suspended at experimental intervals below the grating to get an even distribution of air. Cost of the air tunnel was \$3,000.

A fan capable of delivering 25,000 cfm was installed at the far end of the tunnel. This fan brought in fresh outside air and delivered it uniformly through the interstices of the grating. This formed in effect an air curtain at the breathing zone of the workmen, and effectively diluted the fumes to safe concentrations. Overhead ventilators removed the contaminant from the building.

The observed effects were:

1. The fume was blown away from the men and upward. This can be seen to a small degree in the figure by observing the slight deflection of the flames toward the wall.

2. The cooling effect of the cold air on the men made the atmosphere more comfortable.

3. The installation introduced no heating problem, since the overhead ventilators did not remove intentionally heated air but cold outside air.

4. The system offered absolutely no interference with operations, overhead cranes, or production.

Air samples were collected along the floor grating at the breathing level of the workmen, with the air curtain turned off and with it in operation. The samples were collected at the foreground end, the center, and the background end of the grating. The results are as follows:

Location of Sample		re Conc. of g/10 cm,	Percentage of Reduction in
Station	Air off	Air on	Concentration
Front	5.5	1.7	69.0
Center	7.5	0.7	90.0
Back	5.1	0.6	88.0
Average	6.0	1.0	83.0



Fig. 17 (above)...Reverberatory furnace with a gravity stack. Notice flame discharging into the stack.

Fig. 18 (above right)...Gravity stack for kettle-type furnace. Local exhaust hoods are generally needed even though the fumes are discharged above the breathing level.

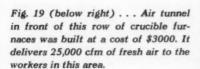


Fig. 20 (below) . . . Hood over bank of pit turnaces. Note effective removal of all tumes here.







# AIR FURNACE REFRACTORY PRACTICE IN CUPOLA-AIR FURNACE DUPLEXING

By

William C. Corbeau*

#### ABSTRACT

In this paper the author presents air furnace dimensions, the manner in which bottom and side walls are laid, refractories used, data on refractory consumption, repair schedule, provisions for metal removal should the furnace bottom come up, a discussion of various repair methods, changes in practice over the years, and recommendations for design changes in new furnace construction.

The Cleveland Works of the company with which the author is associated employs the cupola-air furnace duplex melting unit in the production of malleable cast iron. The melting equipment consists of two cupolas and two air furnaces. The daily melt is between 500 and 600 tons for two 8-hr shifts in the foundry.

The two 120-in. cupolas are lined to 82 in. at the tuyeres and are operated on alternate days. The two air furnaces are alternated on a 3-day cycle, i.e., one furnace is operated on three consecutive foundry operating days; the other furnace follows a like period of operation. The furnace repair is done in the 3-day period the furnace is down.

The air furnaces and a central stack are aligned upon a common centerline. Two dampers, one at each stack opening, are open or closed depending upon the furnace in use. A suction fan connected to a waste heat boiler draws the gases through a breeching attached to the side of the stack. If it is desired, the furnace may be operated on natural stack draft by raising a damper on the top of the stack. In either case the furnace operator controls the furnace draft which is held to zero at the skim door.

The cupolas are in a line parallel to the air furnaces and equidistant from the center of the air furnace stack. An "X" runner layout provides for the flow of metal from either cupola to either furnace.

The temperature of the cupola metal is 2700 F (1482 C) to 2740 F (1504 C), and the furnace metal at the tap spout is 2800 F (1538 C) to 2830 F (1554 C).

The normal metal content of the air furnace is between 30 and 35 tons. The capacity is 40 tons.

Pea sized pocohontas coal, prepared in a unit pulverizer, is used for fuel. A single water-cooled coal burner is set with its centerline at 5 degrees to the horizontal and at such an elevation that a projection of its centerline would strike the surface of the bath opposite the tap holes. Two low-pressure oil burners, mounted one on each side of the coal burner, are stand-by equipment.

The air furnaces were built in 1930. Since then they have been raised 3 ft 10 in. They are constructed solidly from the ground up as opposed to the now common air-cooled bottoms.

The bath measures 7 ft 6 in. in width by 42 ft in length. From the burner wall to the tap holes is 12 ft. The furnace bottom, on the centerline, from a point opposite the tap holes, rises 3 in. to the burner wall and 3 in. to the front of the bridge wall, and drops 2 in. to the bottom of the lower tap hole. The transverse section of the bottom is an inverted arch having a depth of 4 in. The skim door sill is 10 in. and the bridge wall is 13 in. above the low point on the centerline.

Three tap blocks are set in triangular pattern with the center of the two outer blocks 2.5 in. above the center of the lower block. The metal depth at the lower tap hole is 11 in. (maximum, 12 in.).

#### Method of Bonding Joints

The furnace bottom consists of a series of rings (arcs) of superduty clay brick laid on end over a 4.5 in. sub-bottom. Each ring is carried to the  $4\frac{1}{2}$  in. jacket course, at which point it is wedged. All joints are dipped in air setting, high temperature bonding mortar supplied by the brick manufacturer. All bricks used in the furnace are of the standard 9-in. sizes in the  $2\frac{1}{2}$ -in. series. In the bottom the majority of the bricks are 9 in. straights and No. 1 wedges in the ratio of 6 to 1.

The side wall inside the jacket course rests on the ends of the bottom. The first four courses of side wall are of superduty brick. The first three courses are of bonded construction to form a 13.5 in. thick wall. The fourth course, formed by a soap and a header, is 11.25 in. thick. The balance of the wall is

^{*} Melting Superintendent, Cleveland Works, National Malleable & Steel Castings Co., Cleveland.

9 in. thick, and composed of high duty Kentucky clay brick. The top line of the side wall is a straight line having a fall of 24 in. from the burner wall to the stack. The distance between the bottom, on a centerline opposite the tap holes, and the top of the side wall is 45 in.

The roof is constructed of a series of 42 bungs having a rise of 12 in. The bungs are lined with 13½ in. No. 101 and No. 105 standard bung brick made of Missouri clay. A mix consisting of ganister, fireclay and goulac, to which sufficient water has been added to make it workable, is prepared in a wet pan mill and used for a seal where the bung rests on the side wall, for the joints between the bungs, and for lining the furnace runners and spouts.

The bottom repair is made in three sections: (1) burner wall to bung 15; (2) bung 16 to bung 25; and (3) bung 26 to bridge wall.

Usually 12 days' production, or four periods of three foundry days each, are run over each section of the bottom before it is replaced. Only one section is repaired in a repair period. In one of the repair periods no bottom is replaced. In each repair period tap blocks and the section of wall over the tap blocks are replaced. In our practice full rings of bottom brick are replaced, never a part of a ring.

It is the author's opinion that too much care and attention cannot be given to the furnace repair. There must be good workmanship and close supervision. For that reason the furnace repair is assigned to the mason foreman. His group, which includes furnace masons and mason laborers, removes the old brick work, repairs the furnace bottom, walls, neck, arch and stack, and relines the bungs. Placing and sealing the bungs, repairing the runners and spouts, and the weekly removal of the slag and ash from the stack is done by the furnace men.

Operators of duplex air furnaces know how disheartening it is to see the bottom floating on the metal. In anticipation of such a condition and as an

Below—(left) View of Cupola and air furnace. Air furnace stack is shown in center background; (right) Trough constructed in front of furnace is used to drain metal in emergencies.

#### AIR FURNACE REPAIR DATA

Bottom Brick (includes lower 4 courses of side wall)				
Melt,	530,470	tons	1.31	bricks/ton*
Melt,	90,118	tons	1.296	bricks/ton*
k (incl	udes bu	rner v	wall and	l neck)
Melt,	90,118	tons	0.863	bricks/ton*
Melt,	90,118	tons	0.250	bricks/ton
				-
			0.154	hr/ton-Mason
Melt,	90,118	tons		hr/ton-Melt.
•			0.168	hr/ton-Total
	Melt, Melt, k (incl Melt, Melt,	Melt, 530,476 Melt, 90,118 k (includes bu Melt, 90,118 Melt, 90,118	Melt, 530,476 tons  Melt, 90,118 tons  k (includes burner v  Melt, 90,118 tons  Melt, 90,118 tons	Melt, 530,476 tons       1.31         Melt, 90,118 tons       1.296         k (includes burner wall and Melt, 90,118 tons       0.863         Melt, 90,118 tons       0.250         Melt, 90,118 tons       0.014

 Each split, soap, wedge, straight, etc., used is considered as one brick in the computations.

aid in expediting the repair work after it occurs, we have located drain holes on one side of the furnace at the level of the sub-bottom at about 5 ft. intervals.

Should the bottom come up, the drain hole is burned through opposite the pool of metal and the metal is drained into a trough constructed parallel to the two furnaces. This trough also serves to take the metal should there be a breakout through the bridge wall into the stack. Thus, the delay encountered in removing the salamander from the furnace is avoided. The metal in the trough can be reduced to readily removable small chunks by slicing it with a bar just before it solidifies.

Many furnace operators wedge the furnace bottom between the side walls with good results. We believe it to be better construction to carry the bottom under the side wall and wedge it to the jacket course. Thus, with a good tight job of brick work in the arch, with the ends wedged to the jacket course and the weight of the side wall holding it down, there is little chance of the bottom rising.

Some believe that economy lies in using the large 9-in. brick in the bottom because there are fewer bricks to lay, fewer joints and less labor required. The lower unit cost of the regular 9 in. straight and its greater uniformity have kept it in our favor. Actually, no labor saving was found with large 9-in. brick.

Many lay the brick dry in the bottom and, when





finished, brush fine silica sand or clay wash into the joints. The author prefers the dipped brick joint because the brick stays where set and a bond along the entire joint is made. The finished surface is also

brushed with high-temperature cement.

In this plant the superduty brick have been used in the air-furnace bottoms since 1946. From 1931 to 1946 the 60 per cent high-alumina brick was used. Because of some bottom failures with this brick and the reported success of one of our plants in using the lower cost superduty brick (in air-cooled bottoms), it was thought advisable to try them. The results were so satisfactory that the practice was continued. In one furnace the superduty brick from one manufacturer are used, and in the other furnace the superduty brick from another manufacturer are used with equally good results.

The bung brick are from one source and the side-

wall brick from another source.

#### Revise Construction Details

In 1930 when the plant ventured into cupola-air furnace duplexing there was little information to be had from the experience of others. It was natural to continue much of the batch type, hand-fired air furnace practice such as wind bung for top air, cameltype back roof, rammed tap holes and sand bottom.

In the beginning all of the above were included with the exception of the sand bottom. Later it was learned that the wind bung could be eliminated with a saving on maintenance. The additional top air through the wind bung created a turbulence which was hard on the side walls and bungs. The cameltype roof caused the bungs to burn unevenly. The straight-line roof as adopted proved to be much more economical both in increased bung life and in labor

setting the bungs.

The change from rammed tap holes to the use of tap blocks was by far the most important factor. Without the tap blocks the daily long periods of high tonnage could not be maintained. Having had experience with brick bottoms in the electric furnaces, triplex melting, it was decided to use brick bottoms in the air furnaces. The plant was fortunate in having the guidance of a mason foreman who had had much experience. It is to him and the brick manufacturers who worked with us that credit is given for the success we have had in duplex air furnace refractory practice.

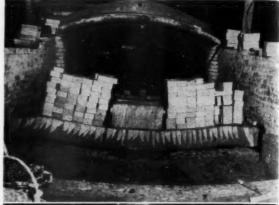
#### Conclusion

If new duplex air furnaces were to be installed, it is believed that most of the present practice would be continued, with the inclusion of the following changes in design:

(1) The air furnace bath would be widened between 12 in. and 15 in. for better heat efficiency;

(2) The air furnace would have an air-cooled or ventilated bottom for economy in bottom brick;

(3) The common stack would be divided at the base so that the slag and ash removal and repair work in the stack could be done at straight-time hours during the week rather than at premium time over the week end.







Top—Air furnace section through bottom looking toward burner wall before side walls are laid. Center—Section through bottom showing side-wall construction. Usually side wall is not laid until the bottom has been completed. Bottom—Detail of side wall and bottom construction.

#### DISCUSSION

Chairman: W. G. FERRELL, Auto Specialties Manufacturing Co., S.t Joseph, Mich.

Co-Chairman: ERIC WELANDER, John Deere Malleable Works, East Moline, Ill.

Recorder: WM. ZEUNIK, National Malleable & Steel Castings Co., Indianapolis.

H. E. LEICKLY: 1 Would you clarify your statement on draft at the skim door opening?

MR. CORBEAU: The pressure is zero. We use waste heat boilers so with the exhaust fan we can readily control the draft conditions in the furnace.

MR, LEICKLY: Does this improve your carbon control and also

improve the life of refractories?

MR. CORBEAU: We do not use draft to control carbon but it does improve refractory life.

CO-CHAIRMAN WELANDER: How do you install the rings or courses in your furnace bottom?

Mr. Corbeau: We work each way from the center and key on the shoulder against the jacket course. The center brick in each course forms our base line for the bottom slope.

MEMBER: Have you tried water-jacketed sidewalls?

MR. CORBEAU: No, we have not.

MR. LEICKLY: Do you provide for expansion in each course?

MR. CORBEAU: No, we do not.
R. A. WITSCHEY: Do you not use high-burned super-duty and not regular super-duty brick?
MR. CORBEAU: That is correct.

MR. LEICKLY: Is your cost figure of bungs expressed in terms of bungs or 9-in. equivalents?

MR. CORBEAU: It is expressed as bungs.

L. B. Winings: 3 Do you drain the furnace after each heat? MR. CORBEAU: Yes, we do.

MR. WININGS: What kind of brick are used in the sub-bottoms? MR. CORBEAU: First quality fire clay brick are used.

Co-Chairman Welander: Do you use any material between the sub-bottom and the service bottom?

Mr. Corbeau: Yes, 1/4 to 1/2 in. of ganister-fire clay mix as described in the paper. It makes it easier to remove the service bottom and protects the sub-bottom.

Fanner Manufacturing Co., Cleveland.
 A. P. Green Fire Brick Co., Chicago.
 Wagner Malleable Iron Co., Decatur, Ill.

## PATTERNS FOR HIGH PRODUCTION MALLEABLE AND STEEL FOUNDRIES

By

Joseph M. Kreiner*

#### ABSTRACT

This paper describes the design, construction and complete standardization of pattern and corebox equipment for modern, high production foundry use. The type of patterns described are constructed on interchangeable basis, since at least four foundries are being supplied with identical patterns. The castings made from the patterns must interchange in an assembly with castings made in eight different foundries.

#### Introductory

This paper describes in detail the design, construction and complete standardization of pattern and corebox equipment for modern, high production foundry use. The scope of the title is so large that one standard railroad article has been selected as the subject for discussion.

The device selected is identified as the standard "E" freight car coupler of the Association of American Railroads. Before launching into the main topic of the paper, i.e., the foundry equipment, a brief description of the device is in order. Also some historical data as to the adoption of the device as standard may be helpful in order that one may appreciate the necessity for the pattern standardization program that will be outlined.

The freight car coupler (Fig. 1) is an assembly of cast steel and malleable iron parts. Sometimes the smaller parts are made as forgings but this is an optional procedure. This assembly, or, at least, the most important part thereof may be seen projecting from both ends of all cars operated on railroad lines. The coupler connects the cars into continuous trains for haulage. It is locked automatically be being bumped into the adjacent coupler and must remain locked until released by the operator using an uncoupling rod attached to the end sill of the car body. That part of the coupler that extends into the car underframe is connected to a shock absorbing mechanism that decreases the impact forces caused by starting and stopping the train.

The standard freight car coupler was adopted by all the railroads in 1916. Previous to this time over 100 different styles of automatic couplers were in

use. The maintenance of repair parts for these numerous varieties of couplers operating on all the railroads in our vast rail system must have been an appalling problem. The first standard coupler adopted in 1916 was called the Type "D" coupler and it was supplanted by a new, heavier and improved Type E coupler in 1931. There is now a still more improved coupler undergoing service tests. The railroads and manufacturers believe it will go a long way toward reducing car derailments and wrecks. This latest design of freight car coupler is identified as the Type F coupler. During the span of life of any of these various standards many improvements are constantly being worked into the design by the manufacturers. Since interchangeability with all existing equipment must be maintained some desirable improvements cannot be made as soon as they are recommended but must be deferred until a completely new standard is to be adopted by the Railroad Association.

The standard coupler (Fig. 1) is manufactured by six foundries in the United States and two in Canada. Each of these manufacturers' supplies complete couplers as well as repair parts therefor. Wherever these

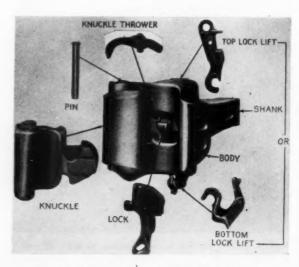


Fig. 1-Exploded view of coupler and parts.

Assistant Chief Engineer, Railway Division, National Malleable and Steel Castings Company, Cleveland.

various assemblies or parts are made they must meet all gaging requirements and interchange in any combination of parts as made by the several other foundries. In order to insure this interchangeability of parts, standard gages are used at each point of manufacture. A complete set of gages consists of 81 separate and individual gages. The largest manufacturer of the group is charged with the responsibility of furnishing one complete set of gages and checking masters for each foundry. Complete gage drawings are also available and distributed to every manufacturer. With this information all foundries can make additional gages and check them for accuracy or subsequent wear in service.

Another set of drawings is maintained showing in very elaborate detail the various castings involved in the device (Fig. 2). These drawings are also issued, through the coupler manufacturers' Technical Committee, to each foundry. All drawings are made to full scale and arranged on 18 x 24 sheets. The complete set of gage drawings number 63 separate sheets and the casting details number 50 sheets,

The yearly requirements of couplers for new car construction is large but of course varies to suit current car building programs. The maintenance of over one and one half million cars in service also requires a considerable quantity of complete couplers and repair parts. Maintenance requirements are closely related to existing business conditions as affecting freight shipments.

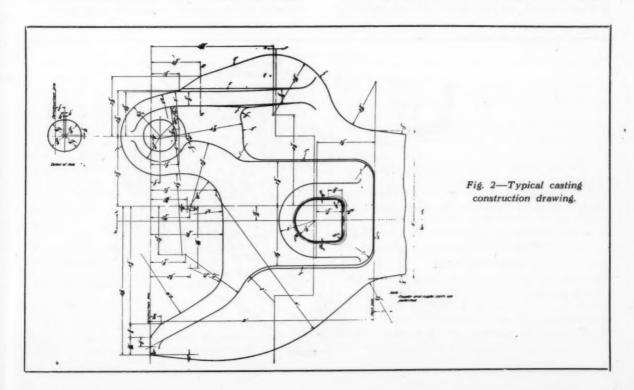
A complete coupler weighs 423 lb. The main part or body is a Grade B steel casting weighing 312 lb. The knuckle is of a medium manganese H. T. steel normalized and tempered, and weighs 76 lb. The lock of the same H. T. steel weighs 14 lb. The knuckle thrower made in Grade B cast steel or forged

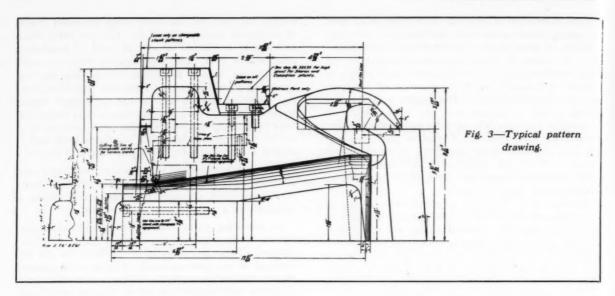
weighs 5½ lb. The top lifter plug, generally made of malleable iron, weighs 3 lb. The top lifter link, which may be forged or cast of steel or malleable iron, weighs 2 lb. The same coupler body can also be fitted for bottom operation by omitting the top operating parts and using a malleable iron lever casting weighing 3½ lb and a cast or forged toggle weighing 1 lb. The pivot pin which only keeps the knuckle from falling out when in open position, is a round rolled rod section with an upset head and a drilled cotter pin hole. This pin weight is 8½ lb.

This rather prolonged introduction should give one a better understanding of the rather involved technical procedures necessary to adopt and maintain a standard on a vast transportation system.

In preceding paragraphs the complete detail construction drawings of the castings parts were explained. These drawings form the basis for the beginning of our pattern work. The General Pattern Department of the author's company starts with these drawings and designs and builds suitable wood master pattern equipment, which, as soon as possible, is reproduced into metal master patterns. The expense of metal master patterns is justified by the fact that this master equipment must be available for constant reproduction for 15 to 20 years.

The General Pattern Department mentioned above, supplies metal working patterns and coreboxes to two of the company's foundries and also to the two Canadian foundries. Company policy has been established to insure that all patterns and coreboxes are completely interchangeable from one foundry to another. Economically the standardization is justified due to the fact that patterns and coreboxes can be made in quantity and stored at the General Pattern

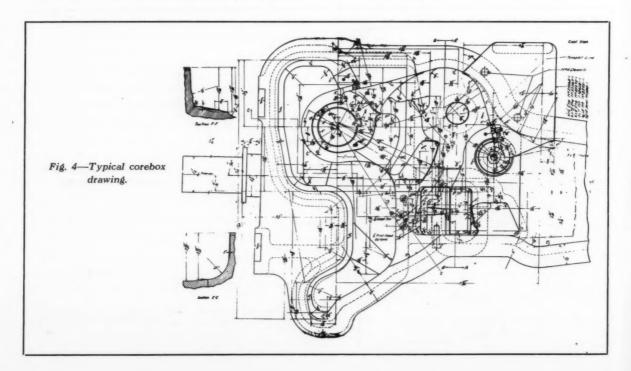




Department until requisitions are placed by any of the four foundries being supplied from this source. The best of materials and workmanship are insisted upon. Accuracy of machine operation is expected to be within  $\pm~0.002$  in.

The standardization of patterns and coreboxes is carried out by means of complete working drawings of each individual pattern and corebox. Loose pieces, pins, seats, etc., are completely dimensioned. The detail drawings (Fig. 3) of the patterns are dimensioned to include shrinkage allowance. Our several reasons for including the shrinkage allowance in the dimensioning are as follows: (a) Due to the complicated internal coring and close gaging requirements of as-cast surfaces, we have found through experience

that several shrinkage allowances are necessary. For example, ½ in. per ft shrinkage in length, for the coupler body; ½ in. in depth and ½ in. in width is also used. In addition grinding allowance is also required on the joint of the largest cores. (b) Standard machine tool measuring instruments and indicators may be used as in straight machine work. (c) Adjustments to patterns to take care of variations from standard shrinkage rule allowance are easily arranged by recording changes in these drawings. Any changes or adjustments of permanent nature, if made at the foundry, are immediately reported to Engineering Department and incorporated in the standard tracings. In this way all improvements are automatically included in new patterns in stock or being built.



# Steel Foundry Pattern Equipment

The various types of patterns required to produce this standard coupler and component parts will now be reviewed. The largest part, or body, is cast two in a mold and requires eight coreboxes. The patterns are cast in a good grade of brass and have a machine finish allowance of about 3/32 in. These patterns are completely machined to the pattern drawings. Layout and contour templets are used and a model pattern is also provided to be used whenever profiling or

duplicating is desirable.

Coreboxes are made for hand ramming or blower machine operation. The large body box is made of cast iron, steel faced on the top surface. Various inserts and loose pieces are made of brass or hardened steel (Fig. 4). This box is fitted with trunnions cast integral with the box and when in use, is mounted into a roll-over type of coremaking machine. The aluminum boxes designed for blower machines are also machined to size on all the inside surfaces and provided with armor on top, as well as sides and bottom. The blower holes and vents are applied at the foundries.

The knuckle patterns are also made of cast brass and machine finished. The unusually curved surfaces on this pattern make it ideally suited for duplicator machine finishing. The patterns are mounted either four per mold or eight per mold, as desired. After sampling and final approval we have found hard chromium plating to be a desirable procedure to increase pattern life. This pattern requires three coreboxes which are also designed for blower machine operation. They can be supplied in single cavity or multiple cavity design to suit various sizes of blowing machines. These cores are mainly lightening cores and the aluminum boxes are cast to size in most areas and the machining is restricted to fitting points and coreprints. Suitable aluminum coredriers are

provided.

The lock is a smaller steel casting and multiple patterns are required. As many as 14 patterns are mounted on one pattern plate. These parts require an irregular parting line so we have found the cast aluminum type of pattern with cope and drag plate incorporating the parting lines to be desirable. These patterns are produced by the plaster molding process and thus are accurately matched in pairs. The Pattern Machine Shop simply cleans up each pair of cope and drag castings and, by means of indicator lines, machines the edges of the plate portion accurately to the same external sizes. The bottom surfaces of the plates are also cut to produce even plate thicknesses. The foundry that receives a lot of these patterns has a comparatively simple job to mount and match these patterns on the molding pattern plates. The two coreboxes required for this casting are conventional aluminum coreboxes with multiple cavities. Here, however, the insides are completely machined since they form working holes or slots that are closely gaged in the castings.

The knuckle thrower, while sometimes forged by some suppliers, can also be cast of steel. This is also made by the General Pattern Department as a cope and drag plate pattern since it has a slightly irregular parting. This type of pattern is quite simple since no cores are required. The master pattern must be accurately made, however, because the gaging tolerances are close. This is perhaps the one reason why some manufacturers prefer to forge the part.

# Malleable Foundry Pattern Equipment

Smaller operating parts are made of malleable iron. Cast steel or malleable iron are optional materials allowed by the specifications but the small size and intricacy of the casting indicates that malleable iron is the preferred metal. The malleable production pattern equipment is not unusual in appearance or design. Aluminum squeezer plates for snap flask operation being used. The coreboxes are of aluminum in multiple cavity designs with as much armor as the convenient handling of the box will allow. Aluminum coredriers are used where necessary. Here again, however, the standardization program is carried to the ultimate by having completely detailed drawings of the pattern squeezer plates. The pattern is dimensioned and the spacing of patterns on the plate indicated on the drawing. The coreboxes are similarly drawn in detail to interchange with any plate pattern.

#### Conclusion

We have attempted to describe a pattern manufacturing system that we believe to be remarkable if not unique. The standard established practice of providing complete construction drawings, including shrinkage allowance, of patterns and coreboxes has not been observed in use elsewhere by the writer during 30 years with the Foundry Industry. This system has much to its credit in economy and accuracy, but, of course, can be justified only in a large volume, long extending casting design. The advantages are several and very important under these conditions:

1. Pattern equipment can be completely interchangeable.

2. Pattern equipment can be economically manufactured in quantity.

3. Adequate stocks of patterns can be maintained for quick expansion or replacement in case of accidents.

4. Large expansion can be made quickly since contract jobbing pattern shops can be utilized to full advantage

5. Orders for patterns can be spread out with several or many contract pattern shops to expedite deliveries without loss of interchangeability.

6. Standards of the customers who, in this case, include all the railroads in the United States, Canada and Mexico, are adhered to and maintained to the full extent of a most rigid set of specifications.

# Acknowledgments

The author wishes to acknowledge the encouragement of Mr. K. L. Selby, Chief Engineer, Railway Division, and members of his staff; Mr. W. L. Woody, Vice President of Operations. Also the kind assistance of Mr. William Ewing, Manager of Sharon Works, and members of his staff.

# APPLIED WASTE CONTROL PRINCIPLES

By

John R. Walley*

Waste control—the recognition and elimination of wasted time, effort, space, materials, and power—is a sound basis on which a foundry-wide cost reduction program can be initiated.

The principles of waste control are simple and few in number. Actually, there are only six steps involved in applying waste control:

(1) Spot a source of waste; (2) record the work done; (3) ask questions about the work; (4) write down your ideas; (5) get your ideas approved; (6) put the new method into effect.

The first principle of applying waste control is to "spot a source of waste." The beginner who applies waste-control principles for the first time encounters one big problem, and that is in determining the job he is to analyze. This problem is a stumbling block, not because there is a scarcity of wasted time, effort, space, materials or power, but because there are so many sources of waste within the average foundry that the beginner has trouble in narrowing his focus down to the one problem on which he will work.

The author suggests that the supervisor or timestudy man pick any small job at random, such as making a small core, cleaning a small casting or, if he is in an indirect department, analyze delivery of patterns, storage of core boxes or a similar operation. A safe rule to apply here is that any job being done in the foundry is a possible source of waste until it has been analyzed and proved otherwise. The main thing is to pick one small job. At this step of the analysis, consideration should not be given to whether or not the job has a source of waste in it.

#### Start on Small Scale

One pitfall encountered by many who hunt for a source of waste is that they want to start applying waste control principles on too big a scale. As an example, in many foundries, the beginner might want to install an overhead sand system, or otherwise solve some major problem by a tremendous capital expenditure. Improvements of that type have their place, but they are not a part of waste control as it can be applied by every management man in the shop.

Waste control is a procedure that is to be applied day in and day out to eliminate the sources of waste right within a supervisor's department, so the first principle is to "spot a source of waste" within a small operation in that department, or among the many jobs that supervisor does himself.

The second step in applying waste control is to "write out the main work elements of the operation selected." This simply means to divide the operation into a number of separate elements. Each element of the operation should be listed in its regular sequence as the work is done. The easiest way to handle this is to observe the job while it is being done, and write down the elements exactly as they are seen.

#### Write Down Job Elements

Many practical operating men shy away from the task of writing out steps involved in an operation. They prefer to watch a job and, as it is being done, take whatever action seems to offer a quick way to reduce or eliminate waste. Generally, only a small part of the potential savings can be realized in this way.

The very act of writing out separate steps in an operation generates ideas. Many foremen state that as they observe a job closely enough to break it down into a number of separate steps, they immediately see sources of waste. This is excellent. It is exactly as it should be. However, a note of warning should be sounded now. Do not take action on any ideas to reduce waste while writing out the elements. Wait. Go through all six steps of waste control instead. Generally, when immediate action is taken only a small amount of waste is eliminated, and the balance of the operation continues to be performed in the same inefficient manner. When all six steps of waste control are applied, quite often the entire operation can be improved, not just a small part of it.

The practice of writing out all steps involved in an operation assures that the job is being analyzed in an orderly manner, and is almost a guarantee that the analysis will result in a saving.

The third step in applying waste control is the "ask questions about the work" to find a source of waste in each work element of the operation being analyzed.

[•] Foundry Specialist, Helmick & Associates, Canton, Ohio.

TABLE 1-JOB ELEMENTS AND WASTE CONTROL

Department		Item ob Breakd	lown	Date	
1					
2					
3					
4		14.			
5					
6		16.			
				*******	
7		17.		******	
8					
9					
10		m.			
10					
	Time	Effort	Space	Material	Power
What Causes					
Waste?					
Why is it					
done?	******				
	******			*****	*****
	*****	******		******	******
How can	******	*******		******	******
Waste be	******	******		******	******
eliminated?		******		******	
				*****	
		*****		******	
		******		******	******
	******	******			

When questioning elements to find a source of waste, the first question to ask is "What causes wasted time?" This question is asked of each work element. Not one element is missed. As the sources of waste are located, notes should be made so that they are not forgotten.

After asking "What causes wasted time?", each element is questioned as to "What causes wasted effort?", and notes are again taken. This same procedure is repeated for the other common causes of waste—space, materials and effort.

Many foremen ask, "What is the difference between wasted time and wasted effort?" There is a difference. Wasted time means idleness, loafing, or having nothing to do. Wasted effort means actually doing something, but not accomplishing any real work.

Information is gained by asking questions. Any foundryman can apply this set of questions and find sources of waste in operations in his department. Using questions such as these is a logical and intelligent method of attack on the problem of waste.

After locating the sources of waste something should be done about them. The next question to ask, then, "Why is it done?" This question frequently leads to eliminating the source of waste. This question alone—"Why is it done?" is the greatest single known way to eliminate waste. Simply by asking why, and answering the question honestly, eliminates so

much waste that often operations are not recognizable afterward because of changes, and some operations may even be eliminated completely.

As answers to questions are found, it is important that all ideas be written down, whether or not they seem impractical at the moment. Writing down ideas as they occur is the fourth step in applying waste control. An impractical idea may suddenly become the best one. Anyone who uses this systematic attack on waste will get many sound ideas, and writing them down is the best way to insure retaining them.

In applying these waste control principles, most effective results can be obtained if a simple form is used, such as that shown in Table 1. Space is provided on this form for listing the work elements of the operation, as well as for ideas to eliminate waste.

The fifth step in applying waste control principles is to obtain approval of the ideas. This step is not always necessary if the idea can be put into effect by the individual. If the idea involves an expenditure, changes inspection requirements, removes a hazard from an operation, or requires a worker to perform an operation differently, approval generally is necessary.

No one likes to change his method of work. Most people are prone to go along in the same routine. Therefore, the foundryman who applies waste control principles is likely to meet some resistance in obtaining acceptance of his ideas.

## Sell the Idea

While it is not the intent of this paper to list techniques of salesmanship, it is vital that the supervisor applying waste control principles observe five simple but fundamental rules of successful selling: (1) approach the proper person cheerfully with the idea; (2) get him interested quickly; (3) stress the advantages of the idea such as cost reduction, labor savings, safety; (4) keep asking for a trial of the new idea; and (5) persist in the presentation until approval is obtained. It goes without saying that the presentation should be enthusiastic.

There is another point of importance in getting ideas accepted. Remember to bait the hook to suit the fish. Before approaching anyone for approval, try to put yourself in his shoes. Visualize the things of interest to him in your proposition, then build your approach around your idea, keeping in mind those points that will appeal to the person you are trying to sell. Many sales managers state that the prospect makes up his mind to buy or not to buy during the first ten words a salesman speaks. After developing sound ideas for eliminating wasted time, effort, space, materials and power, do not allow your efforts to be wasted because of ineffective selling.

The sixth and final step in applying waste control principles is to "put the ideas into effect." What good is it to develop a way to eliminate or reduce waste in an operation if it is not made to work? None at all.

Another selling job must be done, when putting the idea into effect. The supervisor or time-study man usually must get someone to change his way of work-

#### TABLE 2-SUMMARY-WASTE CONTROL PRINCIPLES

1. Spot a Source of Waste:	Any operation being done in a department is a source of waste until it has been questioned.
2. Write Out the Work Done:	Consider the main steps re- quired to complete an opera- tion. Writing down the steps requires thought and reveals sources of waste.
3. Ask Questions About the Work:	Question each element. What causes waste of time, effort, space, materials, power? Why is each element done? How can this waste be eliminated?
4. WRITE DOWN IDEAS:	Assume every idea is sound, un- til it is rejected. Do not take action until sure the idea is sound.
5. GET THE IDEAS APPROVED:	Talk it over with others. Know all the angles to it. Do not overlook some minor point which may result in an idea being rejected.
6. Put the Idea Into Effect:	A good idea, worked out, but not put into effect, is also wasted time and effort.

ing. Rarely can he be persuaded to change without at least some resistance. People just are not made that way. Again, the idea must be enthusiastically presented, so much so that the other person will want the advantages pointed out to him. He must feel that he will gain by using the proposed method. Only when he feels this way will he want to put the idea into effect.

After a new idea is put into effect, the foundryman must follow up to see that it remains so. Many good systems are installed, procedures changed, operations improved, and then after a time the employee on the job slips back into the old familiar routine. Do not let this happen. Follow up the idea religiously after it is put into effect. Make sure that it is carried out as planned.

Supervisors can learn waste control principles in three 1-hr meetings. At the first meeting the instructor should present a well known foundry operation and guide the group in applying waste control principles. In this manner the group participates in solving an interesting problem, meanwhile learning the method of recognizing and eliminating waste.

A summary of waste control principles should be outlined before closing the first session. Each participant should be given a copy of the summary (Table 2). Applications of waste control principles should be requested on a volunteer basis for presentation when the group meets again. There is no better way to learn to reduce costs than by actually working on a problem in a systematic manner. It should be explained that before the next session the instructor will be glad to work with each volunteer on his problem.

The second meeting on waste control principles should emphasize questioning each work step in an operation, and the use of the form shown in Table 1 to note all ideas derived from the questioning. This meeting and part of the third meeting should also be devoted to a review and discussion of the "projects" submitted by the volunteers, to stimulate a practical

interest. The remainder of the third meeting should include a thorough discussion of the fifth and sixth steps in the Principles.

Waste control will work. The principles outlined in this paper get results. Our nation needs output of foundries more today than at any other time in its history. There is no better or simpler way for management men to increase tonnage per man-hour than to eliminate wasted time, effort, space, materials and power.

#### DISCUSSION

Chairman: L. W. LEHMAN, John Deere Van Brunt Co., Horicon, Wis.

Co-Chairman: H. R. WILLIAMS, Williams Management Engineering, Milwaukee.

GERALD NADLER (Written Discussion): 1 Mr. Walley's approach to the subject is excellent and can and does result in effective methods improvement.

Perhaps something that can be added to the paper might be an elaboration of what is meant by "main steps . . . (of) the operation." For example, it is possible to record as the main steps of an assembly job as "get parts 1 and 2 and assemble, get part 3 and assemble, place on paper, wrap, and aside." This type of breakdown overlooks much waste, whereas a somewhat more detailed breakdown would have given the complete picture of how the job was performed.

Time spent on emphasizing the human relations problem was especially worthwhile. I do not think that Mr. Walley need imply that the parts of his paper on the techniques of salesmanship are not part of work simplification. He has done a rather fine job on this aspect and it should be commended along with the general tenor and approach of his paper.

J. I. Hogan (Written Discussion): Suggestions recommended by the author to apply principles of waste control to foundry operations are applicable and practical to any foundry.

The fundamental procedure suggested—to list all elements of an operation or procedure of foundry operations is a standard practice used in recording the elements of an operation when taking a timestudy.

These foundries using timestudy as a means of determining man-hour production measurement will find a wealth of cost reduction material already listed in their timestudy files.

To make an additional recapitulation or spread sheet of these observed elements into the values of time, effort, space, material and power requirements would provide interesting data for cost analysis. Additional selections of these elements which might be classified as wasted time, wasted effort, wasted space, wasted material and power could easily be made. No search is required—select a timestudy and start an analysis and the results will be amazing.

However, it is the writer's firm opinion that a most important fundamental of successful cost reduction is entirely ignored by Mr. Walley. This is the human element in the shop. No mention is made or consideration given to "How does the proposed change effect other departments and plant operations as a whole?" "How will it effect labor relations and earnings of the operator?" "How will it effect other people in the supervisory group?" These elements are more important than analysis of work content in an operation.

The most important element of any successful cost reduction program is leadership and the approval and sanction of executive management. Mr. Walley places this important element of "sell the idea and get approval" as fifth in the procedure. This should be No. 1 on the list. Before consulting the files or making any analysis as suggested by Mr. Walley, the writer believes that proper procedure is to sell the idea of a cost analysis program to executive management. Only with such approval and sanction can such procedure be successful.

Furthermore, a shrewd, tactful and experienced man, well versed in company policies and shop practice must be appointed as a leader. Only with proper leadership can a cost analysis program be applicable and practical. Waste control will never be completed in three easy lessons.

Asst. Prof. of Industrial Engineering, Washington University, St. Louis.
 Industrial Engineer, Williams Management Engineering Co., Milwaukee.

# ALLOYING ZIRCONIUM TO MAGNESIUM

By

W. P. Saunders and F. P. Strieter*

#### ABSTRACT

An investigation was conducted on the merits of several zirconium containing materials for alloying this metal to magnesium. The study included the use of various mixtures containing ZrCl, several agents containing ZrF, various forms of metallic zirconium, and some master alloys. In general, it was found that several of the above materials were suitable. The use of a master alloy containing about 40% zirconium, balance magnesium, appears to possess many advantages over both metallic zirconium and zirconium salts.

#### Introduction

Extreme interest in magnesium alloys containing zirconium has stimulated much published work. The extent of this work has perhaps resulted in confusion regarding methods of alloying zirconium. This paper has as its aims the clarification of some of the confusion concerning alloying agents, and the presentation of conclusions which show that each type of agent has some use, but that the zirconium chlorides and certain master alloys are suitable from a practical and economic standpoint.

Properties of Mg-Zn-Zr alloys, Mg-rare earth metal-Zr alloys, and Mg-Zn-rare earth metal-Zr alloys have been published by German, American, Canadian, and British workers, and a companion paper covering properties of the Mg-5Zn-Zr alloy (designated ZK51A alloy according to the ASTM system of nomenclature) is currently being published.

The experience of many years has been evaluated and is described in this report. It deals essentially with laboratory-size melts, but much of the work discussed here has been successfully used in production facilities.

It has been found that careful but, nevertheless, straight-forward techniques are all that are necessary for the successful production of these alloys. Since the development of proper metal-handling techniques, the authors' company has produced over one-half million pounds of magnesium-zirconium alloys made via chloride alloying agents. Experience has shown no flux problem to exist with the production techniques and, therefore, a study of flux inclusions was not an object of this investigation.

 Metallurgical Laboratories, The Dow Chemical Co., Midland, Mich. The alloying of zirconium to magnesium originated in Germany shortly before World War II. These investigators alloyed zirconium from ZrCl₄ or from mixtures of chlorides containing ZrCl₄. British investigators carried on the study of zirconium alloying agents, working initially with chloride agents and then going over to the fluorides of zirconium with or without other salts.

American and Canadian investigators carried on parallel development work during the past 10 years. This work was based primarily on the chlorides for production operations, but covered also the use of fluorides, zirconium metal in various forms, and hardeners of zirconium with other metals. A rather complete historical development on the alloying of zirconium is included in an appendix to this paper. The bibliography includes the references cited in the appendix.

#### **Experimental Procedures**

The materials used in the study of various agents for alloying zirconium were: commercial cell magnesium of 99.8 per cent purity and, in some instances, high purity sublimed magnesium; zinc of 99.99 per cent purity; commercial mischmetal (50.5 per cent Ce, 22.5 per cent La, 18 per cent Nd, 6.5 per cent Pr, and 2.5 per cent other rare earth metals and impurities); commercially available zirconium materials; and , laboratory developed zirconium materials.

Because of the higher maintenance and operating costs involved at high temperatures and the increased solubility of iron in magnesium at temperatures above 1400 F, most alloying work reported here was at about 1400 F. All results shown herein, unless stated otherwise, are those obtained at 1400 F. Some melts required higher temperatures during the alloying operations, e.g., >1700 F, but in all cases the final analyses were obtained after lowering the temperature to 1400 F.

The greater part of this work was done with standard steel crucibles, steel tools, and steel thermocouple protection tubes. In certain instances, clay-graphite crucibles and graphite tools were used.

Most melts were of 10-20-lb size, although a few

melts as low as 1000 grams and as high as 500-lb were made.

Most of the melts were made in a regular gas-fired crucible setting, and in a few instances oil-fired settings or an electric-resistance furnace were used.

In all melts not involving rare-earth metals, regular and modified crucible fluxes were used. Those melts containing rare-earth metals were protected with a flux which was free from MgCl₂. In general, the melts were made by the "crucible method" as described by Nelson.⁵²

For routine analyses, spectroscopic analyses were used. In all cases of zirconium analyses, gravimetric⁵⁸ or colorimetric⁵⁴ determinations were made. Electron and x-ray diffraction analyses were occasionally obtained.

Zirconium analyses reported herein are all as "soluble zirconium," or that portion of the zirconium present in the alloy which is soluble in dilute sulphuric acid. Analyses of hundreds of magnesium-zirconium melts show that after settling at 1400 F some "insoluble zirconium" is present. Statistically, this averages about 0.07 per cent in laboratory-controlled melts, and may be higher if the melt is stirred or otherwise disturbed before the sample is taken. Spectrographic analysis of Zr reports "total" Zr, or the sum of the insoluble and soluble forms, and may thus be used as a tool in production if the melt is properly controlled and sampled.

Analytical procedures^{53,54} have been established to determine accurately the soluble and insoluble portions of the zirconium in magnesium alloys. It has been found that any time from 15 to 60 min is satisfactory for dissolving the sample in dilute H₂SO₄. Refluxing experiments with an alloy containing a high amount of insoluble zirconium have shown that only a small amount of the insoluble zirconium can

be taken into solution. This information indicates that the time for normal solution in the acid is not critical, but the standard procedures were followed in order to assure that the correct soluble zirconium values were obtained.

The melting and alloying techniques were as follows: the crucible was fired to a red heat and carefully spudded down with a steel bar. The loosened material was scraped out and discarded. The walls of the crucible were then wetted with an inspissating type of flux and solid magnesium was charged and lightly sprinkled with flux. Flux was added as needed to control burning during the melt-down. At 1400 F zinc and/or mischmetal (when used) were added. Normally, the zirconium agent was then added, in a fashion dictated by its form, at 1400 F. The melt was then held quiescent for 10-15 min at 1400 F and sampled for analyses. Sometimes it was necessary to increase the temperature in order to try to increase solution of zirconium.

When this was done the melt was cooled to 1400 F and held quietly before it was sampled. A dusting agent of sulphur and boric acid was used to protect the molten alloy against oxidation during the pouring operation.

#### Discussion

Occasionally, the molten zirconium-free alloy was poured onto the zirconium agent in another crucible.

The zirconium was added in many different forms. The various categories were: metallic forms of zirconium, zirconium chloride combinations, zirconium fluoride mixtures, and master alloys containing zirconium. Since different techniques were frequently required with these various alloying agents, a brief description of the procedure used in each case is included in the discussion.

Table 1-Alloying Efficiencies of Various Metallic Forms of Zirconium and of Briquetted  $ZrCl_4$  Added to Magnesium or Mg + 5 Per Cent Zn at 1400 F

		Soluble Zi	rconium for Ad	ditions of		
	Alloved	1% Zr	2% Zr	3% Zr		
Material, Form	to	%	%	%	Remarks	
Fused Zr, lump	Mg	0.03		-	30 min stirring	
Iodide Zr sheet squares,	Mg					
0.005 x 1/4 x 1/4 in.	0	0.02			20 min	
		0.03			40 min Intermittent	
		0.05			55 min stirring	
		0.10			65 min	
er sponge	Mg	0.32	0.48	0.62	3-4 min stirring	
		0.50	0.57	0.66		
		0.52	-		,	
er sponge						
$\frac{1}{4}$ + 6 mesh	Mg + 5Zn	0.44				
6 + 10 mesh	Mg + 5Zn	0.44			`	
10 + 20 mesh	Mg + 5Zn	0.38				
28	Mg + 5Zn	0.30			Some burning	
100	Mg + 5Zn	0.0			Flash burning	
owdered Zr, Briquetted					Excessive alloying	
nd Sintered	Mg + 5Zn	0.39	0.63	0.76	time required	
riquetted ZrCl.	Mg + 5Zn	0.56	0.78	0.80		
		0.65	0.80	0.85		
			0.87			

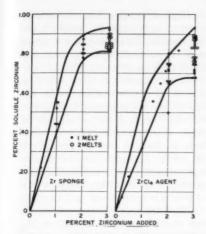


Fig. 1 — Alloying zirconium as Zr sponge and fused ZrCl_i + KCl or KCl + NaCl salts to Mg + 5% Zn alloy at 1400 F.

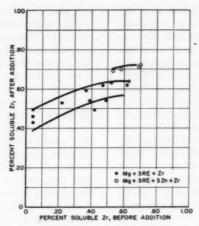


Fig. 2—Effect of the addition of 1% Zr sponge at 1400 F on the zirconium content of Mg + 3RE + Zr & Mg + 3RE + 3Zn + Zr alloys.

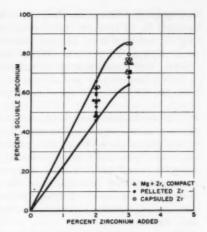


Fig. 3—Alloying zirconium as pelleted or capsuled zirconium powder to MG + 5% Zn alloy at 1400 F.

Metallic Forms of Zirconium: The alloying of Zr with the following metallic forms was studied: zirconium sponge, 56,57 fused zirconium, iodide-decomposed ductile Zr55 rolled to 0.005 in., and zirconium powder produced by the co-reduction of CaH₂ and ZrO₂.

Figure 1 shows the results of adding zirconium sponge to Mg-5Zn; Fig. 2 presents the results of adding an additional 1 per cent zirconium as sponge to Mg-rare earth metal-Zr and Mg-Zn-rare earth metal-Zr alloys; and Fig. 3 contains some information on the alloying efficiency of zirconium powder.

The data indicate that zirconium sponge produced a soluble zirconium content in Mg-5Zn-Zr alloys of 0.80 to 0.90 per cent with a 3 per cent addition of the sponge. The sponge used in these tests was obtained from the U. S. Bureau of Mines, and was made by the Kroll process. 56,57 The results of these experiments should be kept in mind throughout this paper as they represent the most consistently good alloying efficiencies obtained.

# Alloying With Zirconium Sponge

An important disadvantage in the commercial use of sponge in the magnesium alloying field is the rather laborious effort required to alloy the material. In these experiments, the sponge was essentially ground while submerged in the molten Mg-Zr alloy. Metallographic examination shows that the average particle size of the sponge is about 0.0005 in. The particles appear to be lightly sintered to one another. Thus the apparent mass of the sponge is dense enough to permit sinking in the melt and, once submerged, may be crushed or abraded down to a surface to mass ratio which permits fairly rapid solution of the zirconium in magnesium.

Table 1 presents some additional information showing that 0.62 to 0.66 per cent zirconium can be alloyed to commercial magnesium with a 3 per cent addition of sponge. This indicates that 5 per cent zinc increases the solubility of Zr in magnesium by about 0.2 per cent.

The data shown in Fig. 2 are included to show the efficiency with which sponge can be used to "sweeten" the Mg-rare earth metal-Zr and the Mg-rare earth metal-Zn-Zr alloys. It also indicates approximately the differences in Zr solubility limits for these two alloys. The alloy containing zinc is seen to have a solubility of slightly over 0.7 per cent Zr, while the alloy without zinc has about 0.1 per cent less soluble Zr.

The fused lump zirconium was added as 1/4-in. pieces in a small steel ladle and stirred in the ladle with a steel rod. No apparent solution had occurred after 30 min of stirring. Analysis of the melt as indicated in Table 1 showed an alloying efficiency of only 3 per cent with this material.

Iodide zirconium sheet, rolled to about 0.005–0.010 in. and cut into 1/4-in. squares was added in a manner similar to that used for the fused lump zirconium. It was stirred for several minutes in the ladle and then intermittently stirred for several minutes at a time during the 65-min period tested. As shown in Table 1 solution occurred very slowly and only 10 per cent of it was alloyed after more than an hour had elapsed.

Since the solution of metallic zirconium in magnesium (as with all difficultly soluble materials) is dependent upon time and particle size, the use of finer particles of zirconium metal was evaluated.

Table I shows that as zirconium sponge fragments were decreased in size, the alloying efficiency decreased. In fact, when the particles became finer than 100-mesh powder, the material burned up before it could be submerged beneath the metal.

Zirconium powder is pyrophoric, and some means of protecting the powder from oxidation had to be applied. Materials tested were Zr powder pelleted with various binders, Zr powder enclosed in tight magnesium capsules, sintered Zr powder briquettes, and Zr powder compacted with magnesium powder.

It can be seen by comparing Fig. 3 with Fig. 1 that the zirconium powder, when compacted with magnesium or when pelleted with organic binders, is not quite as efficient in alloying to Mg-5Zn as is the sponge Zr. Three per cent additions of zirconium powder resulted in an average soluble zirconium content of about 0.7 per cent. When the Zr powder was packed into capsules made of magnesium, it became a little more efficient and alloyed to the level of 0.75 per cent to 0.85 per cent at the 3 per cent addition level.

The zirconium powder alloyed more rapidly than sponge, as was expected because of the particle size. It has been reported⁵⁸ that the reaction of Zr with H₂, O₂, and N₂ increases as temperature increases. Since the pellets and compacts were not protected from the atmosphere, it is probable that the zirconium combined with one or all of these elements, and thus the amount of metallic zirconium added was less than that stated.

The sintered briquetted zirconium powder reacted somewhat like the fused lump zirconium in that after 30 min of stirring of the 3/8-in. diameter briuse for direct alloying to magnesium. Briquetted ZrCl₄, however, is much more readily handled and presented interesting results. This material was added to the molten alloy as 1½ in. diameter briquettes, and held below the surface with an inverted ladle or with a perforated plate. The magnesium alloy failed to wet the briquettes and no fragmentation occurred.

The time required for the alloying of the zirconium from the briquettes was slightly longer than that required by the sponge. However, the briquettes were not crushed beneath the melt as was the sponge. It is believed that the ZrCl₄ vaporized slowly from the briquettes because of a vapor barrier which insulated the heat of the melt from the main body of the briquette. Table 1 shows not only that a 3 per cent addition of Zr in this form produced a soluble Zr content in the range of 0.80 to 0.85 per cent in Mg-5Zn alloys, but also that a 2 per cent addition gave results about as good as the 3 per cent addition. In this respect the ZrCl₄ briquettes were essentially as

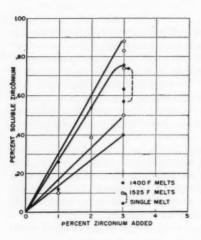


Fig. 4 (left)—The effect of temperature on the alloying of zirconium as fused ZrCl_i + ZrF_i + KCl salts to Mg + 5% Zn alloy.

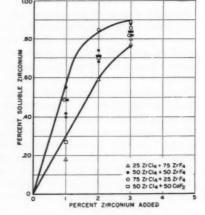


Fig. 5 (right)—Alloying zirconium as briquettes of ZrCl₁ + ZrF₁ or ZrCl₂ + CaF₂ to Mg + 5% Zn alloy at 1400 F.

quettes little crushing had occurred. In the next series of tests the briquettes were quartered with a cold chisel, and after 30 min of stirring at 1400 F the 3 per cent addition had dissolved 0.76 per cent zirconium into the Mg-5Zn alloy, as shown in Table 1. Metallographic examination of the briquettes showed that the average particle size was about 0.001 in., and that the particles were quite strongly sintered together.

Chlorides of Zirconium: Alloying studies with the following chloride forms were conducted:

1. Briquetted ZrCl4.

2. Fused ZrCl₄ + KCl, or ZrCl₄ + KCl + NaCl.

3. Fused  $ZrCl_4 + ZrF_4 + KCl$ .

4. Briquetted ZrCl₄ + ZrF₄, or ZrCl₄ + CaF₂.

The use of some of these forms has been suggested in the literature as described in the appendix, but little data have been shown. Table 1 and Figs. 1, 4, and 5 present the analytical results.

ZrCl₄ is a light, fluffy powder which sublimes at 331 C (628 F) and is extremely hygroscopic. When a small sample is poured on a dry watchglass, the results of hydrolysis are immediately apparent by the odor of HCl and the visible color change that takes place. The nature of the material discourages its

good as sponge in alloying efficiency.

Proprietary fused salt mixtures containing 50 per cent ZrCl₄ + 25 per cent NaCl + 25 per cent KCl and 50 per cent ZrCl₄ + 50 KCl are commercially available. Figure 1 shows the efficiencies experienced when the solid fused salt was added to Mg + 5 per cent Zn at 1400 F. A comparison of the Zr sponge range with the range for these salts indicated that the salts were slightly lower in their average efficiency and the range of scatter was somewhat broader.

The ease of alloying with this type of salt was readily evident in the fact that only several minutes were required to completely react the material. Since the salts contain only about 20 per cent zirconium, it is evident that a larger amount of alloying material has to be handled compared to the pure metallic additions. This minor disadvantage will be evident also in the discussion of other alloying agents.

Incidentally, the sodium and potassium chlorides are inert to magnesium so there is essentially no alloying of other elements to the melt coincidental to the Zr alloying.

The data in Fig. 4 demonstrate that 40-50 per cent ZrCl₄ + 20-25 per cent KCl fused with 40-25 per cent ZrF₄ is alloyable, and that the alloying effi-

ciency is increased by the use of temperatures higher than 1400 F. A 3 per cent Zr addition, when alloyed at 1525 F, produced soluble Zr contents of 0.50 to 0.85 per cent in Mg-5Zn alloys. However, the average zirconium content was lower, and the scatter was greater, than that obtained with the sponge or with the all-chloride fused-salt mixtures.

Figure 5 presents interesting data when compared to Figs. 1 and 4. Here, when ZrF4 was present in the solid dry briquetted state, with varying amounts of ZrCl4, the alloying efficiencies were higher than they were in the fused condition and they compared quite well with those experienced with the fused all-chlor-

ide salts and metallic zirconium sponge.

The time of alloying was less than that required for briquetted ZrCl4, alone. It is assumed that the poorer results obtained by the fusion process in the previous tests, the results of which are shown in Fig. 4, may have been caused by the formation of the oxyfluoride or the formation of a more difficultly reducible complex salt. It is also interesting to note that a briquetted product consisting of ZrCl4 and CaF₂ was about as efficient in alloying zirconium to the melt as briquetted ZrF₄ + ZrCl₄. This might be a less expensive product that the ZrCl₄ + ZrF₄ combination.

Fluorides of Zirconium: The following types of fluoride salts were studied:

- 1. ZrF4, powder and briquetted.
- K₂ZrF₆ + alkali metal chlorides.
   ZrF₄ · KF + chlorides.
- 4. ZrF₄ + other fluorides.

5. ZrF₄ + KCl salts.

Figures 6, 7, 8, and 9 show the results of these ex-

Early experimental work with ZrF4 indicated that high temperatures (above 1500 F) and large additions (5 per cent Zr) were necessary for moderate success. The experiments reported here tend to confirm the

earlier experiments.

Adding ZrF4 powder to molten magnesium at any temperature is unpleasant work. The powder is dry and not wet by the magnesium, and much flux must be added to control the burning. Figure 6 shows the wide (and low) range of alloying efficiency obtained when ZrF4 was added to magnesium + 5 per cent Zn alloy. The data indicate that at 1400 F a 5 per cent zirconium addition produced only about 0.50 to 0.60 per cent soluble zirconium in a Mg-5Zn alloy. Raising the temperature to 1550-1650 F increased the zirconium content about 0.1 per cent, as shown in Table 2.

The reaction product MgF2 was hard and dry and separated well from the melt, but probably inhibited the reaction by coating the zirconium with an insulating product. Extra stirring, in an attempt to wear off this coating, with the use of an air-motor powered mechanical stirrer failed to improve the efficiency over that of hand stirring, as indicated in

The alloying of K₂ZrF₆ has been studied and the alloying results are found in Fig. 7.

The majority of the work was done with K₂ZrF₆ + 20 to 50 per cent KCl, and the salts were added in various manners: as prefused lumps, prepared as molten salts and the alloy poured on, or prepared as molten salts and added to the molten alloy. All were found to be equally efficient; however, a note of explanation as to why these techniques were used is

In the early stages of work, the alloy was prepared by adding the alloy to the molten salts. In several occasions, violent reactions resulted and caused immediate concern. By means of several thermocouples the temperatures in various parts of the crucible were determined. If the salt was at 1300 F and the metal being poured onto the salt at 1325 F, the temperature at the bottom of the crucible frequently ran to greater than 1700 F within a matter of seconds. Local hot spots at the salt-Mg interface probably exceeded the boiling point of magnesium and thus hurled the molten alloy from the crucible. By pour-

TABLE 2-ALLOYING EFFICIENCY OF ZRF4 ADDED TO MG + 5% ZN Alloy

	*		Soluble Zirconium for additions of								
		Temp.,	1.25 % Zr	2.5 % Zr	3.75 % Zr	5.0 % Zr	10.0 % Zr				
Form	Treatment	F	%	%	%	%	%				
Powder	hand stir	1400	_	-	_	0.19	_				
	motor stir	1400	-	-	_	0.58	_				
	hand stir	1400	-	-	_	0.56	-				
	motor stir )	1400	_	_	_	0.16	_				
	Add 5% Zr	1400	-	_	_	-	0.59				
	motor stir	1400	_	-	-	0.45	_				
	Add 5% Zr	1400		-	_	-	0.53				
Briquette	hand stir	1400	-	0.43	-	-	_				
	hand stir	1400	-	_		0.50	-				
	heat & stir	1650	_	_	-	0.50	_				
	hand stir	1500		0.41	_	_	_				
	heat & stir	1650	_	0.51	_	-	-				
	hand stir	1400	0.11	-		_	_				
	Add 21/2% Zr	1400		_	0.35	_	_				
	heat & stir	1550	_	-	0.47	-	_				

ing the smaller quantity of salts into the Mg, the heat generated could be harmlessly dissipated.

The exothermic nature of the ZrF4-base reductions was always noticed when measured by ordinary thermocopules, and was much greater than that in the ZrCl4-base reductions. This rise in temperature of the melt was directly related to the higher rate of reaction of the fluoride, or to the higher heat of formation of MgF2 (than MgCl2), or probably to both. In no case has a violent reaction resulted when the salt, either liquid or solid, was added to the molten metal and immediately stirred.

Figure 7 shows that the alloying efficiency of K₂ZrF₆ · KCl was lower and the results were more scattered than in the cases of the metallic zirconium or ZrCl4-base reactions. A 5 per cent addition of zirconium produced zirconium contents ranging from 0.33 to 0.77 per cent.

There is some interest in the ZrF₄ · KF (KZrF₅) salts. Table 3 presents the data for this type of material. The alloying efficiencies were low with only 0.34 per cent soluble zirconium obtained with a 4 per cent addition.

Another significant fact shown in this table is the

presence of barium which was alloyed to the melt.

It is seen that in the  $ZrF_4 + KF + KCl$  melt, 0.0004 per cent Ba was found; in the  $ZrF_4 + KF + KCl + BaCl_2$  mixture, 0.0155 per cent Ba was detected; and in the  $ZrF_4 + KF + BaCl_2$  melt, 0.0165 per cent Ba was found. It can be calculated that if only a small amount of  $BaCl_2$  were converted to  $BaF_2$ , a rather large amount, by comparison, of Ba could be expected in the melt.

The presence of Ba in the alloy made with the  $ZrF_4 + KF + KCl$  addition is explained by the presence of about 20 per cent  $BaCl_2$  in the melting and

refining flux.

In melts alloyed with chloride salts or metallic Zr derived from ZrCl₄ and using the same BaCl₂-containing flux, the amount of Ba present is less than 0.0003 per cent (if present at all), 0.0003 per cent being the lower limit of accuracy for Ba analyses.

Emley³⁶ has shown the probable order of the more

1600 F yielded soluble zirconium contents above 0.80 per cent. This was accomplished with the addition of 4.2 per cent Zr. At 1400 F there was less than 0.7 per cent Zr alloyed with the same 4.2 per cent addition of "inert fluorides." Table 5 shows this same information, and in addition lists the analyses of the impurity elements. Even the inert fluorides reacted to some extent with magnesium since several hundredths per cent of lithium and barium were reduced and alloyed.

It has been stated in the literature³⁶ that ZrF₄ is insoluble in alkali metal chlorides. Our analyses lead us to disagree with this statement. Fused mixtures of ZrF₄ and KCl, when examined by x-ray diffraction techniques, showed neither salt to be present, but rather an unidentified single-phase product.

This fused salt mixture was added to Mg + 5Zn alloys with results as shown in Fig. 9. A 5 per cent Zr addition produced soluble zirconium contents of

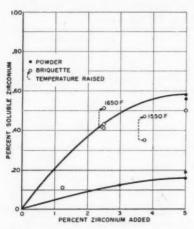


Fig. 6—Alloying zirconium as ZrF, to Mg + 5% Zn alloy at 1400 F.

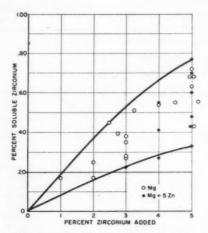


Fig. 7—Alloying zirconium as fused K:ZrF:-alkali metal chloride salts to magnesium or Mg + 5% Zn alloy.

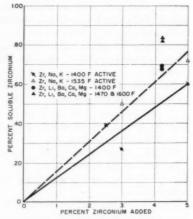


Fig. 8—The effect of temperature on the alloying of zirconium as "active" and "inert" florides of zirconium added to Mg + 5% Zn alloy.

electro-positive metals in both the chloride and fluoride electropotential series. His series is reproduced in Table 4. This important work indicates that magnesium is considerably more electropositive in the fluoride series than it is in the chloride series, and also shows that sodium and potassium fluorides are less positive than magnesium.

Also, BaCl₂ is considerably more electropositive in respect to magnesium than is BaF₂. This situation probably accounts for the reason why some barium was found in the melt when ZrF₄ + BaCl₂ was used as compared with ZrCl₄ + BaCl₂—a certain amount of the BaCl₂ being converted to BaF₂ prior to being reduced to elemental barium.

A few tests were conducted to study the effect of the position of the various metals in the above electropotential series. An alloying agent consisted of ZrF₄
plus NaF and KF (reducible by Mg), while others were made of ZrF₄ plus "inert fluorides" (Ca, Li, and Ba).

Figure 8 lists the results of these tests and shows that the "inert fluorides" when alloyed at 1470 to

0.64 to 0.76 per cent. In this respect the material was comparable in alloying efficiency to the  $K_2ZrF_6$  · KCl salt mixtures.

Master Alloys Containing Metallic Zirconium: Several different master alloys containing zirconium and one Zr + MgCl₂ mix were investigated in an attempt to improve upon the alloying characteristics of the metallic zirconium and the various salt mixtures. Several proprietary master alloys were studied as well as a laboratory-produced Mg-Zr hardener.

Since the last operation of producing zirconium sponge by the Kroll process^{56,57} (sublimation of Mg and MgCl₂ from the reaction product of Mg + ZrCl₄) is expensive, the product, before sublimation, was tested. The problem of storing this hygroscopic material is present, and Table 6 indicates the efficiencies are not good in either the as-received or water-leached and dried condition, 5 per cent efficiency for the 2 per cent water-leached addition and 10 per cent efficiency for the 5 per cent unleached addition.

Ferro-zirconium has been manufactured for many years and has been used in the iron and steel in-

TABLE 3—ALLOYING EFFICIENCY OF VARIOUS ZrF₄—KF SALTS ADDED TO MAGNESIUM

	Soluble Zr for additions of										
	1% Zr	2% Zr	3% Zr	4% Zr	5% Zr						
Material	%	%	%	%	%	Ba, %					
ZrF ₄ +KF	_	_	_	0.06	_						
ZrF4+KF+KCl	0.18	_	_	_	_	0.0004					
ZrF+KF+KCl+BaCl	_	-	_	0.34	-	0.0155					
ZrF4+KF+BaCl2	-		-	0.16	_	0.0165					
$ZrF_4 + K_2ZrF_6 = 2ZrF_4$	+2KF			0.12							

TABLE 4—PROBABLE POSITIONS OF THE MORE ELECTRO-POSITIVE METALS IN THE CHLORIDE AND FLUORIDE ELEC-TROPOTENTIAL SERIES AT OPERATING TEMPERATURES

Chlorides: (Ba, K), (Sr, Li, Ca, Na), (Ce, Mg), Al, . . . Fluorides: Ca, Li, Ba, (Mg, Co), Al, Na, K, . . . (The electropotential decreases from left to right) (Information obtained from Table 1 of Emley*)

dustry. Even though it has been a well known fact that iron co-precipitates zirconium from magnesium melts, it was felt necessary to check this material as a zirconium alloying agent. The results of the tests are shown in Table 6 where it is shown that there was no retained zirconium in the melt when 5 per cent Zr was added.

A master alloy containing 60Zr and 40Mg has been commercially available for the past several years. Results with it were erratic and high zirconium contents were not consistently achieved. A typical example of its alloyability is shown in Table 6.

Another master alloy, purchased from a different company and containing 28 per cent Zr+71 per cent Mg, showed promise but was excessive in cost. Table 6 shows that 0.74 per cent soluble Zr was alloyed to Mg + 5 per cent Zr with a 3 per cent Zr addition.

Another master alloy which contained 14 per cent Zr-70 per cent Zn-balance Mg, showed excellent Zr recovery for a 1 per cent Zr addition, from 0.60-0.72 per cent soluble Zr, and simultaneously alloyed 4.5 per cent zinc. Thus, it is not possible to add more

zirconium to a melt (to obtain a higher Zr level) without also increasing the zinc content. Therefore, the utility of this material is limited to high zinc content alloys.

The use of a master alloy made in the laboratory by the reduction of ZrCl₄, with magnesium was tested under laboratory and production conditions. The material was prepared by reacting magnesium at 1400 F with about four times its weight of a proprietary flux containing about 50 per cent ZrCl₄ and 50 per cent KCl. A metallic reaction product was obtained which contained 30 to 50 per cent zirconium, balance magnesium, except for a small amount of residual chlorides. This product for convenience is labeled K30A alloy in accordance with the ASTM system of designating magnesium alloys.

It is a convenient and efficient material for alloying zirconium to magnesium alloys as shown by the data in Fig. 10. Three per cent additions of zirconium as K30A alloy resulted in soluble zirconium contents of 0.70 to 0.90 per cent in Mg-5Zn alloys. This is nearly as good as zirconium sponge and is equal to the ZrCl₄ fused salt mixtures.

This master alloy combines the several advantages which are individually exhibited by the other various

Table 6—Alloying Efficiencies of Various Zr Master Alloys and Mixes Added to Mg + 5% Zn Alloy at 1400 F

	So	luble Zi	for Ad	lditions	of	
	1% Zr	2% Zr	3% Zr	4% Zr	5% Zr	
Material	%	%	%	%	%	
40Zr+60MgCl ₂ (Above—H ₂ O	-	_	_		0.51	
leached)	Statement .	0.11	_	_	-	
Ferro-Zirconium	_	_	_	_	0.00	1440F
	_	-	-	_	0.00	1500F
	-	-	-	-	0.00	1600F
60Zr-40Mg	_	_	0.38	_	_	
28Zr-71Mg	_	_	0.74	-	-	
	_	_	0.74	_	_	
14Zr-70Zn						
(Bal. Mg)	0.72	_	_	_	_	
,	0.66	-	-	_	_	
	0.60	_	-	-	-	

TABLE 5-THE ALLOYING EFFICIENCY OF ZR FROM ZRF4 FUSED WITH "ACTIVE" AND "INERT" FLUORIDE MIXTURES

		Per	r Cent S	ioluble Zr + 1	mpuriti	es for Addition	n of		Alloying
	2.5% Zr			3.0% Zr		4.2% Zr		Tempera ture,	
Material	% Zr	% Impurity	% Zr	% Impurity	% Zr	% Impurity	% Zr	% Impurity	F
84ZrF ₄ +9KF+7NaF	0.39	0.005 Na					0.60	0.005 Na	1400
							0.72	0.010 Na	1550
			0.27	0.005 Na					1400
			0.50	0.005 Na					1525
43ZrF ₄ +8LiF+41BaF ₂ +8MgF ₄					0.68	0.10 Ba			1400
						0.045 Li			
						0.0012 Ca			
					0.84	0.096 Ba			1600
*						0.06 Li			
4						0.0025 Ce			
43ZrF ₄ +16LiF+27BaF ₂ +13CaF ₂					0.69	0.070 Ba			1400
						0.06 Li			
						0.0086 Ca			
					0.83	0.047 Ba			1470
						0.06 Li			
						0.010 Ca			

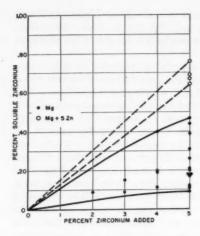


Fig. 9-Alloying zirconium as ZrF, fused with KCl salts to magnesium or Mg + 5% Zn alloy at 1400 F.

agents. Metallic zirconium possesses the advantage of alloving zirconium without the necessity for adding extraneous salts, and K30A does essentially the same thing. The major component in the master alloy aside from zirconium is magnesium, and it combines readily with the alloy and is therefore useful.

ZrCl4 fused salt mixtures are readily alloyable to magnesium as compared with the sponge zirconium and the fluorides, and K30A alloy also alloys readily to magnesium alloys. Both the fluorides and the metallic forms of zirconium have the advantage of forming the alloy without the large amount of fluid salts which need to be carefully settled out of the melt, such as is required with the chloride agents. K30A likewise possesses an insignificant amount of chloride salt, so that this problem is not present.

The problem of large amounts of fluid salt being present from the chloride agent has another disadvantage which is solved by the use of K30A alloy. The fluid salts contain rather large quantities of MgCl2 as a result of the reaction between Mg and ZrCl₄. The MgCl₂ is detrimental in Mg alloys containing rare earth metals and zirconium because the MgCl2 reacts with the rare earth metals, taking them out of the alloy. The K30A alloy contains at most only a small amount of MgCl2 and its reaction with the rare earth metals is insignificant,

Table 7 presents some data on the "sweetening" ability of K30A alloy on some 300-500 lb production melts of Mg-3 rare earth metal-zirconium melts. It shows that 0.5 per cent additions are about 40 per cent efficient at the level of zero to 0.3 per cent initial zirconium content.

#### Conclusions

The foregoing information indicates that several zirconium-containing materials can be used successfully to alloy zirconium to magnesium and its alloys.

Zirconium sponge produced by the U.S. Bureau of Mines is particularly useful in this respect because it produces consistently high soluble zirconium contents. The alloying technique is slow and laborious.

ZrCl4 base materials, both fused and briquetted, alloy readily and nearly as efficiently as the sponge. Their use is practical in many applications.

ZrF4 combined in its many forms with other salts may be used as alloying agents if desired. They seem to be less efficient and to require a higher alloying temperature. More labor is required to alloy zirconium from fluorides than from chlorides because of the more viscous nature of the resulting sludge.

A master alloy made by reacting magnesium with ZrCl4 is an especially good agent for alloying zirconium to magnesium alloys. It alloys readily, possesses essentially no excess salt constituents, and is readily available for use by any foundry handling

molten magnesium alloys.

The decision regarding exactly which one of the above agents to use must be made by the individual foundry according to its own desires. Not only must the raw material costs be considered in arriving at the economics of the various zirconium alloying agents, but also other factors such as the labor of alloying, operating temperatures, reactivity of resulting salt residues with the other alloy constituents, maintenance costs, convenience, etc., must be con-

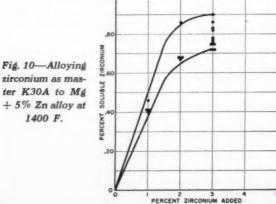
It is recognized that the general field of alloying of zirconium to magnesium is one which is very active at this time. Many different groups have been and still are working to find improved methods for obtaining the desired zirconium contents in the alloys. Many patents have been granted in this field, many of which are mentioned in the appendix and in the bibliography. More patents will undoubtedly be issued in the future. Each patent must be thoroughly evaluated before its use can be considered.

## **APPENDIX**

# Historical Development

Early German work¹ reports the alloying of zirconium to magnesium by the reduction of ZrCl4 or K₂ZrCl₆ with magnesium. A British patent² in 1939, covering German work, teaches that Al, Si, Mn, Sn, Co, Ni, and Sb are elements which inhibit the alloying of Zr; while Zn, Cd, Ce, Ag, Tl, Th, Bi, Be, Pb, and Ca are compatible with Zr. Another British patent in the same year3 indicates that Zr counters the grain coarsening caused by the addition of Be to magnesium alloys.

In 19404 it was stated that a successful way to alloy zirconium was with the use of molten ZrCl₄ + KCl.



zirconium as master K30A to Mg + 5% Zn alloy at

Sauerwald⁵ described the I. G. Farbenindustrie work from 1937-1939 and tells of the use of powdered zirconium, and that the solid solubility of zirconium in magnesium is about 4.5 per cent. British work in 1941⁶ claims that molded pieces of ZrCl₄ were used, and that CaF₂ was added to separate the MgCl₂ liquid reaction product from the alloy, and that zirconium metal alloys only slowly to magnesium. In published work of 1943 Sauerwald et al.^{7,8,9} indicate that alloys of Mg-Zr with Ag, Cu, Ce, Th, and Ca contained no microshrinkage.

The patents of Stroup and Sager in 1945^{10,11} deal with the use of ZrCl₄ in settling Fe in magnesium. In 1946 Doan and Ansel¹² described alloying with 50 per cent ZrCl₄ and 50 per cent alkali metal chlorides. Sauerwald¹³ in 1947 claims that the effect of grain refinement in Mg-Zr alloys depends upon true solution, that zirconium in solid solution in magnesium dissolves readily in HCl and any elemental zirconium

Table 7—Effect of the Addition of 0.5% Zr as Master Alloy K30A Upon the Zr Content of 300-to 500-lb Melis of Mg-3 Rare Earth Metal-Zr Alloys

Soluble Zr Before Addition, %	Soluble Zr After Addition, %
0.01	0.18
0.18	0.41
0.01	0.14
0.14	0.38
0.04	0.29
0.29	0.47
0.003	0.15
0.15	0.37
0.02	0.28
0.28	0.45
0.02	0.31
0.31	0.46
0.11	0.35
0.10	0.25

is not dissolved in the acid, and that hydrogen precipitates zirconium.

Ball¹⁴ in the same year describes the British work and states that metallic zirconium dissolves in magnesium under an argon atmosphere at 900-1000 C (1652-1832 F) and is a difficult and costly process; K₂ZrF₆, ZrF₄, ZrO₂, ZrN, and ZrS were tried and only the fluorides showed promise; and that sinks are found in poorly-fed sections of Mg + 5Zn + Zr alloy rather than microporosity. Emley¹⁵ in 1948 indicates that zirconium powder is expensive and inflammable, that the oxide is refractory, double chloride salts with alkali metal chlorides are volatile and hygroscopic, that high temperatures of 920-950 C (1688-1742 F) and prolonged stirring are necessary if ZrF4 is used, that ZrF₄ + KF results in alloying zirconium and explosively reducing the KF, but that certain chlorides suppress the violence, and that MgCl2 reacts violently with ZrF4 to form ZrCl4.

Altwicker¹⁶ describes Sauerwald's work in 1938-41 with K₂ZrF₆, ZrCl₄ + KCl, ZrF₄, and ZrO₂. Emley^{17,18} claims the use of a double fluoride salt with the empirical formula KZrF₅. Altwicker¹⁹ in 1949 states that in Mg-Zn-Zr systems microporosity increases with increasing Zn content, that substituting Cd for Zn re-

sults in no microporosity problem, and that cerium additions increase the pressure tightness of the Mg-Zn-Zr alloys. Emley's paper²⁰ in 1949 states that his experience with Mg-Zr alloys indicates that alloys made with ZrCl₄ or alkali metal chlorozirconates are inherently unreliable from a viewpoint of chloride contamination.

Swedish work²¹ claims a master alloy of 13 per cent Zr-86 per cent Zn as having an excellent alloying efficiency (ratio of zirconium found in the product to the zirconium added to the melt), and that iron will precipitate about four parts by weight of zirconium to form an insoluble product which sinks to the bottom of the melt. Sauerwald²² explains that zinc increases the solubility of zirconium in magnesium by 0.1–0.2 per cent, and states that 1.6–2 per cent cerium is needed to produce sound Mg + 4–6 per cent Zn + Zr alloys.

Ball and Jessup²³ teach that Ba and Sr in Mg–Zr alloys increase the burning tendency and yield strength and decrease the grain size of the alloy. Ball and Fox²⁴ indicate the use of lumps of ZrCl₄ + KCl or ZrCl₄ + NaCl and 45 ZrCl₄ + 35 ZrF₄ + 20 NaCl. Wilkenson²⁵ compares properties of Mg + 5 Zn + Zr, Mg + 3Zn + 3 RE (rare earth metals) + Zr, and Mg + 3 RE + Zr alloys with Mg + 8 Al and Mg + 9 Al + 1 Zn and states that the Mg + 5Zn + Zr alloy is less prone to microporosity, while the Mg + RE + Zr alloys (with or without Zn) are completely free of microporosity.

In 1950 Murphy et al.²⁶ suggest a method of alloying beryllium and zirconium and cerium with the use of silver-beryllium and silver-zirconium master alloys. Ball et al.²⁷ claim that NaF causes Mg-Zr alloys to burn more readily, and also claim the use of ZrF₄ + fluorides of lithium, strontium, calcium and magnesium. They further claim²⁸ a master alloy made from fluorides of zirconium, and state that traces of lithium, strontium, calcium, and barium may be removed from the melt with MgCl₂.

Ball et al.²⁹ claim a master alloy made by the reduction of  $\rm ZrCl_4$  with Mg in a flux made heavy with 30 per cent or more  $\rm BaCl_2$ . They also teach that rare earth metals reduce the porosity in Mg + Zn + Zr alloys at the expense of embrittling the alloys, that the porosity comes from hydrogen in the melt or by the reaction of the more electropositive metals with moisture in the molding sands, and that the microporosity is lessened if the zirconium in solution exceeds 0.6 per cent.

Emley et al.³⁰ describe an agent of ZrF₄ + alkaline earth metal chlorides and KCl and/or NaCl and free from NaF or KF. Emley and Jessup^{31,32} indicate that Mg-Zr alloys have a strong tendency toward flux inclusions, which is diminished by treating in inspissated or fluid fluxes made heavy with large amounts of Ba or Sr salts. Emley³³ states that MgCl₂ and MgF₂ will react with the rare earth metals in magnesium alloys to lower the rare earth metal content, but if MgF₂ is wet with certain chloride salts it will be inactive to the rare earth metals. Thus MgF₂ can be used to inspissate fluxes which are free from MgCl₂ for treating Mg-RE alloys.

Jessup and Wilson³⁴ report that the addition of

beryllium to Mg-Zr alloys causes coarse grain, but if a reducible halide of beryllium is added in specified small amounts along with the zirconium the problem is eliminated. Meier and Martinson35 discuss Canadian experience with Mg + 6 Zn + Zr alloys. Emley³⁶ treats the electropotential series of certain chlorides and fluorides and shows that chlorides of sodium and potassium are not easily reduced by magnesium, but that fluorides of these metals are readily reduced by magnesium. He also states that ZrF4 is insoluble in alkali metal chlorides, that the sulphide of zirconium is not reduced by magnesium, that the nitride is reduced by magnesium only to ZrN, that phosphorus and carbon are also zirconium alloying inhibitors, and that while manganese is an alloying inhibitor the precipitated material is of such a size that settling does not occur readily.

Nelson and Strieter in their 1950 paper³⁷ discuss the Mg-rare earth metal-Zr alloys and also state that MgCl₂ reduces the rare earth metal contents, Orton³⁸ prefers a mixture of ZrCl₄ and alkali metal chlorides to Na₂ZrF₆ and a master alloy and shows properties of Mg + 4Zn + Zr alloys. Ball³⁹ further discusses properties of the alloys. Van Ewijk⁴⁰ limits Ca to <0.01 per cent in casting alloys of Mg-Zr. Ball et al.⁴¹ state that bubbling ZrCl₄ vapor into magnesium has not given satisfactory results, and that ZrO₂, if highly reactive, can be reduced by magnesium.

Ball and Jessup⁴² report that barium and strontium are helpful in eliminating flux inclusions, and Ball et al.⁴³ state that metallic zirconium has not given satisfactory results, that hydroxide groups with zirconium fluorides give difficulty in alloying, and that NaF₂ should be water-leached out of alloying agents. Ball et al.⁴⁴ claim that Na imparts poor properties to the alloys. Emley⁴⁵ reports on the use of K₂ZrF₆ + KCl.

Jessup and Wilson⁴⁶ claim that in co-alloying beryllium and zirconium as  $ZrCl_4 + BeF_2$  the MgCl₂ formed will dissolve the MgF₂. Ball et al.⁴⁷ teach that combined water has a strong inhibiting effect on Zr alloying, and that MgO should be absent during alloying. Emley and Jessup⁴⁸ state that traces of barium and strontium are harmful in Mg–Zn–Zr casting alloys, and that the dark spots sometimes seen on the surface of the Mg–Zn–Zr alloys are not related to corrosion resistance but are only surface discolorations.

Nelson and Strieter in their 1951 paper⁴⁹ describe castability and properties of Mg–RE–Zn–Zr and Mg–RE–Zr alloys, and report the use of zirconium sponge as an alloying agent. Emley⁵⁰ reports that zirconium precipitates phosphorus as an insoluble compound, and that P is <0.005 per cent in commercial Mg–Zr alloys. McDonald⁵¹ describes the use of a master alloy prepared by the reduction of ZrCl₄ with Mg in alloying operations.

# Bibliography

- Helmut von Zeppelin, I.G. Mg Reports, Microfilm BB209/ 4627-4636, 24 Aug., 1938.
- 2. Mag. Electron Ltd., Brit. Pat. 511,137, 1939.
- 3. Compagnie de Produits Chimiques et Electrometallurgiques Alais, Froges et Camargue, Brit. Pat. 511,791, 1939.
- 4. H. von Zeppelin and T. Duisberg, U.S.P. No. 2,214,211, 1940.
- 5. Franz Sauerwald, U.S.P. No. 2,228,781, Jan., 1941.

- 6. Mag. Electron Ltd., Brit. Pat. 533,264, Feb., 1941.
- 7. Sauerwald et al., U.S.P. No. Applic. 369,749, (Applic. 6-15-43).
- 8. Sauerwald et al., U.S.P. No. Applic. 369,746, (Applic. 6-15-43).
- 9. Sauerwald et al., U.S.P. No. Applic. 369,824, (Applic. 6-15-43).
- 10. P. T. Stroup and G. F. Sager, U.S.P. No. 2,373,515, April, 1945.
- 11. P. T. Stroup and G. F. Sager, U.S.P. No. 2,373,516, April,
- J. P. Doan and G. Ansel, AIME Transactions, 171, 1946,
   p. 286.
- 13. Franz Sauerwald, Z. anorg Chemie, 225, 1947, p. 212-220.
- C. J. P. Ball, Metallurgia, 35, 1947, p. 125-129; 211.
   E. F. Emley, Disc. of the Faraday Soc., 47, no. 4, 1948-49,
- 16. H. Altwicker, ATI, No. 27205, Hg AMC, Aug. 1948.
- 17. E. F. Emley, U.S.P. No. 2,452,914, Nov. 1948.
- 18. E. F. Emley, U.S.P. No. 2,452,894, Nov., 1948.
- 19. H. Altwicker, ATI, No. 22439, Hg AMC, Jan., 1949.
- 20. E. F. Emley, Journal Inst. Met., 75, pt. 6, Feb., 1949.
- 21. A. Johnson & Co., Brit. Applic. 1576/1949.
- F. Sauerwald, Z.f. Metallkunde, 40, Feb., 1949, p. 41-46.
   C. J. P. Ball and A. C. Jessup, U.S.P. No. 2,464,918, March,
- 24. C. J. P. Ball and F. A. Fox, Brit. Pat. 624,304, June, 1949.
- 25. R. G. Wilkinson, Metallurgia, 41, no. 242, Dec., 1949.
- 26. Murphy et al., U.S.P. No. 2,461,229, Feb., 1950.
- 27. Ball et al., U.S.P. No. 2,497,529, Feb., 1950.
- 28. Ball et al., U.S.P. No. 2,497,530, Feb., 1950.
- 29. Ball et al., U.S.P. No. 2,497,531, Feb., 1950.
- 30. Emley et al., U.S.P. No. 2,497,537, Feb., 1950.
- 31. E. F. Emley and A. C. Jessup, U.S.P. No. 2,497,538, Feb., 1950.
- 32. E. F. Emley and A. C. Jessup, U.S.P. No. 2,497,539, Feb., 1950.
- 33. E. F. Emley, U.S.P. No. 2,497,540, Feb., 1950.
- 34. A. C. Jessup and J. B. Wilson, U.S.P. No. 2,497,551, Feb., 1950.
- 35. J. W. Meier and M. W. Martinson, A.F.S. Transactions, vol. 58 (1950).
- 36. E. F. Emley, Inst. of Min. & Met., Refining of Non-Ferrous Metals Symposium, July, 1949.
- 37. K. E. Nelson and F. P. Strieter, A.F.S. TRANSACTIONS, vol. 58 (1950).
- 38. G. W. Orton, A. F. Tech. Report 6173, Hg AMC, June, 1950.
- 39. C. J. P. Ball, Magnesium, Nov., 1950.
- L. J. G. Van Ewijk, Overdukuit Congresnummer, Metalen, Dec. 1, 1950.
- 41. Ball et al., Brit. Pat. 657,222, April, 1951.
- 42. C. J. P. Ball and A. C. Jessup, Brit. Pat. 652,223, April, 1951.
- 43. Ball et al., Brit. Pat. 652,224, April, 1951.
- 44. Ball et al., Brit. Pat. 652,226, April, 1951.
- 45. E. F. Emley, Brit. Pat. 652,227, April, 1951.
- 46. A. C. Jessup and J. B. Wilson: Brit. Pat. 652,229, April, 1951.
  - 47. Ball et al, Brit. Pat. 652,230, April, 1951.
- 48. E. F. Emley and A. C. Jessup, Brit. Pat. 652,234, April, 1951. 49. K. E. Nelson and F. P. Strieter, A.F.S. TRANSACTIONS, vol.
- 59, 1951.
- Emley et al, Journal, Inst. Met., 80, Sept., 1951, p. 23.
   J. C. McDonald, Light Metal Age, 9, Nos. 11, 12, Dec.,
- 52. C. E. Nelson, Transactions, AIME, 159, 1944, p. 392.
- 53. Dow Chemical Co., Analytical Method, MLW.51.13, May 31, 1951.
- 54. Dow Chemical Co., Analytical Method, MLW.51.5, June 11, 1951.
- 55. J. H. deBoer, Foote Prints, III, No. 2, 1930.
- 56. Kroll et al., Transactions, Elect. Chem. Soc., 89, 1946, p. 233.
- Kroll et al., *Transactions*, Elect. Chem. Soc., 92, 1947, p. 99.
   E. A. Gulbranson and K. F. Andrews, *Journal*, Elect. Chem. Soc., 96 (6), 1949, p. 364.
- 59. J. F. Hildebrand and F. P. Strieter, A.F.S. Transactions, vol. 60 (1952).

#### DISCUSSION

Chairman: A. Cristello, American Light Alloys, Inc., Little Falls, N. J.

Co-Chairman: H. E. Elliott, Dow Chemical Co., Bay City, Mich.

Recorder: A. CRISTELLO.

E. F. EMLEY (Written Discussion): This interesting paper sets out clearly the comparative amounts of zirconium introduced solubly into magnesium using various alloying agents. The zirconium alloying field is quite a large one since a considerable number of salt mixtures have now been mentioned in the patent literature. Examples of most types have been tested by the authors. Anyone who has done research work on alloying zirconium with magnesium will know that there are some snags which can arise and the whole business is not entirely straightforward. The paper raises some general matters which are of importance and call for discussion, and there are many points of detail on which comments might be made.

1. The authors do not claim that their work enables a decision to be made as to which is the best alloying method for zirconium. They rightly stress that economic factors must be taken into consideration. A further weeding out of the processes on technical grounds is however required. No mention is made of the tensile properties or tendency to microporosity shown by alloys produced with various reagents. Such information is vital where ZK51A sand castings are to be made. Quite a few of the alloying mixtures tested would produce ZK51A alloy showing pronounced tendency to microporosity and unacceptably low tensile properties, even with a full zirconium content.

2. The authors dismiss the question of flux contamination on the grounds that since the development of proper handling techniques over half a million pounds of zirconium alloy have been made and that this has shown no flux problem to exist with the techniques used. I think I am correct in believing that the material referred to is ZK60 alloy extrusion with a minor

proportion of EK30 sand castings.

Now it is a relatively simple matter to produce acceptable zirconium alloy extrusions, since not only can metal be gently siphoned or decanted from the alloying crucible, but the subsequent extrusion process breaks up and disperses the characteristic primary chloride contamination so that only quite gross inclusions remain detectable after working. I would like to show in accompanying illustrations EK31 castings made with a proprietary American brand of ZrCl4 and KV1, American electrolytic magnesium, and the standard British Melrasal fluxes. (Figs. 1A-2A) These castings were made by one of the leading British sand casting foundries who have themselves already handled over a million pounds of zirconium alloys. It will be

seen that the chloride contamination after machining and exposure for 48 hr to moist air is quite marked. This chloride contamination arises from entrapment of small amounts of the chloride alloying reaction product in clusters of insoluble zirconium-rich particles, and it differs fundamentally from all chloride contamination in non-zirconium alloys in that it is persistent and cumulative, and is not removable on remelting or fluxing. As far back as June 1943 Magnesium Elektron Ltd. made several tons of ZK31 alloy via chlorozirconates, but it was decided that the chloride contamination, though not gross, was such as to render the metal unacceptable by British A.I.D. standards, and accordingly other alloying methods were investigated. We found, however, that, by incorporating about 15 per cent BaCl, in the chlorozirconate salt mixture used for alloying, primary chloride contamination can be considerably reduced (see Ref. 29), and with over 30 per cent BaCla it can be entirely prevented. (Fluxes rich in BaCla are also required to obtain zirconium alloys free from chlorides).

Nearly all of the chloride contamination appearing in Figs. 1 and 2 is too light to be seen except on a machined surface after exposure to damp air, and is probably in itself of no consequence from any point of view. Unfortunately a few more serious inclusions are present, Figs. 1 and 2 each showing one such. These castings are in my experience typical of what can be achieved using ZrCl₄-KCl mixtures and normal fluxes without additions of BaCl₂. Examination of many such castings has shown that it is only necessary to expose a sufficiently large surface area in order to find one or more really objectionable inclusions, and that the best way to avoid these is to employ fluxes and alloying methods which do not introduce even light contamination into

the final alloy.

A further point about this chloride contamination is that, where insufficient BaCl2 is used to ensure absence of chloride contamination, ZK51 castings may show objectionable contamination and EK31 castings very appreciable contamination under conditions where EK30 castings are quite acceptable: a full zirconium content makes it more difficult for any chloride contaminated particles to settle through the melt; and moreover EK31 alloy possesses acceptable properties at 0.5-0.6 per cent zirconium which permits the metal after alloying to be transferred away from the fluid salt residue into a separate crucible where it may be further refined before pouring castings. Such transfers are impracticable with ZK51, since if the melt is not saturated with zirconium at the time of pouring the castings may show microporosity, hot cracking and unacceptable tensile properties. The object of these remarks is to warn the authors against dismissing the chloride contamination problem as of no account merely on the results of extensive production experience of ZK60 and EK30. I make this point because I feel that the question of chloride contamination has not been given the attention on this side of the Atlantic that it deserves.

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Fig. 1A—EK31 alloy sand casting machined and exposed to 80 pct relative humidity for 48 hr. (Alloy made with proprietary salt mixture of ZCl, and KCl, using standard British fluxes. Nominal Zr addition, 2.4 pct. Approx. ½ actual size.

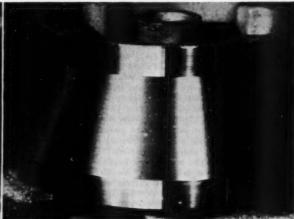


Fig. 2A—EK30 alloy sand casting machined and exposed to 80 pct relative humidity for 48 hr. (Alloy made with proprietary salt mixture of ZCl₄ and KCl using standard British fluxes.

Nominal Zr addition, 1 pct. Approx. 1/3 actual size.

In both the present paper and that on ZK51A a modified flux containing 20 per cent BaCl₂ is mentioned. We found that 20 per cent BaCl₂ is about the minimum BaCl₂ content which will prevent appreciable dispersion of flux in the alloy (see Ref. 31) but that at least 30 per cent BaCl₂ is desirable. We would not use or recommend a flux containing less than 30 per cent BaCl₂ for alloys with full zirconium content, eg. ZK51A.

3. The conclusions reached by the authors on alloying via fluozirconates are quite misleading. No information appears to be given about the proportions of the other salts added to KZrFs (Table 3) but with almost any proportions there should not have been difficulty in consistently achieving saturation with a 3 per cent addition of zirconium. Had the authors investigated the effect of alloying temperature on zirconium introduction, as they did with other mixtures, they would have found that at 1500 F these mixtures alloy quite satisfactorily. The authors are right to call attention to a barium introduced with the zirconium, and they indicate how this has arisen. Nothing is said as to how entry of barium might be presented. Small amounts of barium damage tensile properties in the sand-cast state, and in any commercial process involving fluozirconates it is essential to arrange that the alloying salt mixture used shall not introduce barium. None of the salt mixtures indicated in Table 3 would give alloys with acceptable tensile properties in sand castings, even had the full zirconium content been introduced. Again the salt mixtures of Table 5 would be useless for making alloys for sand casting.

4. Violent reactions between magnesium and K₂ZrF₆ with small KCl additions are promoted mainly by partial reduction of KF with liberation of potassium vapor. With sufficient KCl to suppress reduction of KF, it is quite safe to pour molten magnesium on to K₂ZrF₆-KCl mixtures. We made a 1-ton melt of ZK31 alloy by this process in 1945. With little or no KCl present the reaction is violent even if the salts are poured into the

magnesium.

5. ZrF₄ without other additions can be alloyed much more efficiently than the authors' results suggest. For instance with a 4 per cent zirconium addition we have introduced 0.6-0.75 per cent into pure magnesium. It appears that the authors have stirred the powder into the metal using much flux to prevent burning, and it seems likely that they may have wetted the ZrF₄ particles with flux and so partially blanketed the ZrF₄ from contact with the magnesium. In such a case it would seem better either to pour the magnesium on to the fluoride, so preventing wetting by flux at the outset, or to dissolve the ZrF₄ first in some suitable halide mixture. The alloying properties of ZrF₄ however depend to some extent on its physical state and method of preparation, and it may be that the authors have not used the most favorable material in their tests.

6. The information given about the K30A zirconium master alloy is of interest to us since we have had a good deal of experience with master alloys of this type. Originally in 1944 we made master alloys by pouring magnesium saturated with zirconium on to alkali metal chlorozirconates, stirring to effect reaction, pouring off the excess metal and salt reaction product, and expressing the residual chlorides as far as possible. (For method used see Ref. 29). In 1946 we made half a ton of master alloy from salt mixtures containing ZrCl, alkali metal chlorides and BaCl₂, the BaCl₂ being added to avoid introducing chloride contamination into casting from the residual chloride in the master alloy (about 7 per cent). We found that with over about 30 per cent BaCl, in the salt mixture, residual contamination in the castings was suppressed, and that with 15 per cent BaCl₂ it was considerably reduced. This master alloy contained about 30 per cent zirconium. It was used to make prototype zirconium alloy sand castings in British foundries. Despite many attractive features we decided to abandon this master alloy in favor of a master salt process. The reasons were partly technical and partly economic. With these master alloys we obtained results which agree closely with those of the authors shown in Fig. 10, and which incidentally are similar to those achieved by the master salt process in which a 2 per cent addition of zirconium will introduce 0.6 per cent zirconium solubly into magnesium. (The latter process has now been in commercial operation for over five years, and well over three million pounds of master salt have been used in foundries in many countries, including the

It would be interesting to hear from the authors whether the master alloy they describe would differ significantly from the type we have used. What for example is the chloride content of the K30A master alloy, and how much BaCl₂ does it contain? Although detailed economics of zirconium alloying probably lie rather beyond the intended scope of the present paper, it would be interesting if some indication could be given of the probable price for the K30A master alloy, and the cost of using it to make zirconium alloys, e.g. in relation to that of alloying with proprietary salt mixtures containing ZrCl₄. One would assume that if the master alloy is made from the latter it would be appreciably more expensive in use, as we ourselves found.

7. The authors query a published statement that ZrF₄ is insoluble in KCl. When we added only a very small quantity of ZrF₄ to molten KCL, the clear melt became cloudy; and on adding an equal weight of ZrF₄ the melt gave a distinct KCl pattern when examined by X-ray diffraction (Fig. 3A). The remaining lines do not necessarily indicate a compound of ZrF₄ and KCl: they may be due to some decomposition product of ZrF₄

of which there are a considerable number.

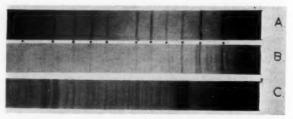


Fig. 3A—X-ray diffraction pattern of a prefused melt of  $ZrF_i$  and KCl(B) showing presence of KCl(A). Top (A),  $KCl_i^*$  Center (B) fused  $KCl-ZrF_i$  (50/50 by wght.) and Bottom (C)  $ZrF_i$  (Monoclinic).

With ZrF₄·KCl mixtures the authors have introduced considerable amounts of zirconium into magnesium. If the fluxes used contained MgCl₂ and perhaps BaCl₂, I would expect this, since these salts will chlorinate ZrF₄ to ZrCl₄ which alloys readily. To test ZrF₄·KCl mixtures *per se* it would be necessary to use a NaCl-KCl flux.

8. The excellent alloying results obtained with Kroll process zirconium sponge and with pure zirconium powder are of great interest, as they show quite definitely that, when sufficiently pure, zirconium metal can be alloyed with magnesium up to the highest levels. We have never managed to incorporate more than 0.55 per cent zirconium in true solution in magnesium using any metallic form of zirconium, and had been inclined to wonder whether it was possible to exceed the peritectic concentration in this way. On the other hand, to what do the authors attribute the poor alloying efficiency of the raw (unsublimed) sponge containing MgCl₂, both as received and after leaching? And might the same condition sometimes affect the K30A hardener which also contains some MgCl₂?

The Appendix is of considerable value as an index to the published literature on the subject. I think however that for the purposes of tracing the historical development it would have been better if the patents, which form a considerable proportion of the references, had been mentioned in the order of their filing dates rather than their publication dates. This applies mainly to British patents; most of those forming references 41-48 and published in April, 1951, were actually filed in 1945. A remark in the appendix that Ball states that sinks rather than microporosity form in poorly fed sections of ZK51 requires cor-

rection: the remark related to EK31 alloy.

Finally, may I ask some questions on analytical methods: Why was sulphuric acid used to distinguish between metallographically soluble and insoluble zirconium? Do the high analyzed soluble zirconium contents obtained on alloying with sponge in comparison with ZrCl₄ correspond with metallographical reality, or do they result from the presence of larger proportions of undissolved zirconium which are appreciably attacked during dissolution of the sample? How much "soluble zirconium" would this analytical method indicate if applied to K30A hardener containing, say, 50 per cent zirconium? The paper refers to some exceedingly small barium contents. Can the authors please say how these were determined? And what reagent is used for the colorimetric determination of zirconium? (We use Alizarin S.)

MESSRS. SAUNDERS AND STRIETER (Written Reply to Mr. Emley): Dr. Emley states that the paper does not enable a decision to be made as to which is the best alloying method for zirconium. The conclusions point out that each foundry must, because of its individuality, choose its own method. The paper rather clearly shows that the master alloy described in the text is a particularly

good alloying agent.

Dr. Emley mentions that "quite a few of the alloying mixtures tested would produce ZK51 showing pronounced tendency to microporosity and unacceptably low tensile properties, even with a full zirconium content." It is unfortunate that Dr. Emley neither presents data nor explains to which mixtures he refers. If he refers to the fluozirconate or "inert fluorides" which furnish certain of the more electropositive metal contamination, (i.e., Li, K, Ba, etc.) we should tend to agree with him. On the other hand, if the remarks are aimed at the chlorozirconate type of agent, we can state that our properties, as shown in the Hildebrand and Strieter⁵⁰ paper, are equal to those properties reported by both Ball14 and Wilkinson.25

Dr. Emley is concerned that American attention to flux contamination has not been rigid. Dr. Emley's past work has established a noteworthy reputation in laboratory phases of work related to flux contamination and he has made definite contributions to the art. One must, however, recognize that a boundary exists between research techniques and the practical considerations of producing a salable item. Certainly we would agree that Dr. Emley's photos of EK31 castings appear to be badly contaminated. Castings of this quality would indicate unusually poor foundry practice. It certainly is not typical of the hundreds of thousands of pounds of rare earth metal and zirconium containing magnesium castings which have been satisfactorily pro-

duced in this country for jet aircraft engine parts.

The mixtures shown in Table 5 are claimed by Dr. Emley to be "useless" for making sand castings. The authors are uncertain as to the basis for this blanket statement, but assume that it is based on the fear that contamination by such impurities as Na, K, Be, etc. (which seems to be inherent with fluozirconate-type agents), would cause lowered tensile properties. Actually, the results achieved by using some of these mixtures were quite interesting. For example, the first fluoride mixture shown in Table 5 has resulted in ZK51A alloy which, although containing as little as 0.52 per cent soluble Zr and 0.004 per cent Na, displayed the following tensile properties in separately sand-cast test bars tested in the T6 condition: 7% El, 27,000 psi yield strength, and 42,000 psi ultimate strength. Certainly these properties would not support a conclusion that this alloying material would have no utility.

The master alloy K30A contains about 7 per cent chlorides with little or no BaCl2 present. Our experience has shown that it is not necessary to add BaCl2 to the master alloy to achieve satisfactory freedom from flux contamination. It is currently slightly more expensive per pound of contained zirconium than, say, ZrCl4-KCl salts; however, the final cost can be measured only when all economic and technical aspects of the process are considered. Thus, although the master alloy initially costs more than chlorozirconates and some fluozirconates, rare earth metals are not lost in alloying, and the convenience of using a hardener rather than large quantities of salts more than counterbalances the initial price differential.

The X-ray diffraction patterns shown by Dr. Emley are interesting. The ZrF4 pattern agrees with our ZrF4 standards but contains, in addition, what seems to be another phase.

The ZrF4-KCl pattern is different from the pattern we obtain when fusing equal parts of KCl and ZrF4 Dr. Emley's pattern contains KCl and a phase which we would call K2ZrF6 · KCl. Our melts show KCl and a different phase which may be isomorphous and larger than the K2ZrF6 · KCl.

An answer to the question about the poor alloying efficiency of the raw (unsublimed) zirconium sponge containing MgCl₂ might be found in the reference of Ball, Jessup, Emley, and Fisher⁴⁷ which states that combined H₂O has a strong inhibiting effect on Zr alloying: that is, the hygroscopic MgCl₂ phase (which constituted about 50 per cent of the agent) probably picked-up some H2O and the H2O did inhibit the final alloying. Since the MgCl2 content of K30A is less than 10 per cent of that in the raw sponge mixture, there is a much lower possibility of absorbing H2O. In addition, practical experience, as well as laboratory test, has shown that there is no problem relative to moisture absorption when the hardener is properly handled.

We are grateful for Dr. Emley's correction of the statement of the Appendix which quotes Ball14 as stating that sinks are found in poorly-fed sections of Mg+5Zn+Zr alloy (ZK51A) rather than microporosity. As Dr. Emley points out, the statement is related to Mg+3rare earth+Zr alloy (EK31).

With reference to the analytical method question about the use of H₂SO₄, we have tested HCl, HNO₂, and H₂SO₄ to distinguish between soluble and insoluble zirconium. No differences were detected and all three acids yielded the same results.

It is doubted that higher amounts of undissolved zirconium exist in alloy made from zirconium sponge. One might actually expect more undissolved Zr in any Zr salt-produced alloy because there should be smaller particles of zirconium in the latter melts and these smaller particles would settle more slowly. Further, if the solute acid has a tendency to dissolve Zr particles, the smaller particles should, of course, be expected to be more appreciably attacked during dissolution.

Barium contents were determined by spectrochemical techniques. The alloy sample was dissolved and loaded on purified graphite electrodes and sparked in an Enns-type spark source in

a 3 meter grating spectrograph.

Our colorimetric analysis technique uses Alizarin red S as a reagent.54

J. W. MEIER (Written Discussion): 2 This paper has to be considered as a major contribution to published information on the growing family of zirconium-containing magnesium alloys and the authors are to be congratulated for their fine work. As reported earlier²⁵ we conducted a similar investigation of various zirconium-containing alloying agents and our findings confirm, in general, the results obtained in this paper, with the exception that we were unable to obtain highest "effective" zirconium contents using pure zirconium. We found that the most consistent and most effective alloying mixtures are those based on or, at least, containing some zirconium chloride compounds. We believe that the exceptionally high properties obtained using zirconium chloride compounds are due to the efficient refining and degassing action of these fluxes. Especially important seems to be the removal of hydrogen which affects greatly the solubility of zirconium in magnesium and magnesium alloys.

It is interesting to note that attempts at using proprietary alloying methods based on various zirconium fluoride compounds in the production of the high strength magnesium casting alloy ZK61 were rather unsuccessful. Fully heat treated test bars cast from fluoride-treated melts show consistently lower tensile properties than similar test bars from chloride-treated melts and these differences are quite substantial, i.e. 5-6,000 psi UTS and

2-3,000 psi 0.2% PS.

Recent experiments with a master alloy similar to that designated by the authors as K30A, showed very good results and it can be recommended in all cases where the addition of large amounts of salts has to be avoided, although the two-step melt-

ing operation may be more costly.

Another very interesting master alloy technique, not mentioned in the paper, proposed by S. C. Pyk in U.S. Patent 2,558,-627 (1951), consists of using various zinc-zirconium or zinczirconium-magnesium master alloys produced by an electrolytic process whereby the proportion of these metals can be altered by high-vacuum treatments at various temperatures (partial distillation of zinc). The practical application and efficiency of this method has to be proven.

MESSRS. SAUNDERS AND STRIETER (Written Reply to Mr. Meier): We are grateful for Mr. Meier's addition to the U.S. Patent 2,558,627. We are also pleased that Mr. Meier's experience would lead him to essentially agree with our work related specifically

to the master alloy.

It is doubted that the ZrCl, compounds would actually remove hydrogen, but we would agree that the resulting MgCl₂ salts should be efficient refining materials.

It is possible that the lower physical properties for alloy prepared from ZrF4 compounds might be related to contamination by some of the more electropositive metals introduced via the fluoride decomposition by magnesium.

C. E. NELSON: 3 It should be kept in mind that the purpose of this paper was to present, for all to see, a general survey of

² Canadian Department of Mines & Technical Surveys, Ottawa, Ontario,

² Dow Chemical Co., Midland, Mich.

some of the various chemical possibilitites and methods of alloying zirconium with magnesium. The data show that full soluble or effective zirconium contents can be obtained by many methods and there was no intent in the paper to either promote or discredit any particular proprietary method.

and there was no intent in the paper to either promote or discredit any particular proprietary method.

The discussion of this paper has mentioned proprietary methods quite extensively and I would just like to say that we are certain that the fluoride alloying methods used in England are good and adequate methods, but furthermore, the entire range of zirconium alloys, with or without rare-earths or zinc, have

been and are being made on large commercial scale in the United States, using the Mg-Zr hardener derived from ZrCl₄ reduction with magnesium or using ZrCl₄ base materials. There is no problem in getting the full effective Zr content and, contrary to implications in some of the discussion, there is no problem in making high quality castings free from harmful inclusions.

It would appear that the actual choice of the specific alloying method rests primarily on economics of the over-all process and this may conceivably be different in various countries.

# MECHANICAL PROPERTIES OF SAND CAST MAGNESIUM-ZINC-ZIRCONIUM ALLOYS

By

J. F. Hildebrand* and F. P. Strieter*

#### Introduction

During the past decade the increasing use of magnesium has been further extended by the appearance of new and better alloys. The field of application for these new alloys is ever broadening.

The recent papers by Leontis, 1.2 Nelson and Strieter, 3.4 and their predecessors illustrate the improved elevated temperature properties obtained on magnesium-rare earth metal-zirconium alloys. Aircraft engine manufacturers have been making increased use of these alloys.

New to the list of potential future magnesium sand casting alloys is ZK51A, also referred to by British investigators as Z5Z. ZK51A is the proposed ASTM designation for a magnesium alloy containing about 5 per cent zinc and 0.7 per cent zirconium. This alloy, while showing promise of having interesting mechanical properties, has not been proven in service to the same extent as the magnesium-rare earth metal-zirconium alloys. It presents a number of foundry castability problems, such as slight porosity and surface defects, which will need to be solved before the alloy can take an accepted place among standard magnesium casting alloys.

The initial work on this alloy system was done in Germany⁵ where the alloying of zirconium to magnesium was intensively studied in the years just prior to World War II. This early work was expanded upon by investigators in Great Britain⁶ where a magnesium-5 per cent zirc-0.6 per cent zirconium composition was developed as a sand casting alloy.

Coincidently with this development was that of a magnesium-6 per cent zinc-0.6 per cent zirconium composition which was exploited in the United States⁷ for extrusions. More recently considerable discussion and experimental development has taken place in Canada and the United States on the sand casting of Mg-Zn-Zr alloys. Meier and Martinson⁸ have shown the tensile properties which can be obtained in a Mg-6 per cent Zn-0.7 per cent Zr alloy in the heat treated and aged condition. It is the intent of this paper to present some of the mechanical properties

which can be achieved in sand castings of Mg-5 per cent Zn-0.7 per cent Zr alloys following a simple aging treatment.

The tests show that ZK51A alloy, when aged at 350 F for 12 hr, exhibits a good combination of static and dynamic mechanical properties. In most respects the properties of this alloy are similar to the present commercial alloys; AZ91C, AZ92A, and AZ 63A. The new alloy does have better ductility at a high level of yield strength. This may be an indication that castings in the new alloy will sustain a higher static load than the present alloys.

#### **Procedures**

The melts required for this work were made up using 99.8 per cent purity electrolytic cell magnesium and 99.99 per cent purity zinc. The zirconium was obtained from two different sources. The zirconium in some of the melts came from a fused salt containing about 50 per cent ZrCl₄ and 50 per cent alkali metal chlorides. The procedure for its use was described by Doan and Ansel.⁷ The balance of the melts were alloyed with a laboratory produced Mg-Zr hardener currently described by Saunders and Strieter.⁹ The melting, handling and pouring procedures followed the crucible practice described by Nelson, ¹⁰ except that a flux containing 20 per cent BaCl₂ was used.

Sand cast test bars and panels were poured from 20- or 60-lb melts. The molds were made of regular magnesium foundry sand containing sulphur and boric acid inhibitors. The test bars were standard cast to shape ½-in, diameter reduced section tensile bars. The panels, which were machined to the desired dimensions for fatigue tests, were 8 x 4 x 5/16 in. Special panels 7 x 2 x 7/32 in. were cast for relaxation specimens.

Radiographs' of the test bars and panels showed that they were substantially free of porosity, segregation and oxide skins.

Separately cast samples, poured immediately before the test bars or panels, were used for chemical and spectrographic analyses. Except for the expected decrease in iron content, the analysis of impurity elements revealed little, if any, variation from that of

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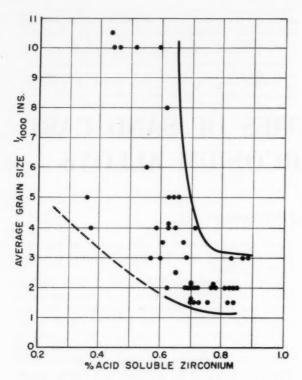


Fig. 1—Effect of zirconium content on grain size of ZK51 alloy.

the cell magnesium. The iron content was reduced by the settling action of the zirconium as described by Beck¹¹ and by Stroup and Sager.¹²

A chemical analysis was used to establish the zinc and zirconium contents of the alloys. Both acid soluble and acid insoluble zirconium analyses were made on several melts, but the acid soluble zirconium obtained on all alloys was considered indicative of the zirconium content. The zirconium contents were obtained by either of two different chemical methods—gravimetric13 and colorimetric.14

#### **Experimental Work**

Before obtaining the mechanical properties of ZK-51A alloy, some preliminary studies were made to establish the desired composition range and aging treatments.

The dependence of the grain refinement of magnesium-zirconium alloys on the amount of zirconium in solution with the magnesium was reported by Sauerwald.⁵ Meier and Martinson⁸ also showed the importance of the zirconium content on the grain size and mechanical properties of ZK61 alloy (a magnesium alloy containing 6 per cent zinc and over 0.7 per cent zirconium). Their tensile data, however, are correlated to the total zirconium content.

The effect of the acid soluble zirconium content on the grain size of the test bars of individual Mg-5 per cent Zn-Zr alloys is shown in Fig. 1. Several melts with intentionally low zirconium were made to determine what grain size might be anticipated. As the curve shows, there is a wide range of grain size when the soluble zirconium content falls below 0.7 per cent, and only when the soluble zirconium content exceeds this value is there assurance of obtaining a fine grain alloy.

The importance of the zirconium content on the tensile properties was reflected in the various heat treatments tried on magnesium-zinc-zirconium alloys. Sauerwald⁵ suggested tempering Mg-2 per cent Cd-2 per cent Zn-Zr alloys at 250 C (482 F) for 15 hr to increase the yield strength 50 per cent while retaining a 10 per cent elongation. The German ZB alloy (Mg-3 per cent Zn-0.5 per cent Zr) was annealed at 280 C (536 F) for 5 hours according to Beck.¹¹ Ball⁶ has stated that Z5Z alloy (Mg-4.5 per cent Zn-0.7 per cent Zr) should be aged at 180 C (356 F) for 24 hr in order to obtain the best properties on sand cast material. More recently Ball¹⁵ recommends heat treating 2 hr at 330 C (625 F) and then aging 24 hr at 180 C (356 F) in order to relieve casting stresses.

Doan and Ansel⁷ surveyed heat treatment temperatures up to 950 F and aging temperatures from 200-400 F on extruded ZK60A alloys (ASTM designation for a magnesium wrought alloy containing nominally 5.5 per cent Zn and 0.6 per cent Zr but a specified minimum of 0.45 per cent Zr). They stated that the high temperature treatments were detrimental to the tensile properties. Leontis² used a two-stage solution heat treatment on cast Mg-Zn-Zr alloys consisting of a 2-hr treatment at 675 F followed by a 14-hr treatment at 725 F. These alloys were subsequently aged at 350 F for 16 hr.

More recently Meier and Martinson⁸ reported the results of an extensive program on the heat treatment and aging cycles for cast ZK61 alloy. They concluded that the best properties were obtained after

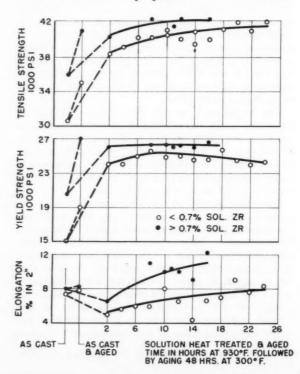


Fig. 2—Effect of time at solution heat treatment temperature on the tensile properties of ZK51A alloys.

solution heat treating at 930 F for 2 hr and then aging at 300 F for 48 hr.

Since the purpose of this investigation was to determine the properties of material aged directly from the as-cast condition, a group of test bars having greater than 0.7 per cent soluble zirconium was aged at various temperatures for different periods of time. The results of these tests, as shown in Table 1, revealed that essentially equal properties were obtained over a rather wide range of aging times and temperatures. It was decided that a treatment at 350 F for 12 hr gave satisfactory properties and would be a practical treatment for commercial consideration.

The possibility of a solution heat treatment was studied but will not be presented in detail. A group of tensile bars from several alloys which had soluble zirconium contents from 0.6 to 0.8 per cent were heat treated for various lengths of time at 930 F and subsequently aged at 300 F for 48 hr. The tensile properties of these alloys were grouped according to the soluble zirconium content, above and below 0.7 per cent, and the resulting average properties are shown in Fig. 2.

It may be seen that the alloys with less than 0.7 per cent soluble zirconium which were solution heat treated at 930 F for 10 to 12 hr followed by aging have tensile properties nearly equivalent to those obtained on the alloys with greater than 0.7 per cent soluble Zr which were only aged at 350 F for 12 hr. Or, observing the data in a different light, it may be seen that the alloys with soluble zirconium contents above 0.7 per cent and aged only developed tensile properties as high as those obtained by solution heat treating and aging.

Equivalent tensile properties cannot be obtained with a simple aging treatment if the soluble zirconium content is less than 0.7 per cent. Table 1, which shows the tensile properties in the various aged conditions, also lists for comparison the properties obtainable in as-cast, solution treated, and solution treated and aged test bars.

Further tensile tests were made on alloys contain-

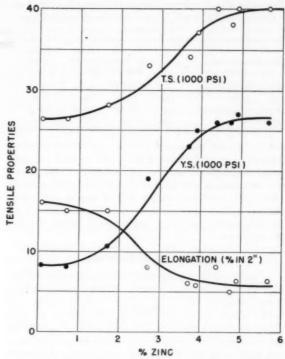


Fig. 3—Effect of zinc on tensile properties of Mg-Zn-Zr alloys in the T-5 condition.

ing essentially the maximum soluble zirconium content, but with increasing zinc contents, i.e., from zero to 6 per cent. These alloys were given the previously selected aging treatment at 350 F for 12 hr. The results of the tensile tests are plotted as a function of zinc content in Fig. 3. It may be seen that at 4.5 per cent zinc the curve of the yield strength levels off showing no significant change from 4.5 to 5.5 per cent zinc.

The facts presented seem to establish a composition for ZK51A alloy around 4.5 to 5.0 per cent zinc, with

TABLE 1—TENSILE PROPERTIES OF SEPARATELY CAST ZK51A TEST BARS AFTER VARIOUS HEAT TREATMENTS

ASTM		Trea	tment	Tensile	Yield	Elong-
Temper Designation	Condition	Time,	Temp.,	Strength, 1000 psi	Strength,* 1000 psi	ation, % in 2 in
F	As-Cast			38	21	9
T5.	Aged	12	275	40	25	9
		24	275	40	26	8
		48	275	39	26	5
		12	300	41	26	8
		24	300	41	27	8
		48	300	41	28	8
		6	350	40	26	8
		12	350	41	27	8
		24	350	41	26	8
		6	400	40	25	8
		12	400	40	26	9
	Solution					
T4	Heat	11	930	36	14	15
	Treated					
	Solution					
T6	Treated	11	930			
	and Aged	48	300	42	28	7
	sted in all tables were using an autographic					

TABLE 2—COMPARATIVE TENSILE PROPERTIES OF BARS
CUT FROM ZK51A CASTINGS AND SEPARATELY CAST
TEST BARS

Alloy Designation and Condition	Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elong- ation, % in 2 in.	
ZK51A-T5 bars cut	range	31-41	21-28	2-13
from casting	ave.	37	24	6
ZK51A-T5 separately	range	39-42	26-28	5-10
cast test bars	ave.	41	27	8

a 0.7 per cent minimum soluble zirconium content. This alloy is aged at 350 F for 12 hr to produce optimum tensile properties. For convenience, in this paper the composition and heat treatment are designated as ZK51A-T5 according to the ASTM system of nomenclature. Alloys meeting these specifications were used to determine the other mechanical properties reported here.

# **Evaluation of Properties**

A group of 48 tensile test bars was machined from four ZK51A sand cast aircraft landing wheels which were about 18 in. in diameter. The castings had been aged from the cast state for 12 hr at 350 F. The range of properties obtained and the average values are recorded in Table 2. For comparison purposes this table also lists the range and average properties typical of separately cast test bars. From these rather meager data it appears that the spread between the average properties of the separately cast test bars and the bars cut from the castings is smaller in this alloy than is normally encountered in the conventional Mg-Al-Zn alloys.

A group of important mechanical properties of ZK51A-T5 alloy is listed in Table 3 where they can be compared with the corresponding properties of AZ92A-T6 and AZ63A-T4.* The ductility of ZK51A-T5 as illustrated here by the elongation and reduction in area (the latter was measured on a machined surface test bar) shows that ZK51A-T5 alloy is intermediate between AZ63A-T4 and AZ92A-T6.

The hardness data were obtained on test bars and castings using a Brinell machine with a 500-kg load and a 10-mm steel ball. The impact tests were made on a modified Charpy specimen having a 45° V notch, but otherwise conforming to the standard Charpy

The shear and bearing strengths are also shown in Table 3. The shear strength was determined using the standard double shear test method. The bearing strengths were determined using an 8 x 2 x 1/8-in. plate machined from a cast panel. The load was applied through a 1/4-in. pin inserted in a 1/4-in. centerline hole drilled and reamed to size at 1.5, 2.0, 2.5 and 4.0 hole diameters from the end of the specimen.

The area under test above the pin (between the pin and specimen edge) was unrestrained against

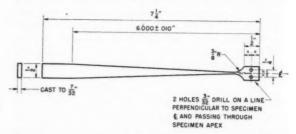


Fig. 4—Dimensions of the cantilever beam specimen used for relaxation tests.

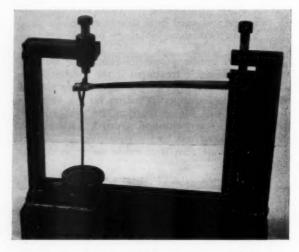


Fig. 5—Jig for cast cantilever relaxation specimen with weight pan and specimen.

TABLE 3—Some Properties of ZK51A-T5 Compared with AZ63A-T4 and AZ92A-T6

Alloy Tensile Designation Properties			Brinell Hardness Number, Impact Strength					Shear	Bearing Strength (1000 psi)						)		
and	TS,	YS,	Elong.,	Red. in	500-kg	load	notched,	unnotched,	Strength,	1.5	5D•	2.0	D*	2.5	D*	4.0	D•
Condition	1000 psi	1000 psi	% in 2 in.	area, %	10-mm	ball	ft-lb	ft-lb	1000 psi	YS	TS	YS	TS	YS	TS	YS	TS
ZK51A-T5	41	26	8	10	65		13/4	13	21.7	45	61	54	75	56	82	59	86
AZ63A-T4**	40	14	12	14	55		41/2	27	18	38	44	44	60	46	70		
AZ92A-T6**	40	23	2	. 1	84		3/4	4	21	54	62	65	80	65	85		
• D = the hole	diameter	(1/4 in.).	1.5D, etc.,	is the	distance		2.4										

from hole center to the edge of the specimen through which the load is applied.

bar dimensions. Unnotched Charpy bars were also tested, but the notched bar is believed to be more indicative of the actual service characteristics of the alloys. Both the hardness and impact strength of ZK-51A-T5 are intermediate between AZ63A-T4 and AZ-92A-T6.

^{*}AZ92A-T6 is a Mg alloy containing 9.0 per cent Al and 2 per cent Zn in the solution heat treated and aged condition. AZ63A-T4 is a Mg alloy containing 6 per cent Al and 3 per cent Zn in the solution heat treated condition.

^{**} Typical Test Bar Data.

Table 4—Fatigue Properties of ZK51A-T5 Compared with AZ63A-T4 and T6 and AZ92A-T6 at Room and Elevated Temperatures

			R. R. M	oore-Rota	ating Bea	* Krouse Plate Bending Fatigue **							
Alloy Designation and	Test Temp.,	Plain Bar stress (1000 psi) at cycles		SCF stress (1	-	Notched Bar, SCF*-5 stress (1000 psi) at cycles		stre	Cast Surfa ss (1000 p at cycles	psi)	Machined Surface, stress (1000 psi) at cycles		
Condition	F	106	10 ^a	10°	100	10°	10 ⁸	10 ⁸	10°	107	105	10°	107
ZK51A-T5	70	11	9	11	9	10	6	15	7	6	17	10	9
AZ63A-T4	70	18	14	12	8	8	7	12	10	8	15	13	12
AZ92A-T6	70	18	15	11	9	6	5	14	10	9	20	16	13
ZK51A-T5	300	11	8	7	5						8	7	7
AZ63A-T4	275			8	5								
AZ63A-T6	300										12	9	6
AZ92A-T6	275	13.5	10	8.7	6								
eometry of the	Concentration notch and na e for complete	ture of th	ne applie		ef. 16).								

buckling by any clevis arrangement. The strain was recorded autographically in a manner that would give a precise measurement of the elongation of the hole. The yield strength was ascertained at a 0.2 per cent offset from the stress strain curve. Both the bearing and the shear strengths are better than those of AZ63A-T4 and approach closely the values reported for AZ92A-T6.

# **Tests Represent Service Conditions**

The fatigue properties of ZK51A-T5 alloy are presented in Table 4. R. R. Moore rotating beam and Krouse plate bending fatigue data were obtained at 70 and 300 F. Similar data are shown in this table for AZ92A-T6 and AZ63A-T4 alloys. The test specimens and procedures are fully described by Found. Although fatigue tests are routinely made and reported by many laboratories, the results may not be a direct prediction of the serviceability of actual castings. The rotating beam tests are made on polished surface bars rotated, under load, like an axle.

The introduction of a notch is probably a closer approach to some of the conditions of service. The plate-bending tests, with either the as-cast or machined surface, may be even more representative of service conditions. Because of the small number of tests run, the data are based on an average curve and hence do not adequately describe the scatter band which is characteristic of the laboratory fatigue tests. The data in Table 4, therefore, are interpreted to indicate that ZK51A-T5 cannot be expected to give better service performance under fatigue conditions than AZ92A or AZ63A alloys.

The possibility of using ZK51A alloy at moderately elevated temperatures required an evaluation of its creep strength at these temperatures. The 100-hr creep tests were run using the method described by Moore and McDonald¹⁷ and the results are recorded in Table 5. It must be remembered that scatter is also common in creep tests. ZK51A-T5 is slightly better on the average than AZ92A or AZ63A up to 300 F, but at higher temperatures all three alloys lose strength rapidly. The differences in creep strength shown in Table 5 are not great enough to insure that there would be any difference in the creep strength of castings made from these alloys.

Additional investigations were conducted to determine the growth characteristics of ZK51A alloy. Several measurements of growth were made on this alloy starting with as-cast material. The alloy did not show any growth during exposure for 200 hr at temperatures up to and including 400 F. The test procedure followed has been described by Leontis and Nelson¹⁸ who have shown that this was ample time to bring ascast AZ92A or AZ63A alloys to essentially maximum growth at 350 F. It may be concluded that any precipitation aging of ZK51A alloy causes very little linear dimensional change.

Some stress relaxation tests were performed on ascast ZK51A alloy using a uniformly stressed cantilever beam. The specimen used for this test was contoured to the dimensions shown in Fig. 4 from a cast panel 7/32 in. thick. The specimen was then mounted in the steel jig as shown in Fig. 5. A load was applied to the unsupported end to produce a predetermined surface stress in the specimen.

The deflection of the bar was maintained by a set screw. The load was removed before each heat treatment. The specimen and the jib were heated and cooled separately to prevent heat transfer from the specimen to the jig or vice versa. The jig and specimen were carefully marked after the initial setting to insure accurate remounting after each stage of the test. The initial set screw adjustment and subsequent loadings were determined by using an electric current to signal the instant at which contact was made or broken between the specimen and the set screw.

TABLE 5—CREEP LIMIT IN 100 HOURS OF ZK51A-T5 AS COMPARED WITH AZ63A-T6 AND AZ92A-T6

Alloy		Creep Limit*					
Designation and Condition	Test Temp., F	0.1% Creep Extension	0.2% Total Extension	0.5% Total Extension			
ZK51A-T5	200	10.0	8.1	14.0			
	300	5.0	5.0	11.4			
	400	1.6	2.2	4.2			
AZ63A-T6	300	3.5	4.0	8.5			
	400	1.3	1.8	2.8			
AZ92A-T6	200	8.5	7.8	13.5			
	300	3.2	. 3.9	7.0			
• Stress in 10	000 psi.						

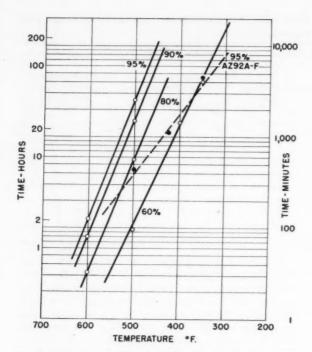


Fig. 6—Stress relaxation of ZK51A-F alloy as a function of time and temperature.

After each heat treatment the load required to break the electrical circuit was converted to stress in the specimen using the equation for a uniformly

stressed cantilever beam; 
$$S = \frac{6 PL}{bd^2}$$
, where S is

the stress in the beam, P is the applied load, L is the beam length, and b and d are the beam width and thickness, respectively. The difference between the initial stress and the stress obtained after successive times at heat treatment temperatures represent the degree of stress relief.

These tests show that ZK51A in the as-cast condition requires a prohibitively long time for stress relief at 350 F, but as the curves in Fig. 6 show, 95 per cent stress relief can be attained by a short 2-hr heat treatment at 600 F. This fact essentially substantiates the work reported by Payne¹⁹ who used the split ring and wedge technique to measure relaxation.

# Conclusions

An alloy containing at least 0.7 per cent soluble zirconium and around 4.5 to 5.0 per cent zinc possesses an interesting set of mechanical properties.

The heat treatment selected depends on the zirconium content of the alloy, or the purpose for which the casting is to be used. One of the most interesting features of this alloy is the attainment of relatively good properties with only an aging treatment.

In general, the properties are not unlike those of AZ92A-T6 alloy with one notable exception; ZK51A-T5 has a better ductility-yield strength ratio.

ZK51A-T5 alloy has potential future uses in the structural casting field whenever this combination of ductility and strength can be efficiently used. How-

ever, designs in which fatigue life is critical will require actual service tests before the fatigue characteristics of this alloy can be established.

The higher cost of this alloy and present difficulties in the foundry, such as surface imperfections, do not warrant extensive commercial use at the present time.

#### References

 T. E. Leontis and J. P. Murphy, Transactions, AIME, vol. 166, p. 295 (1946).

T. E. Leontis, Transactions, AIME, vol. 180, p. 287 (1949).
 K. E. Nelson and F. P. Strieter, Transactions, A.F.S., vol. 58, p. 400 (1950).

4. K. E. Nelson and F. P. Strieter, Transactions, A.F.S., vol. 59, p. 532 (1951).

F. Sauerwald, Z. F. Metallkunde, vol. 40, p. 41, Feb. 1949.
 C. J. P. Ball, Metallurgia, vol. 35, p. 125, 211 (1947).

 J. P. Doan and G. Ansel, Transactions, AIME, vol. 171, p. 286 (1947).

8. J. W. Meier and M. W. Martinson, Transactions, A.F.S., vol. 58, p. 742 (1950).

9. W. P. Saunders and F. P. Strieter, A.F.S. PREPRINT No. 52-78: to be published in A.F.S. Transactions, vol. 60 (1952).

 C. E. Nelson, Transactions, AIME, vol. 159, p. 392 (1944).
 Magnesium und Seine Legierungen, Edited by Dr. Adolph Beck, Berlin, Julius Springer, 1939.

12. Stroup and Sager, U. S. Patents 2,373,515 and 2,373,516.
13. Dow Chem. Co., Analytical Method, MLW 51.13, May 31,

1951.14. Dow Chem. Co., Analytical Method, MLW 51.5, June 11, 1951.

15. U. S. Patent 2,520,753.

G. H. Found, Proceedings, ASTM, vol. 46, p. 715 (1946).
 A. A. Moore and J. C. McDonald, Proceedings, ASTM, vol. 46, p. 970 (1946).

18. T. E. Leontis and C. E. Nelson, *Transactions*, AIME, vol. 191, p. 120 (1951).

19. R. J. M. Payne, *Journal*, Inst. of Metals, vol. 18, p. 147 (1950).

#### DISCUSSION

Chairman: W. J. Klayer, Aluminum Industries, Inc., Cincinnati.

Co-Chairman: R. C. Военм, Wellman Bronze & Aluminum Co., Cleveland.

Recorder: W. J. KLAYER.

P. A. FISHER (Written Discussion): It is, I think, generally appreciated in the light alloy industry that the alloys based on the addition of zirconium to magnesium mark a most important advance in this field, and, having been engaged in this work myself for the past 8 or 9 years, it really is a pleasure to discuss various aspects of this work with other students of the same problems.

I believe it is generally known in the trade that my Company and its associates have carried out a considerable amount of research and development work on magnesium-zirconium alloys. I find it rather surprising that the authors have not referred to this work in greater detail. There have been at least three papers published dealing with the commercial exploitation of these alloys. One of these, by H. Millward, was published in the AMERICAN FOUNDRYMAN as recently as September, 1951.

As pointed out by the authors, work on magnesium-zirconium alloys was first started in Germany, and in 1939 my Company commenced its research work using a mixture of zirconium chloride with sodium or potassium chloride, i.e., essentially the same mixture as used for part of the work described in this paper. A number of years of intensive work followed, from which a sufficient number of promising results were achieved to encourage us to believe that the alloy system was of great potential value, but at the end of which we were forced to conclude that the use of salts of zirconium chloride, potassium and sodium chloride type was technically unacceptable. In other words our best results were outstanding, but owing to the use of the salt based on zirconium chloride, our overall scatter in results was too great.

¹ Metallurgist, Magnesium Elektron Ltd., Manchester, England,

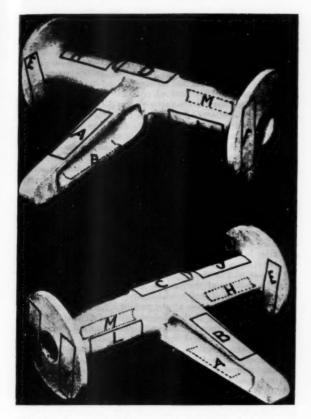


Fig. 1A—Test casting GR14 showing position of test specimens.

MARK		Z42Z.			A 8.	
	O. 12 P. S.	U. T. S. P. S. A.	E %	O. 12.P S.	U. T. S	E%
A.	20, 200	39, 900	12. 5	10, 100 10, 100 9, 200 9, 800	37,000 35,700 36,500 38,400	12.50 13.75 10.00 12.08
С	18, 300	33, 400	6.25 *	10, 700 9, 400 8, 900 9, 670	21,300 21,000 20,000 20,770	5.50 4.00 3.00 4,17
D.	22, 200	42,400	13.0	12, 100 11, 800 12, 100 12, 000	30.700 33,000 31,100 31,600	7.50 7.00 6.50 7.00
L.	20,700	44, 200	12.5	8, 400 8, 400 9, 500 8, 770	24,000 21,800 37,500 27,770	12.50 4.00 2.50 9.67
м.	21, 900	41,600	12,5	11, 200 8, 900 9, 500 9, 870	23,800 23,500 36,000 27,770	5.00 4.00 12.50 7.17

- O MEAN OF THREE VALUES.
- # FRACTURED THROUGH INCLUSIONS
- + FRACTURED OUTSIDE GAUGE LENGTH
- + MICROSHRINKAGE

Fig. 2A—Showing comparison of properties from test casting GR14 for alloys Z4½Z (heat treated 16 hr at 200 C) and Elektron A8 (solution heat treated).



Fig. 3A-Oil sump; weight 101/4 lb.



Fig. 4A-Rear axle casting; weight 671/2 lb.



Fig 5A-Reduction gear casting; weight 75 lb.

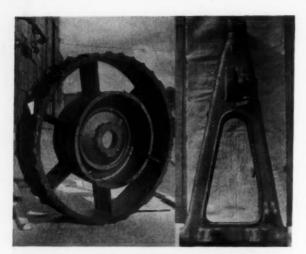


Fig. 6A (Lett)—Air intake; weight 97 lb. Fig. 7A (Right)

Nose wheel leg weight 265½ lb.

At this stage, therefore, my Company started research work to evolve a more satisfactory method of introducing the zirconium. During the course of this work, a considerable number of alloying methods were evolved and tested, including the use of the hardener mentioned in this paper. In carrying out this work, it was found that various alloying materials, while capable of introducing consistently over 0.7 per cent soluble zirconium, gave rise simultaneously to varying degrees of microporosity in the final alloy. As an instance of the effect of this porosity on tensile properties, one alloying material introduced 0.7 per cent soluble zirconium to Z5Z, but the tensile properties were:

0.2% Yield Strength — 22,500 PSI U.T.S. — 24,000 PSI Elongation — 1%

These low properties were not brought about by the presence of traces of foreign metals.

In developing a suitable alloying material it was therefore necessary to provide a material which would consistently introduce at least 0.5 per cent soluble zirconium without giving rise to appreciable microporosity.

Finally, a salt mixture which satisfied these requirements and which we term "Master Salt" was evolved. Using this material, the highest level of results obtained in the earlier work could be achieved consistently and the alloys produced in a simple and economical manner.

At this time, 1946, the alloys were introduced to the trade. The new alloys could be classified into two distinct groups:

1. For use at elevated temperatures. These are the alloys containing the rare earth metals with which we are not concerned at this meeting.

2. An alloy possessing superior properties at room temperature; in particular characterized by a high value for yield strength, coupled with good ductility. This alloy was termed by us Z5Z, and is identical with Zk51, which is the subject of this paper.

I should point out that our Z5Z alloy, although sometimes described as containing about 0.6 to 0.7 per cent zirconium has, in fact, always contained over about 0.7 per cent soluble zirconium

as determined by chemical analysis.

From the earliest days of the introduction of this alloy, my Company has aimed at simplicity of production coupled with a high level of properties. This, we found, was achievable by using a high level of zirconium content, which in itself is essential to ensure good castability, followed by a short simple heat treatment. The difficulty of ensuring a high level of zirconium content while at the same time avoiding various other undesirable factors, which is so common to many alloying methods, was overcome by our development of the particular material and methods upon which our process is based.

At this stage therefore, we had an alloy possessing high yield strength, high ultimate strength, and good ductility. This alloy also had the very great virtue that the properties of castings

made from it were much closer to those from separately cast testbars than is the case with the older alloys. Figure 1A shows a casting from which test coupons were cut. Figure 2A shows the properties obtained on these test coupons. As soon as this alloy was introduced, the aircraft industry showed an immediate interest in it, and very soon prototype castings were being produced. That interest has continued, and the results of the prototype castings fully justified our faith in the alloy and so the use of the alloy has gradually grown. Since its introduction many hundreds of tons of castings in this alloy have been made and used, and it is now accepted by foundries engaged in its manufacture as a normal production alloy, the demand for which increases daily as new aircraft and other types of equipment come into production. As an example of the use of this alloy, I should mention that the British "Comet" Aircraft, which today is making the first flight of the first regular passenger service operated by a jet propelled plane, contains over 1,000 magnesium alloy castings, the majority of which are in this new

To illustrate some of the castings now being produced in this alloy, I would like to show you five illustrations (Figs. 3A-7A).

I think I have now said enough to outline the development of this alloy in England, and to convince you that the alloy is and has been for some years, in regular commercial production.

You may well be wondering, in view of my remarks, why it is that the authors of the present paper are compelled to conclude, as a result of their work, that this alloy "does not warrant intensive commercial use at the present time." The answer to this is quite simple. If the method used for alloying zirconium is unsatisfactory, then the alloy will also be unsatisfactory both in foundry behavior and mechanical properties. The authors have, in effect, I believe, reached the same conclusions which my Company reached a number of years ago, i.e., the method of introducing zirconium based on the use of zirconium chloride is

just not good enough for Zk51 alloy.

With this fact in mind, it is easy to see why the authors have chosen to examine the possibilities of achieving a high level of properties by resorting to a difficult and prolonged heat treatment. In our experience, the materials and techniques described in the paper are inadequate to achieve consistently the high level of properties in castings which can be obtained by a simple short heat treatment. I would ask the authors if they have attempted to heat treat Zk51 castings of any complexity at 930 F, a temperature at which the alloy is almost plastic. The problem is, however, much more complex than this. Our work has shown that the response to heat treatment at 930 F varies with the true soluble zirconium content of the alloy and the section thickness of the casting. In other words, to achieve a high level of properties, the duration and temperature of the solution treatment would require adjustment to suit the soluble zirconium content and section thickness of individual castings. One can imagine the practical consequences of attempting to operate such a process.

In fairness to the authors, I agree that this paper is valuable in confirming some of the favorable characteristics of this alloy. Considering the mechanical properties listed in Table 3 one can see the order of tensile properties achievable. The high level of properties obtained in actual castings is mentioned, and I can confirm that this is a well proved feature of this alloy when made correctly. The authors, in my opinion, do not pay enough attention to the properties achievable from actual castings when comparing the level of tensile properties with those shown by the aluminum containing alloys. Tests on actual castings have shown that the level of ultimate strength in Z5Z is 9 to 40 per cent better than that for the older alloys. This is achieved with good ductility which in AZ92T6 is frequently much lower than that indicated by the value of 2 per cent elongation quoted in this paper.

With regard to fatigue strength, I agree with the authors that it is difficult to correlate the values obtained from separately cast specimens with those of the actual casting, and the desirable aim is to test the casting itself. Probably the most searching fatigue duty to which a magnesium alloy casting is subjected is in the aircraft landing wheel. My Company has experience of endurance tests carried out on landing wheels which show that

^{*} Note: The designation "Z5Z" had not been allocated to this alloy at the time these illustrations were prepared: the illustrations, therefore, show the alloy under its temporary designation, Z4½Z."

Z5Z wheels have improved life over those made from the older alloys. However, fatigue tests on landing wheels are themselves somewhat difficult to evaluate and the ultimate test is service experience. Some years ago, a large number of landing wheels in Z5Z were put into service in England on various new types of aircraft and the necessary service experience is now being obtained. I should add that, so far, no case of failure has been reported.

To sum up then, the alloy investigated by the authors has been in commercial production for a number of years, giving the interesting properties mentioned by the authors, and is commercially available in this country now. I believe that the authors' inability to achieve consistently the high level of properties possible from this system is due to their use of inadequate alloying materials, but I am frankly quite puzzled at their reference to surface defects and would like further details of this.

Finally, I would suggest that the last paragraph of the conclusion of this paper should be modified to restrict the application of this conclusion to the alloy when made from the alloying materials described in this paper.

Messrs. HILDEBRAND AND STRIETER (Reply to Mr. Fisher): Much of the discussion by Mr. Fisher is extraneous to this particular paper and has been covered in a companion paper "Alloying Zirconium to Magnesium" by W. P. Saunders and F. P. Strieter. (See pp. 581-594 in this volume.)

A study of the present paper would indicate specifically that there is no basic conflict between the present authors and Mr. Fisher in the matter of the ZK51 alloy composition, particularly minimum Zr content, tensile properties, either from castings or test bars, or in the preferred method of aging directly from the cast state.

This brings us to the main issue of his discussion, namely, the servicability characteristics. For most parts, in order to lighten the casting or increase its strength-weight ratio, an alloy is needed that possesses better fatigue characteristics than current alloys, since most service failures are due to fatigue. ZK51 does not appear to offer any advantage in this direction.

The information on castings made in England from this alloy is very interesting. We will look forward to learning of some factual data on the servicability of this alloy compared with alloys previously used.

We feel strongly that ZK51 has poor castability. This has been true, regardless of the source of Zr. Surface defects such as cracks, shrinks, and draws are the main factors. It is our impression that similar difficulties have been a problem in England and that they have been working on alloy modifications, such as rare earth additions, to reduce these defects.

J. W. Meier (Written Discussion): The authors are to be congratulated for this interesting paper and their work on magnesium-zinc-zirconium casting alloys. Being interested in this field, for a number of years, I would like to add some comments which seem to be necessary because of the somewhat misleading title of the paper. This title implies that mechanical properties of all known casting alloys based on the Mg-Zn-Zr system are being discussed in this paper and this is obviously wrong, because alloys with higher zinc contents (above 5 per cent Zn) are not included at all and, with the exception of Fig. 3, all of the results given pertain only to one alloy, designated by the authors as ZK51A. The omission of casting alloy ZK61, which has the highest strength-to-weight ratio of all commercial casting alloys, ferrous or non-ferrous, lead the authors to some conclusions which are contrary to our own experience.

First, the choice of the zinc content. In an earlier paper⁸ we showed the effect of the zinc content on the tensile properties of as-cast and fully heat treated Mg-Zn-Zr casting alloys containing 0-10 per cent Zn and over 0.7 per cent soluble zirconium. We later revised this diagram by adding some results in the aged condition (24 hr at 356 F as used in British practice). This diagram shown here (Fig. 8A) indicates very clearly why we consider alloy compositions in the range of 5.5-6.5 per cent Zn as the most outstanding in this alloy system.

Tensile properties are, of course, only one of many characteristics which affect the choice of structural materials. Data on other properties of ZK61 are available and will be included in a more comprehensive comparison of Mg-Zn-Zr casting alloys to

be published in the near future. I want to mention here only one other important characteristic of this alloy and that is its excellent response to heat treatment.

My second comment concerns the choice of proper heat treating cycles for various compositions in this alloy system. As mentioned in our earlier paper⁶ the heat treating cycle consisting of a short high-temperature solution heat treatment (1-5 hr at 800-930 F) and a longer low-temperature ageing treatment (24-8 hr at 265-300 F) was worked out especially for casting alloy ZK61 and, as Fig. 8A shows, is very effective in the range of 5.5-6.5 per cent.

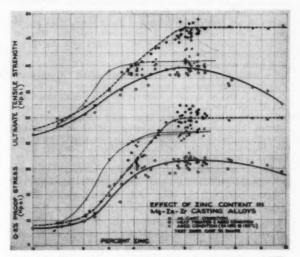


Fig. 8A—Effect of zinc content on tensile properties of ZK alloys.

For alloys with other zinc contents the most efficient heat treating cycles are different and have to be based on solidus temperature determinations for each particular alloy composition. This explains why the authors, using the ZK61 heat treating schedule for alloy ZK51A could not improve, to any significant extent, the tensile properties of this alloy. The prolonged soaking periods used by the authors were unsuccessful, and, we found that the use of higher solution temperature (960-1020 F) for alloys containing 4-5 per cent zinc resulted in only slightly higher tensile properties. It would seem from these experiments that the two-stage heat treatment for alloys having zinc contents below 5 per cent offers no particular advantage over the one-stage treatment (as shown in Fig. 8A).

This supports our former statement that alloy ZK61 covers the most attractive composition range and is the only alloy in the Mg-Zn-Zr system which is amenable to a heat treatment which substantially increases not only the yield strength but also the ultimate tensile strength of the alloy without impairing its high elongation.

Table A shows the effect of various heat treating cycles on ten-

TABLE A—TYPICAL TENSILE PROPERTIES OF

	Sol.	H.T.	A	ging	UTS,	0.2% PS,	Elong.
	Hr	°F	Hr	°F	kpsi	kpsi 9	o in 2 in.
ZK51	(4.5-5.	0% Zn)	— acc	ording t	o Hildebr	and and S	trieter
F		_	-	_	38	21	9
<b>T5</b>		_	12	350	41	27	8
<b>T6</b>	11	930	48	300	42	28	7
ZK61	(5.5-6.5)	% Zn) -	_ obta	ined at	Canadian	Bureau of	Mines
F	_	-	_	-	38-42	20-22	9-14
T51	_	_	24	355	38-42	26-28	5-8
T52		-	48	300	40-43	27-30	5-8
T61	2	930	48	300	44-48	28-33	8-12
T62	2	930	48	265	46-48	30-34	7-10
Canadi	an DND	Specifi	cation	C-28-97			
F	_	_	_		min. 35	min. 18	min. 8
T61	2	930	480	300	min. 42	min. 26	min. 5

 $^{^{2}}$  Canadian Department of Mines & Techical Surveys, Ottawa, Ontario, Canada.

sile properties of casting alloys ZK51A and ZK61. The table shows, also, the minimum requirements of the Canadian Air Force specification for alloy ZK61 issued in August, 1950. A considerable number of ZK61 alloy castings have been produced in Canada and are being successfully used in various modern

aircraft applications.

There is another comment concerning the correlation of grain size and soluble zirconium contents shown in Fig. 1. Our experience, based on many hundreds of ZK alloy melts, indicates that this relationship is much closer than shown in this diagram. From our results it would seem that alloys with over 0.8 per cent soluble Zr must have a much smaller grain size than 3 thousandths and, on the other hand, we never found an alloy with 0.65 per cent or less zirconium which would have a grain size of 2 or 3 thousandths. It would seem that the "effective" zirconium content cannot always be identified with the analytically determined acid-soluble zirconium content. We found that grain size and mechanical properties of ZK casting alloys are a much more reliable indication of the real "effective" zirconium content than its chemical determination.

There is also no mention made on the room-aging characteristics of alloy ZK51A. Our experience⁸ indicates definitely that all ZK casting alloys containing more than 2-3 per cent Zn are affected by room-temperature aging and show increases in strength connected with a slight drop in elongation. Fortunately, we could not find any significant dimensional changes due to this room-temperature aging.

To sum up, it was not my intention to criticize the very good work of the authors or their preference for casting alloy ZK51A

which I consider to be a very useful material, especially for applications where the highest strength-to-weight ratio is not essential. As mentioned already, it seemed necessary to add some data on alloy ZK61 and so to justify, to some extent, the all too general title of the paper.

Messrs, Hildebrand and Strieter (Reply to Mr. Meier): We want to thank Mr. Meier for his discussion of our paper. We would agree with him that ZK61A-T6 offers higher tensile properties than the ZK51A-T5 composition. There was no intent on our part to discredit the good properties which the Canadians have been getting on their alloy containing 6 per cent zinc. In fact, it was felt unnecessary to cover this composition because of the 1950 A.F.S. paper by Meier and Martinson.* Rather, we believed that the presentation of some of the properties obtainable in ZK51A-T5 alloy might help to clear up some of the confusion that exists regarding the utility of this composition.

Mr. Meier indicates his preference for the use of methods to measure the "effective" zirconium content rather than the acid soluble zirconium content. This is undoubtedly a satisfactory method. We prefer to analyze our melts chemically as a control over composition, but we would expect that the grain size or properties could also be used for control purposes so long as consistent cooling rates were always maintained on the control specimens.

Our experiences in regard to room temperature aging and growth are in agreement with those of Mr. Meier.

^{*} See "Development of High-Strength Magnesium Casting Alloy ZK61," by J. W. Meier and M. W. Martinson, A.F.S. Transactions, vol. 58, pp. 742-751 (1950).

# EFFECT OF LEAD AND NICKEL ON GRAIN SIZE OF CERTAIN CAST COPPER-BASE ALLOYS

By

R. A. Colton, F. L. Turk, and D. L. LaVelle*

In an earlier technical paper presented before the American Foundrymen's Society¹ a technique was described for studying the effects of composition on the grain size of copper-base alloys by minimizing cooling rate as a variable. By using the method described, experiments were conducted to determine the effects of nickel on the grain size of a series of binary alloys of copper containing lead, tin, or zinc.

From the results of these experiments it was found that lead had a strong grain refining effect on copper, tin a weaker effect on grain size, and zinc a negligible effect. Nickel was found to have little or no grain refining effect except in some copper-tin alloys. Since the purpose of the experiments had been primarily to develop suitable techniques and to determine qualitative effects of certain elements on grain size, the question of quantitative relationships between grain size and these elements remained unanswered.

To complete the investigations reported earlier and to establish, if possible, practical implications of the findings, additional experiments using the same general techniques were carried out. Rather than consider this phase of the investigation as a distinct entity, the experiments were designed to complement those already conducted. Consequently, quantitative studies of the effect of lead on the grain size of copper and of the effect of nickel in copper-tin alloys were made. In addition, the effects of lead in three widely used commercial tin bronze alloys, SAE 62 (88 Cu, 10 Sn, 2 Zn), SAE 65 (89 Cu, 11 Sn), and Navy "G" (88 Cu, 8 Sn, 4 Zn) were observed.

The bulk of evidence available indicates that cooling rate is by far the most important influence in determining grain size in casting alloys. By obtaining information on possible influences of composition, however, it is hoped that some measure of control can be exercised over what has otherwise been a moreor-less randomly determined property. Such information may also be of some assistance in evaluating specifications for casting alloys.

As mentioned in the earlier report, there has been no entirely satisfactory experimental evidence ever given that fine grain size, as such, is beneficial in casting alloys. In fact, much can be said to contradict the idea that fine grain size improves either the quality of the casting or the physical properties of the metal. Regardless of what may be the final conclusions of such a discussion, it is essential that a better understanding of the factors contributing to grain size be available so that whether fine or coarse grain is desired, control of this property can be exercised. It has been in this spirit that these experiments have been conducted.

# **Experimental Procedures**

It should be apparent that any studies of the effect of chemical composition on grain size of casting alloys must necessarily exclude cooling rate as a variable since it has been demonstrated that this is a more significant factor than composition. Since previous experience has indicated that pouring of castings is an impractical method of studying grain size because of the difficulty of maintaining constant experimental conditions, the experimental technique used was that described in the previously reported work where, to minimize cooling rate as a variable, a method of melting and cooling the various alloys simultaneously in an electric furnace was necessarily used.

Melting was carried out in graphite crucibles 1 in. 1.D., 8 in. long with 3/32-in. wall thickness, using a small Ajax-Northrup high frequency induction furnace. With the lining diameter of the Ajax furnace approximately 6 in. it was possible to group nine of the small graphite crucibles around the periphery, standing on a circular block of graphite. Two more crucibles could be placed in the center of the furnace so that, if desired, 11 separate melts could be made simultaneously, brought to the same temperature, and when the power was turned off, allowed to cool simultaneously.

For all practical purposes, the cooling rate of the 11 melts in the furnace could be considered identical. A graphite cover was placed on the furnace to reduce the heat loss by radiation from the top, and also, to

^{*} American Smelting and Refining Co., Research Dept., Barber, N.J.

slow down the rate of cooling. With this arrangement solidification apparently starts at the bottom of the crucible and proceeds toward the top for at least half of the freezing interval.

Cooling curve experiments described in the previous report gave evidence that the cooling characteristics of the furnace are reasonably uniform so that comparisons can be made between melts made separately as well as among melts made at the same time.

Separate experiments were made to determine quantitatively the effect of lead on the grain size of copper, the effect of nickel on Cu-Sn alloys, and the effect of lead on a variety of commercial tin bronzes.

Only high purity metals were used to make the alloys for these tests. The copper-lead and copper-tin-nickel alloys were made up from the raw materials directly. When alloys containing zinc were used it was necessary to make 25-lb master melts which were cast into graphite molds, the castings being of suitable size for remelting in the crucible used in the experiment. This was done because of the problem of adding zinc to small melts where not only was accurate control of composition impossible, but also, the addition of the zinc experimentally difficult.

The "castings" obtained in these experiments were cylindrical billets 1 in. in diameter, 5 in. long weighing about 475 grams. A thin slice was cut from the bottom of the bar, and the bar was then cut longitudinally in half with a power saw. Both the transverse slice and the longitudinal section were polished, etched and observed at low magnification.

For purposes of delineation of grain size the high tin samples were etched in 10:1 ferric chloride-hydrogen peroxide (30 per cent). Other samples were etched in ammonium persulphate. Drillings for chemical analysis were taken in four positions from the unused longitudinal section and mixed thoroughly to insure proper sampling.

The procedure used was to melt down the copper first, and then, when molten, add the desired amounts of lead. At least 10 min was allowed for alloying. This is believed to be adequate since chemical analysis indicated that there was no marked segregation along the length of the castings. When the temperature was approximately 2150 F (1177 C) the furnace was turned off and solidification allowed to progress in the furnace, with a cover in place. The use of graphite crucibles, base block, and cover is believed to have controlled the oxygen content of the copper melts to a sufficiently low value to have eliminated the possibility of grain size effects due to this element.

The "castings" were then treated as already described. The chemical analysis of each is reported in Table 1. The polished and etched samples were photographed and are shown in Fig. 1.

The technique of comparing grain size by counting grains cut by parallel intercept lines drawn ½ in., 1 in., 1½ in., and 2 in. from the bottom of the casting was used here as it was in the previous study. While not an entirely satisfactory method of measuring grain size, this is adequate for comparing samples of the type used here. The results of these measurements are listed in Table 1.

Examination of the etched castings of Fig. 1 and the data of Table 1, shows well the marked grain refining effect of lead in cast copper. Apparently a minimum lead content is necessary, however, before this grain refining effect can occur. From these experiments it would appear that in pure copper, at least 0.30 per cent lead must be present for any noticeable effect on grain size, although 0.50 per

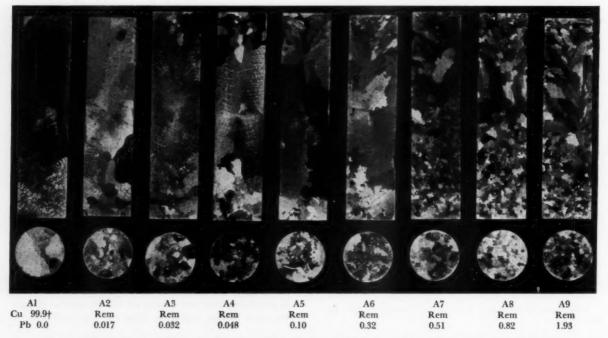


Fig. 1-Macroetched sections showing the effect of increasing lead content on the grain size of copper.

cent lead was required to refine the grain size to an extent where increased lead additions had no further effect. The data indicate a maximum in refining effect at 0.50 per cent with some coarsening found at higher lead contents. This coarsening effect may

not be genuine, however, and would have to be substantiated by test.

It must be remembered that grains are three-dimensional and cut sections show only two dimensions. For this reason allowances must be made for varia-

TABLE 1—CHEMICAL COMPOSITIONS AND GRAIN COUNT DATA

							Gra	in Count		
		(	Composition, %	ó		Dista	nce from	bottom of	bar, in.	
No.	Cu	Pb	Sn	Ni	Zn	1/2	1	11/2	2	
A-1	99.9+	_				2	4	3	4	
A-2	Rem	0.017				5	4	4	3	
A-3	Rem	0.032				8	11	4	1	
A-4	Rem	0.048				10	5	4	3	
A-5	Rem	0.10				6	2 7	2	2	
A-6	Rem	0.32				9	7	6	4	
A-7	Rem	0.51				15	15	16	15	
A-8	Rem	0.82				18	13	10	8	
A-9	Rem	1.93				14	13	12	7	
B-1	Rem		10.36			9	12	10	4	
B-2	Rem		9.00	0.96		12	13	13	7	
B-3	Rem		7.72	1.92		13	12	12	10	
B-4	Rem		6.98	2.57		17	17	13	13	
B-5	Rem		5.94	3.78		10	9	11	5	
B-6	Rem		4.92	4.83		9	8	7	4	
B-7	Rem			9.9		9	6	5	4	
B-8	Rem	1.03	4.69	4.33		14	14	12	13	
B-9	Rem	1.09	9.38			13	14	10	11	
C-1	Rem		11.3			11	11	7	5	
C-2	Rem	0.22	11.4			19	18	19	14	
C-3	Rem	0.51	10.9			22	19	21	17	
D-1	Rem	_	10.2		2.06	18	16	12	10	
D-2	Rem	0.24	10.2		1.95	20	18	18	16	
D-3	Rem	0.52	10.2		1.94	15	18	16	12	
D-4	Rem	1.05	10.3		1.99	19	19	14	15	
E-1	Rem	_	8.6		3.97	12	12	10	11	
E-2	Rem	0.22	8.5		3.95	13	15	14	14	
E-3	Rem	0.56	8.6		3.90	16	16	12	14	
E-4	Rem	1.05	8.3		3.94	15	15	14	13	14

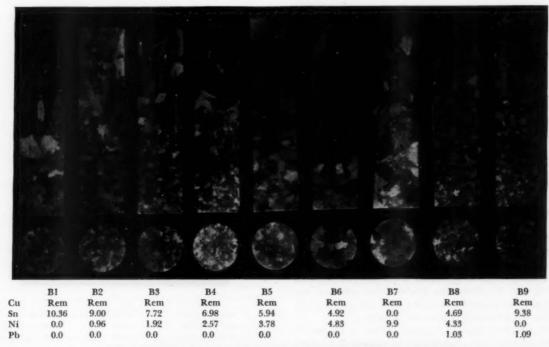


Fig. 2—Macroetched sections showing the effect on grain size of replacing tin with nickel in 90 Cu-10 Sn Alloy; also, the effect of lead in this alloy.

tions in apparent grain size. Only statistical studies of a large number of sections would give a truly

quantitative analysis.

Since the solid solubility of lead in copper is so small (0.002 to 0.005 per cent) it would appear that some grain refining effect should be evidenced with any lead content in excess of the solid solubility. The data do not show this clearly although examination of the etched castings does indicate some difference between the lead-free sample and those containing lead less than 0.30 per cent.

# Copper-Tin-Nickel Alloys

Continuing the work reported in the first paper, a study was made of the effect of nickel on the grain size of copper-tin alloys. It was originally found that small amounts of nickel in a copper plus 5 per cent tin alloy refined grain to a small extent. Larger (up to 2 per cent) nickel additions did not have any

appreciable effect on grain size.

In these experiments, alloys containing copper with tin and nickel contents totaling 10 per cent were used. Limited commercial application has been made of certain alloys in this group which are considered heat-treatable by virtue of the tin-nickel compound formed. This compound apparently behaves in a fashion similar to those found in heat-treatable aluminum alloys.

The base alloy used was 90 per cent copper, 10 per cent tin. In the experiments part of tin was replaced by nickel up to 5 per cent nickel, which is in accordance with commercial practice of using a 5 per cent tin—5 per cent nickel alloy. The polished and etched samples are shown in Fig. 2, and the chemical analyses and grain count data listed in Table 1.

Study of the results of samples B-1 through B-7 shows grain refinement when nickel is present. The binary alloy, Cu plus 10 per cent Sn appears to be generally coarser than alloys where nickel has replaced part of the tin. Maximum effect is noted in B-3 and B-4 where tin contents range from 8 to 7 per cent with nickel from 2 to 3 per cent. Larger increments of nickel seem to cause coarsening of the grain until at 5 per cent Sn, 5 per cent Ni, the grain size is coarser than that of 10 per cent Sn alone. For sake of comparison, an alloy of 10 per cent Ni in copper was included (B-7) and exhibits extremely coarse grain size.

The behavior of nickel in these alloys may be explained by consideration of the intermetallic compound formed when 2 to 3 per cent nickel and 7 to 8 per cent tin are present as shown by the ternary constitution diagram² and the work of Eash and Upthegrove.³ The presence of a fine, well distributed inter-metallic compound would probably refine grain appreciably. This is apparent when B-3 is compared to B-7, Fig. 2, where the Cu-Ni solid solution has the grain size expected in either pure metals or alloys and no intermediate phases are present.

The effect of lead as a grain refiner is shown by examination of the photograph of sample B-8 where 1 per cent lead was added to an alloy of 5 per cent Sn—5 per cent Ni (similar to B-6) and refined the

grain appreciably. This is apparent not only from the grain count but also from consideration of the whole sample (B-8) where the overall grain size is much finer than that of B-6.

Further evidence of how well lead refines grain can be seen in B-9 where 1 per cent Pb has been added to an alloy of 10 per cent Sn in copper. Once again, as with B-8, the grain size of the entire bar has been refined markedly over that of B-1 which contains 10 per cent Sn and no lead.

## Effect of Lead in Commercial Tin Bronzes

Most of the commercial tin bronze alloys have low permissible lead contents, presumably on a physical property basis. There are modifications of these alloys, however, that permit up to 1 per cent lead to be present. To determine whether this higher lead content produced any desirable grain refining effect, three experiments were conducted using:—

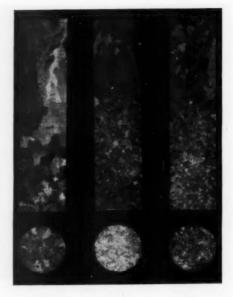
(1) SAE 65 or "Gear Bronze;" 89 Cu, 11 Sn

(2) SAE 62; 88 Cu, 10 Sn, 2 Zn

(3) Navy "G"; 88 Cu, 8 Sn, 4 Zn

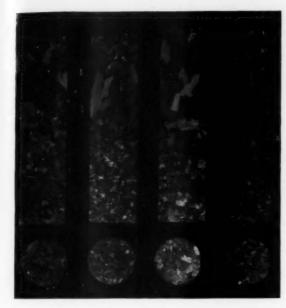
In the test with SAE 65 alloy, lead was added up to 0.50 per cent. The results are shown in Fig. 3 and listed in Table 1. It can be seen that 0.25 per cent Pb has strongly refined the grain of the lead-free alloy. No great improvement is apparent with increased lead up to 0.50 per cent. Current specifications permit 0.50 per cent lead so that whenever possible, from the basis of having fine grain in the casting, at least 0.25 per cent lead should be present.

The experiments with Navy "G" and 88-10-2 alloy indicated that lead is not as important as a grain refiner when zinc is present in the alloy. In both alloys lead was added up to 1.0 per cent. The photographs



	Cl	C2	C3
Cu	Rem	Rem	Rem
Sn	11.3	11.4	10.9
Pb	0.0	0.22	0.51

Fig. 3—Macroetched sections showing the effect of lead on the grain size of SAE Alloy 65 (89 Cu-11 Sn)



	DI	D2	D3	D4
Cu	Rem	Rem	Rem	Rem
Sn	10.2	10.2	10.2	10.3
Zn	2.06	1.95	1.94	1.99
Pb	0.0	0.24	0.52	1.05

Fig. 4—Macroetched sections showing effect of lead on the grain size of SAE Alloy 62 (88 Cu-10 Sn-2 Zn).

of the polished and etched specimens are shown in Figs. 4 and 5 with the chemical analyses in Table 1. Some grain refining effect can be seen but it is of much less significance than was the effect of lead in zinc-free tin bronzes. Consideration of the photographs indicates that the alloys containing 1 per cent lead have somewhat finer grain size than do the lead-free alloys. Lower lead contents have lesser effect but still do influence grain size.

Specifications for these two alloys usually permit up to 0.25 per cent or 0.30 per cent lead. Even this amount has some beneficial effect on grain size although the modified alloys that contain 0.90 per cent lead ("Commercial 88-10-2" and "Navy PC") would probably have finer overall grain size in castings. The higher lead content may reduce physical properties somewhat although inappreciably. The presence of lead improves the machining characteristics of the alloy greatly.

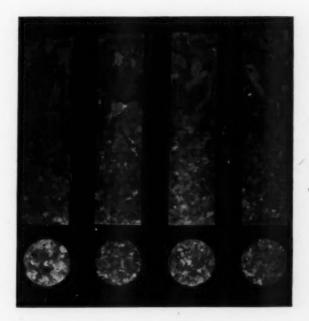
## Discussion

It has been demonstrated in these experiments, as well as those reported previously, that lead has considerable influence on the grain size of cast alloys. Also, the nickel-tin intermetallic compound apparently exerts influence on grain size under certain conditions. In the earlier experiments it was found that tin alone in copper was a stronger influence on grain size than was zinc in copper.

On the basis of these observations it would appear that those things which are not in solution in copper will have the strongest effect on grain size. Lead has low solid and liquid solubility in copper. In this case the effect on grain size may be primarily one of restricting grain growth mechanically during solidification.

Final grain size of a cast metal or alloy depends on two factors—the number of grains that nucleate from the liquid and that rate at which those nuclei or incipient grains grow. If many nuclei are present, finer grained structures may result. Also, if fewer nuclei form, fewer grains will result regardless of whether the cooling rate is fast or slow. Obviously it is difficult to separate the factors in cast alloys but those things which improve nucleation or restrict growth of individual grains (such as impingement of adjacent grains growing from a melt) should promote finer grain size of the cast material.

If the above hypothesis is valid, grain size control (beyond that exercised by cooling rate) in cast alloys will be governed by the manner in which additive elements are present. Those elements which are completely miscible should have minimum effect on grain size. As seen from copper-nickel and copper-zinc alloys, this is apparently the case. With elements out of solution, or present as compounds which may increase the rate of nucleation, greater effect on grain size would be expected. This has been demonstrated. Further experiments exploring the effect of other compound-forming elements or elements with low solid solubility in copper should reveal whether this hypothesis can be used as the basis for deliberate influence of grain size by composition control.



	El	E2	E3	E4
Cu	Rem	Rem	Rem	Rem
Sn	8.6	8.5	8.6	8.3
Zn	3.97	3.95	3.90	3.94
Pb	0.0	0.22	0.56	1.05

Fig. 5—Macroetched sections showing effect of lead on grain size of "G" metal (88 Cu-8 Sn-4 Zn).

# Summary and Conclusions

A group of experiments have been conducted to determine qualitatively the effect on grain size of lead in cast copper, and nickel in a commonly used coppertin alloy. The effect of lead in this alloy and a series of commercial copper-tin alloys has also been investigated. An experimental method which minimizes cooling rate as a variable, has been used. From the results of these experiments it is possible to draw the following conclusions:-

1. Lead in pure copper in excess of 0.25 per cent has considerable grain refining influence. Smaller amounts of lead, in excess of the solid solubility in copper, may have a small effect on grain size.

2. In an alloy of copper and tin, nickel contributes to the refining of the cast grain size when the tin content is about 7 or 8 per cent and the nickel 2 or 3 per cent. This may be related to the formation of the nickel-tin intermetallic compound which makes alloys of this type heat treatable.

3. The presence of 1 per cent lead in a copperbase alloy of 5 per cent tin and 5 per cent nickel, refines the grain as compared to a lead-free alloy.

4. Lead refines the grain of several commercial copper-tin alloys, although the effect is greater in binary copper-tin alloys than it is in copper-tin-zinc alloys.

#### References

1. R. A. Colton and M. Margolis, "Effects of Certain Elements on Grain Size of Cast Copper-Base Alloys, "A.F.S. TransACTIONS, vol. 59, pp. 360-371 (1951).

2. Metals Handbook, 1948, p. 1262, American Society for Metals, Cleveland.

3. J. T. Eash and C. Upthegrove, "The Copper-rich Alloys of the Copper-Nickel-Tin System," Transactions, AIME, vol. 104, p. 221 (1933).

#### DISCUSSION

Chairman: W. A. MADER, Oberdorfer Foundries, Inc., Syra-

Co-Chairman: B. W. SCHAFER, Kuhlman Electric Co., Bay City, Mich.

Recorder: C. P. Kotowicz, Ampco Metal, Inc., Milwaukee. ROBERT KATTUS: 1 Were any deoxidizers used in the coppertin alloys the author touched upon?

MR. COLTON: Nothing was used as a deoxidizer because of melt conditions being limited to very small heats and being handled under exact melting and pouring temperatures.

MR. KATTUS: Was the top of the test bar cooled more slowly? MR. COLTON: The bottom half of the bar was used for test

purposes since it was felt that that part would be more solid. G. P. Halliwell: ² When you mention lead analysis, how did you arrive at the lead determination?

MR. COLTON: We arrived at lead determination by taking drillings along the length of the test bar. Never did we encounter lead segregation.

MEMBER: How long did it take from the molten stage to cool? Mr. Colton: It took 30 degrees per minute.

MEMBER: Why was 2100 F selected?

MR. COLTON: Because of the convenient temperature and everything dissolved and mixed best at that temperature.

B. N. Ames: a Superheating has no effect at all on grain size when melted properly. In tin bronze alloys we want larger grain

¹ Senior Metallurgist, Southern Research Institute, Birmingham, Ala.
² Director of Research, H. Kramer & Co., Chicago.
³ Senior Metallurgist, New York Naval Shipyard, Brooklyn.

# EFFECT OF SULPHUR ON FLUIDITY OF GRAY CAST IRON

By

Lew F. Porter* and Philip C. Rosenthal*

#### ABSTRACT

A series of preliminary experiments was made to develop a laboratory melting technique that would reveal the effects of sulphur. These tests showed that sulphur is detrimental to fluidity only when combined with sufficient manganese to form sulphides high in manganese content. Furthermore, it was found that pouring temperature was an important variable in altering the influence of sulphur.

On the basis of the preliminary experiments, a comprehensive program was designed to study the effect of sulphur on phosphoric gray iron poured from 2400, 2500, 2600, and 2800 F. It was found that no significant decrease in fluidity occurred at 2400 F regardless of sulphur content, but that sulphur did have an effect at 2500 F and above. It appears that the amount of sulphur which can be tolerated at these higher pouring temperatures without adversely affecting fluidity increases with pouring temperature. However, the higher the pouring temperature, the more rapidly will sulphur contents over these threshold value decrease fluidity. A study of size, shape and distribution of sulphides within the spiral showed that no relationship existed between sulphide size and shape and their effect on fluidity, nor could the segregation of sulphur which occurred be correlated with decreased fluidity.

# Introduction

Most gray iron foundrymen agree that high sulphur contents will adversely affect fluidity. Yet, only a few systematic investigations of sulphur have been recorded. That the effect is elusive and difficult to study is reflected by the disagreement found in the literature. For instance, Smith and Riggan¹ concluded that sulphur did not affect fluidity. Holtby and Dowdell² concluded that up to 0.18 per cent sulphur, fluidity was not affected, but at that point fluidity was dractically reduced. Levi,³ on the other hand, showed that sulphur as low as 0.137 per cent affected fluidity since he obtained a 33½ per cent increase in fluidity by lowering sulphur from 0.137 to 0.086 per cent using 6.4 lb of purite per ton of metal as a desulphurizing treatment.

Because of the apparently large yet inconsistent effects reported, it was felt that a detailed investigation of the influence of this element was warranted. A fluidity formula and chart presented by the authors in a previous paper greatly aided the investigation. In order to evaluate the effect of sulphur, it was only necessary to compare the fluidity obtained using the fluidity mold described in the above mentioned paper with that indicated by the chart for the particular analysis and pouring temperature involved.

#### Procedure and Results

Before designing a detailed program, a preliminary investigation was conducted to find out if reduced fluidity due to sulphur could indeed be observed under laboratory melting conditions and, if so, what conditions were necessary.

The four base irons used in the preliminary investigation were selected to cover a wide range of analysis. Their analysis is as follows:

		Composition, %					
		T.C.	Si	Mn	P	S	
1.	Malleable Iron	2.2	1.0	0.30	0.18	0.10	
2.	Phosphoric gray iron	3.2	2.4	0.35	0.74	0.08	
3.	50-50 mixture of 1 and 2 Armco iron with		1.7	0.32	0.46	0.09	
2.	added C, Si and P	3.0	2.0	0.01	0.60	0.014	

The first three irons were melted with iron sulphide added to the charge to increase the sulphur content to 0.18 per cent. After melting, the liquidus temperature was determined using a procedure previously described,⁴ and the iron heated to 350 F superheat. Temperature was measured with a silica-protected platinum-platinum, 10 per cent rhodium thermocouple. None of the fluidity tests poured from this temperature showed any decrease in fluidity from that predicted by the fluidity chart. In fact, if anything, there appeared to be a slight increase of from 1 to 2 in. A higher sulphur addition which increased the sulphur to 0.26 per cent was made to the malleable iron base metal, and even at this value a slight increase in fluidity was noted.

Next, the effect of very low sulphur was investigated by using the fourth base iron which consists of armco iron plus added carbon, silicon, and phos-

^{*}Of the authors, Lew F. Porter is a research metallurgist in the University of Wisconsin Engineering Experiment Station, assigned to the Department of Mining and Metallurgy, and Philip C. Rosenthal is a professor of metallurgy in that department.

phorus. Here again, although the sulphur was reduced to 0.014 per cent, no appreciable improvement in fluidity resulted over that calculated from the

fluidity chart.

Since sulphur in itself did not appear to reduce fluidity, and since it is known that increasing the manganese content will raise the temperature at which the sulphides precipitate from solution, it might be expected that the formation of manganese sulphide* inclusions in the molten iron is responsible for the decreased fluidity associated with high sulphur contents. Therefore, heats were made in which manganese as well as sulphur was added to the charge. When these metal charges were melted, large amounts of an insoluble constituent formed on the surface. This material was removed and the standard procedure of determining the liquidus temperature and superheating 350 F were carried out. Again, no reduction occurred in the fluidity of the metal. However, the analysis showed that the sulphur content had not been increased more than a few points by the addition, and the recovery of manganese was much poorer than had been anticipated. Apparently, the manganese sulphide formed and floated to the surface when the iron melted, thus producing the constituent noted.

#### Retain Manganese and Sulphur

In an attempt to keep the manganese and sulphur in the melt, a gray iron heat was made in which the iron sulphide and ferromanganese were not added until the 350 F superheat temperature (2440 F) had been established. This time less of the insoluble constituent floated to the surface after the addition. This material was removed and the fluidity test poured. Once again, no decrease in fluidity was noted in spite of the fact that the sulphur content of the metal analyzed 0.162 per cent, and 0.92 per cent of manganese was present.

Finally, through still another alteration in melting procedure, a reduction in fluidity from sulphur was obtained. Again, a gray iron base metal was melted, and this time 32 grams of 80 per cent ferromanganese were added to the 8-lb charge. When the charge had melted and the liquidus temperature had been determined the metal was heated to 2800 F. At this temperature 30 grams of iron sulphide were added and the temperature reduced to 2600 F. The fluidity test was poured from 2600 F and a definite decrease in

fluidity of 31/2 in. was obtained.

The sulphur content of the heat was 0.182 per cent and 0.80 per cent manganese was present. An additional number of heats was made to check this result and the following melting technique was established:

1. An 8-lb charge of the desired scrap was melted by induction in a magnesia crucible, the surface of the metal skimmed free of slag, and the melt superheated to a temperature of 2700 F. 2. At this point the desired amount of sulphur was added as iron sulphide. The metal was held 10 min after the iron sulphide addition with the power on.

3. After the 10-min holding period the temperature had reached 2850 F, and 30 grams of 80 per cent ferromanganese were added. This addition raised the manganese content to about 1 per cent and was sufficient to form manganese sulphide with sulphur present in amounts up to 0.60 per cent. The heat was held for 5 min after the manganese addition. After this holding period the temperature was generally over 3000 F.

4. The furnace power was then turned off and the metal cooled to the desired pouring temperature. Samples for analysis were taken from the molten metal at the pouring temperature by drawing them up into a length of pyrex glass tubing using an aspirator bulb to create the necessary suction. The fluidity test was then poured.

The preliminary tests, besides developing a melting technique which was capable of exhibiting reduced fluidity due to sulphur, brought to light a number

of important points:

First, sulphur in itself is not necessarily detrimental to fluidity but must be combined with manganese to form manganese sulphide.

Second, most of the manganese sulphide will float out of the metal at very low temperatures and can be

removed with the slag.

Third, temperature is an important variable, for even though considerable manganese and sulphur were present in the metal, no reduction in fluidity occurred when the pouring temperature was low (2440 F).

Finally, the conditions which were found to develop

TABLE 1—EFFECT OF SULPHUR ON FLUIDITY OF PHOSPHORIC GRAY IRON

Heat	Pouring	C	omposi	tion, %		Sulphu		
No.	Temp., F	C	-si	P	Mn	Metal	Well	Tip
1	2400	3.09	2.30	0.72	1.02	0.08	0.08	0.25
2	2400	3.17	2.35	0.71	0.80	0.16	0.15	
3	2400	3.18	2.28	0.71	0.64	0.22	0.22	
4	2500	3.17	2.38	0.79	1.06	0.10	0.10	
5	2500	3.31	2.42	0.78	1.02	0.14	0.14	
6	2500	3.22	2.33	0.76	1.00	0.155	0.155	
7	2500	3.15	2.27	0.74	0.97	0.18	0.19	
8	2500	3.13	2.23	0.73	0.94	0.21	0.21	
9	2500	3.13	2.20	0.72	0.83	0.22	0.23	
10	2500	3.08	2.08	0.72	0.67	0.26	0.28	
11	2600	3.07	2.24	0.67	1.08	0.11	0.11	
12	2600	3.16	2.31	0.62	1.03	0.18	0.15	0.25
13	2600	3.15	2.34	0.64	0.95	0.19	0.18	
14	2600	3.15	2.33	0.62	0.91	0.26	0.23	0.42
15	2600	3.07	2.21	0.67	0.80	0.29	0.20	
16	2600	3.06	2.25	0.63	0.77	0.30	0.20	0.43
17	2600	3.11	2.13	0.66	0.80	0.31	0.33	
18	2800	3.05	2.36	0.74	1.02	0.08	0.08	
19	2800	3.05	2.39	0.70	0.83	0.23	0.16	
20	2800	2.08	2.30	0.72	0.83	0.23	0.16	
21	2800	3.14	2.32	0.69	0.80	0.25	0.17	
22	2800	3.14	2.31	0.74	0.73	0.30	0.145	
23	2800	3.09	2.28	0.72	0.90	0.35	0.35	
24	2800	3.06	2.30	0.72	0.80	0.39	0.39	
25	2800	3.03	2.25	0.70	0.88	0.40	0.41	
26	2800	3.08	2.26	0.72	0.91	0.42	0.41	

[•] It should be recognized that the term "manganese sulphide inclusions" does not necessarily refer to inclusions chemically identifiable as pure manganese sulphide but, rather, to sulphides high in manganese. Likewise, when the effects of sulphur are mentioned, it will be understood, unless specifically stated otherwise, that these effects occur only in the presence of sufficient manganese to insure the formation of sulphides high in manganese.

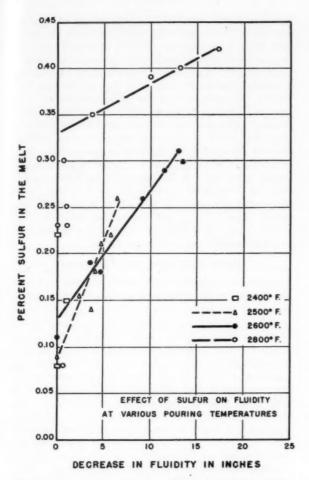


Fig. 1—Effect of sulphur on fluidity at various pouring temperatures when sufficient manganese is present to form high-melting sulphides.

reduced fluidity from sulphur are analogous to those which might be found in commercial practice, particularly in cupola melting. Here, the iron is heated far above its melting point before entering the cupola well, setting up the necessary condition for solution of sulphur from the coke and other sources. Thus, it is believed that the laboratory practice which reveals the effects of sulphur does not represent a specialized case, but that its counterpart is found in routine foundry melting.

Using the melting technique outlined above, a comprehensive program was designed in which the effect of sulphur on the fluidity of phosphoric gray iron was studied at 2400, 2500, 2600, and 2800 F. Each heat was analyzed for carbon, silicon, and phosphorus, and the analysis was applied to the previously mentioned fluidity chart to⁴ obtain a calculated value for the fluidity at normal low sulphur contents. The difference between the calculated fluidity and the actual fluidity gave the decrease in fluidity due to the increased sulphur content. Manganese and sulphur contents were also determined for each heat. The complete analysis of the heats is given in Table 1.

In Table 2 the fluidity of each heat, based on the

carbon, silicon, and phosphorus content and determined by the fluidity chart is compared with the actual fluidity of the heat to determine the decrease in fluidity due to sulphur.

The results given in Table 2 are represented graphically in Fig. 1 by plotting sulphur content against the decrease in fluidity due to the sulphur for each of the pouring temperatures studied.

# Discussion of Results

Figure 1 shows that the effect of sulphur is different at each pouring temperature. As pouring temperature increases above 2500 F it appears that the amount of sulphur that can be tolerated without adversely affecting fluidity increases. Thus, fluidity is reduced for sulphur contents above about 0.09 per cent, 0.13 per cent, and 0.33 per cent at 2500 F, 2600 F, and 2800 F, respectively. However, once these threshold values are exceeded further increases in sulphur drastically reduce fluidity. The higher the pouring temperature, the more rapidly will sulphur contents over the threshold value decrease fluidity as is seen by comparing the slopes of the curves for 2500 F, 2600 F, and 2800 F pouring temperatures. No significant decrease in fluidity occurs at 2400 F, regardless of sulphur content.

At each of the temperatures studied, the maximum sulphur addition was 70 grams of iron sulphide per 8-lb melt. Yet it will be noted that the maximum sulphur content obtained in the melt increases as the pouring temperature increases; being 0.22 per cent at 2400 F, 0.26 per cent at 2500 F, 0.31 per cent at 2600 F, and 0.42 per cent at 2800 F. Thus, it would appear that the solubility of manganese sulphide in the iron

TABLE 2-FLUIDITY DECREASE DUE TO SULPHUR

Heat No.	Pouring Temp., F	Fluidity Obtained, in.	Fluidity Calculated, in.		Sulphur Content, % (liquid metal sample)
1	2400	231/2	231/2	None	0.08
2	2400	233/4	25	11/4	0.15
3	2400	251/8	25	None	0.22
4	2500	321/4	32	None	0.09
5	2500	303/4	341/2	33/4	0.14
6	2500	30	321/2	21/2	0.155
7	2500	27	31	4	0.18
8	2500	253/4	301/2	43/4	0.21
9	2500	243/4	301/2	53/4	0.22
10	2500	221/2	29	61/2	0.26
11	2600	35	34	None	0.11
12	2600	301/2	351/4	43/4	0.18
13	2600	313/4	351/4	31/2	0.19
14	2600	261/8	351/4	91/8	0.26
15	2600	221/2	34	111/2	0.29
16	2600	201/2	34	131/2	0.30
17	2600	21	34	13	0.31
18	2800	441/4	45	3/4	0.08
19	2800	435/8	443/4	11/8	0.23
20	2800	45	45	None	0.23
21	2800	441/2	451/2	1	0.25
22	2800	451/8	46	7/8	0.30
23	2800	411/4	45	33/4	0.35
24	2800	345/8	445/8	10	0.39
25	2800	301/2	433/4	131/4	0.40
26	2800	271/2	443/4	171/4	0.42

#### MODIFIED FLUIDITY SPIRAL

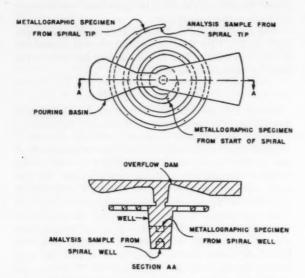


Fig. 2—Sketch of fluidity spiral showing location of analysis samples and metallographic specimens.

increases with increasing temperature; the excess manganese sulphide being rejected to the surface of the melt before the heat is poured. The rejection of excess manganese sulphide was evident during the preparation of the heats.

The analysis of the heats for manganese also indicated that excess manganese sulphide was being rejected. At low sulphur contents the manganese addition gave a manganese content in the melt of about 1.02 to 1.08 per cent; as the sulphur addition increased, the manganese content dropped, the drop being most drastic at the lower pouring temperature. However, when over 0.30 per cent sulphur was retained in the melt, the manganese content rose again and seemed to maintain a value of 1.7 × per cent sulphur + 0.20 per cent.

In the course of the investigation, sulphur analyses were made at various places in the fluidity spiral. It

was found that in some cases segregation of sulphur occurred. This made it necessary to sample the metal just prior to pouring the fluidity mold to obtain an accurate and representative determination of the sulphur content. The sample was obtained by drawing metal from below the surface of the melt as described earlier. An indication of the segregation which took place at various pouring temperatures is obtained by comparing the sulphur contents recorded in Table 1. The first value listed is the representative sulphur content obtained from the melt at the pouring temperature. The second value represents the sulphur found in the metal at the bottom of the fluidity spiral well, while the three values in the third column (2600 F pouring temperature) show the sulphur content found in the very tip of the spiral for the heats indicated. Figure 2 shows more clearly the location of analysis samples taken from the spiral.

#### Sulphur Segregation and Temperature

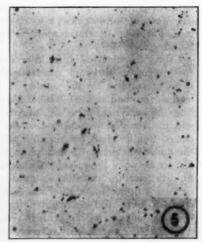
From a comparison of the sulphur in the melt and the sulphur in the spiral well it appears that no segregation of sulphur occurred at 2400 F or at 2500 F pouring temperatures. At 2600 F segregation occurred throughout the sulphur range except for the highest sulphur content. Similarly, at 2800 F the heats containing 0.30 per cent sulphur or less show marked segregation, while those of higher sulphur content do not show segregation. It is interesting to note that at 2800 F the heats without sulphur segregation are the only heats which exhibit reduced fluidity. Thus it appears that pouring temperature influences segregation of sulphur, and that at higher pouring temperatures the amount of sulphur present is important.

Photomicrographs showing the size and shape of the manganese sulphide inclusions in various parts of the fluidity spiral are given in Figs. 3, 4, and 5. These figures show inclusions found in the spiral from Heat

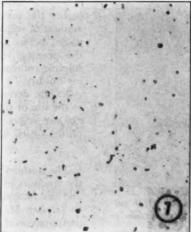
Fig. 3—Sulphide inclusions in the well of the spiral of a high sulphur, high manganese heat. Fig. 4—Sulphide inclusions at the start of the same spiral. Fig. 5—Manganese sulphide inclusions at the tip of the same spiral. All specimens unetched. 250X.











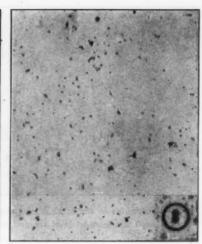


Fig. 6—Sulphides in the well of the spiral of a high sulphur, low manganese heat. Fig. 7—Sulphides in the start of same spiral. Fig. 8—Sulphides in the tip of same spiral.

17 poured at 2600 F. No segregation of sulphur was detected in this heat.

In the well of the spiral under slow freezing conditions the sulphides grow large, forming skeletal polyhedrons. At intermediate freezing rates such as the start of the spiral, the sulphides have a marked dendritic form with their polyhedral crystal habit just beginning to appear. At the tip of the spiral the inclusions are uniformly distributed, generally rounded, and extremely small, indicating rapid freezing with no opportunity for growth. From the variations in form with freezing rate, it is obvious that the sulphides precipitate out of the liquid, well above the start of freezing of the metal, and grow slowly in a dendritic manner forming polyhedral crystals. The growth is stopped off when the metal freezes, leaving either the completed polyhedron or a skeletal polyhedron as the shape of the sulphide.

# Form of Sulphide Inclusions

It is interesting to note that the sulphides present in high sulphur, low manganese heats, which do not show decreased fluidity, were small, light colored, and compact rather than dendritic. The chemical analysis and fluidity of one such heat is given below, along with three photomicrographs (Figs. 6, 8, and 8) taken from the well, start, and tip of the fluidity spiral.

The photomicrographs show that approximately the same amount of sulphide is present as in the previous series of micrographs, the major difference being in the form and color of the sulphides. The lighter color is attributed to the lower manganese content. It is also believed that the smaller size of the inclusions might in part be due to their forming at a lower temperature.

| Composition, % | Pouring Actual Calculated | C | Si | P | Mn | S | Temp., F | Fluidity, in. Fluidity, in. | 3.08 | 2.30 | 0.67 | 0.41 | 0.35 | 2600 | 35 | 341/2

Since normal fluidity was attained when small compast inclusions were present, it was thought that dendritic or skeletal inclusions might always be associated with decreased fluidity due to sulphur. However, this proved to be untrue. When high sulphur, high manganese heats were made and aluminum or titanium was added just before tapping, the inclusions were found to be small and compact, and yet the fluidity was decreased in the amount predicted from the sulphur content and pouring temperature. This was true both at 2600 F and 2800 F pouring temperatures. The typical appearance of the inclusions present at the start of the spiral is shown in Fig. 9 taken from the heat to which 0.15 per cent aluminum was added just before pouring at 2600 F. The composition and fluidity of the heat are as follows:

Composition, % Pouring Actual Calculated
C Si P Mn S Temp., F Fluidity, in. Fluidity, in. 3.12 2.25 0.74 0.77 0.31 2600 24 351/2

Thus, while there is strong evidence to show that manganese sulphide particles are responsible for the reduction in fluidity observed, the manner in which the particles act to reduce fluidity is not necessarily related to the shape and size of the sulphides formed, nor can it be correlated with the presence or lack of sulphur segregation within the fluidity spiral.

It is probable that there are at least four factors, operating in a complex manner, which account for the relationship shown by Fig. 1. These are the sulphur content of the metal, the temperature at which the sulphides begin to form, the pouring temperature, and the temperature at which the iron begins to freeze. Of these factors, the last three primarily determine how high a sulphur content can be tolerated before fluidity is reduced. What determines the slopes of the curves in Fig. 1 is not so readily apparent, but pouring temperature is certainly one of the variables involved.

How far the data presented in Fig. 1 can be extended to actual castings is problematical. The spiral test casting from which the data were obtained represents one particular set of pouring conditions. It would be hazardous at this time to suggest that all castings would behave in a relatively similar manner. Since it has been shown that the effects of sulphur on fluidity are closely associated with pouring temperature and cooling rate of the metal in the mold, it is

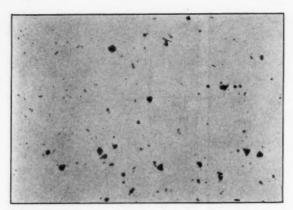


Fig. 9—Sulphides in the start of the spiral of a high sulphur, high manganese heat treated with aluminum.

conceivable that a set of circumstances might arise, particularly in a thin-section casting, where the effects of sulphur might be intensified to the point where sulphur even in the range of 0.10 to 0.15 per cent might produce greater decreases in fluidity than indicated by Fig. 1. However, when sulphur is low the amount of sulphide which can form is small, and thus the possibilities of large decreases in fluidity are minimized.

#### Conclusions

1. Sulphur in itself does not decrease fluidity but must be present in combination with sufficient manganese to form high melting manganese sulphide particles. The sulphide particles are presumed responsible for the decreased fluidity.

2. The decrease in fluidity is related to the amount

of sulphur present, and the pouring temperature. The higher the pouring temperatures, the greater is the amount of sulphur which can be tolerated (threshold value) before a decrease in fluidity occurs. The higher the pouring temperature, the greater the effect of sulphur contents over the threshold value in decreasing fluidity.

3. The decrease in fluidity is not directly related to the size or shape of the sulphide inclusions formed, nor can it be correlated with segregation of the sulphides which was found to occur.

# Acknowledgments

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# References

- 1. E. K. Smith and F. B. Riggan, "The Effect of Sulphur on the Physical Properties of Gray Iron," ASTM *Transactions*, vol. 28, pp. 205-217 (1928).
- 2. F. Holtby and R. L. Dowdell, "Effects of Sulphur on Properties of Electric Furnace Cast Iron," A.F.S. Transactions, vol. 48, pp. 308-346 (1940).
- 3. W. Levi, "Desulphurizing Cupola Cast Iron from the Practical Operating Angle," A.F.S. Transactions, vol 48, pp. 633-640 (1941).
- L. F. Porter and P. C. Rosenthal, "Factors Affecting Fluidity of Gray Cast Iron," A.F.S. TRANSACTIONS (1952) Preprint No. 52-2.

# A SIMPLIFIED ANALYSIS OF RISER TREATMENTS

By

C. M. Adams, Jr.* and H. F. Taylor**

#### General Introduction

In making castings of most commercial alloys, it is necessary to provide reservoirs of liquid metal to compensate for solidification shrinkage. To function properly, a reservoir (riser) must remain liquid longer than the volume of casting it is designed to feed. In general the object of riser treatment is to provide adequate feed metal from a riser of smallest possible volume. Toward this end there are several methods for insulating, shielding, or heating open risers; among these are use of

inert radiation shield on top of riser,
 insulating sleeve surrounding riser,

(3) exothermic material of the metal-producing type on top of a short riser,

(4) mildly exothermic or carbonaceous material on top of riser,

(5) moldable exothermic sleeve surrounding riser, and

(6) heat applied externally to riser by an electric arc or by induction.

These devices are thermal in nature, and the problem of increasing riser efficiency is essentially one of influencing heat flow. As yet, there is very little quantitative understanding of the open riser as a thermal system; it has been largely left to chance and empirical experience which of the above mechanisms are exploited for a given industrial application. This paper is a brief, simplified analysis of heat flow in risers; it is necessary to appreciate the mechanisms by which heat is dissipated from risers before an intelligent choice of riser treatment can be made for a particular job.

# Transfer of Heat from Risers by Convection, Radiation, and Conduction

Heat flows from open risers by (1) convection, (2) radiation, and (3) conduction. In evaluating heat loss, all three mechanisms of heat transfer must be taken into account. Heat leaves the exposed upper

surface of a riser by convection in the air and by radiation, and is transferred to the mold by conduction into the sand. It is of interest to know the relative importance of convection and radiation at the top surface. The two mechanisms are compared in Fig. 1, where heat flow rate per unit area (heat flux) is plotted as a function of temperature. For purposes of comparison, a typical value of 0.5 has been taken for the emissivity of the surface. The radiation curve is a reproduction of the relationship¹

$$\frac{q}{A} = 0.173 \epsilon \left[ \left( \frac{T_2}{100} \right)^4 - \left( \frac{T_1}{100} \right)^4 \right]$$

and the curve for convection reflects the equation1

$$\frac{q}{A} = 0.38 (T_2 - T_1)^{1.25}$$

where  $\frac{\mathbf{q}}{\mathbf{A}}$  = rate of heat flow per unit area,

 $\epsilon = \text{emissivity} = 0.5$ 

T₂ = temperature of the riser surface, °F Absolute

 $T_1$  = temperature of the surroundings (room temperature)  ${}^{\circ}F$ 

Absolute, and 0.173 and 0.38 are empirical constants. • The flow of heat from a riser will be the sum of the radiation and convection components. Figure 1 indicates convection may make a significant contribution to the heat flow from an aluminum riser, but is almost negligible (compared to radiation) at the temperature of molten steel. In copper-base alloys, convection heat flow to the air amounts to a little over 10 per cent of the total. Since any material used as a radiation shield will usually markedly reduce the effect of convection, the word "radiation," in the following paragraphs, will actually mean the sum of radiation and convection. In calculating numerical radiation losses, it is necessary to know the emissivities (or reflectivities) of the radiating surfaces. It has been assumed that any hot metal riser exposed to

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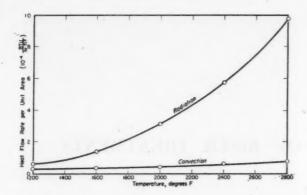


Fig. 1—Radiation and convection from exposed open-top risers. Heat loss as a function of temperature.

the atmosphere will develop an oxide layer; the emissivities of some metals in the oxidized condition at high temperatures are given by McAdams:¹

steel,  $\epsilon = 0.8$  copper,  $\epsilon = 0.6$  aluminum,  $\epsilon = 0.2$ 

Heat flow into the mold is a process of conduction. The mechanism of conduction of heat in an aggregate such as sand is complicated, involving radiation from grain to grain, flow of gases, and conduction through the grains. However, A.F.S. No. 80 molding sand has been found experimentally, at M.I.T., to act as if it were a continuous medium having the following properties—

Thermal Conductivity, k = 
$$0.57 - \frac{BTU}{(Hr) (Ft) (^{\circ}F)}$$
  
Volumetric Specific Heat,  $\rho C_p = 28 - \frac{BTU}{(Ft)^3}$ 

If, instead of sand, a gypsum insulator surrounds the riser, k=0.30 and  $\rho C_p=13.1$ 

No simple comparison between conduction and radiation can be made, since each depends in a different way upon size, geometry, and solidification time.

#### Analysis of Unshielded Steel Risers in Sand

Since the heat flow pattern in a system such as a radiating riser is dependent upon both size and shape, it is necessary to select a shape and indicate specifically what effect size will have upon it. The geometry chosen for this purpose is a right circular cylinder, of height equal to its diameter. Heat is assumed to flow from this riser by (1) radiation from the top surface, and (2) lateral heat flow into the sand. Vertical heat flow in the sand is neglected, and no heat is considered to enter or leave the riser through the bottom surface, i.e., no exchange of heat between the riser and the casting. (This is somewhat the same as saying that the base of the riser is the thermal center of the system.) Actually, heat will enter the riser by convection in the liquid metal during early stages, and leave later with the feed metal. It is further assumed any temperature

differences existing within the riser metal are small compared to the total temperature drop between the metal and the surroundings (air or sand).

The results of mathematical analysis for heat flow from 8-in. x 8-in. and 4-in. x 4-in. steel risers, in sand, are best illustrated graphically in Fig. 2. Details of the computations involved in the solution of this problem have been included in Appendix I. Heat flow rate by conduction, and by radiation, are plotted independently as a function of time. All calculations are based upon the time required for complete solidification of the riser. The curves show that when metal is first poured into a cold mold, heat flow into the mold (conduction) is very rapid, and decreases with time. Radiation heat flow is more or less constant during solidification. In other words, the sand saturates progressively with heat, while the air does not. It is for this reason that size is important; with larger risers and longer solidification times, radiation losses become relatively more important. This is indicated in Fig. 2 when one examines the point where the radiation and conduction curves intersect. Radiation "overtakes" conduction at a much earlier relative time in the large riser. The numerical results for this case are presented in Fig. 3. The large riser loses 55 per cent of its heat by radiation, and the smaller one loses 42 per cent. The radiation e

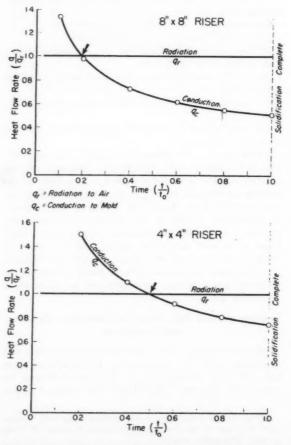


Fig. 2—Size effect on radiation losses. Heat flow rate vs. time.

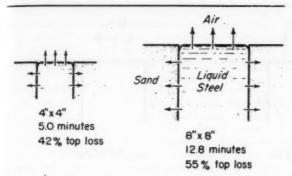


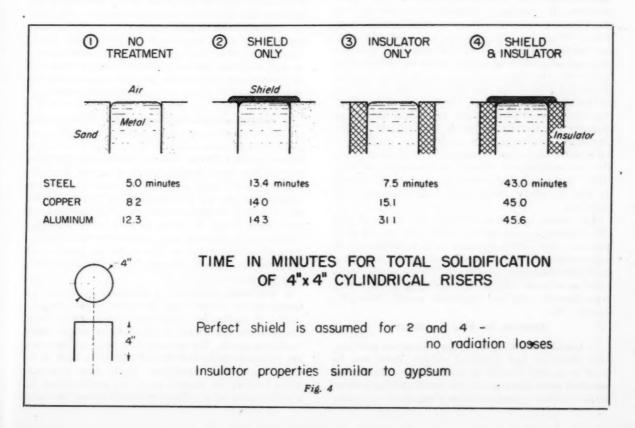
Fig. 3—Size effect on radiation losses. Time for total solidification of 4-in.x4-in. and 8-in.x8-in. cylindrical risers (steel). Without radiation the ratio of times for solidification would be 4:1. With radiation the ratio of times is less than 3:1, showing that the larger riser suffers relatively more from radiation losses. 42% of total heat is lost from top of 4-in.x4-in. riser. 55% of total heat is lost from top of 8-in.x8-in. riser.

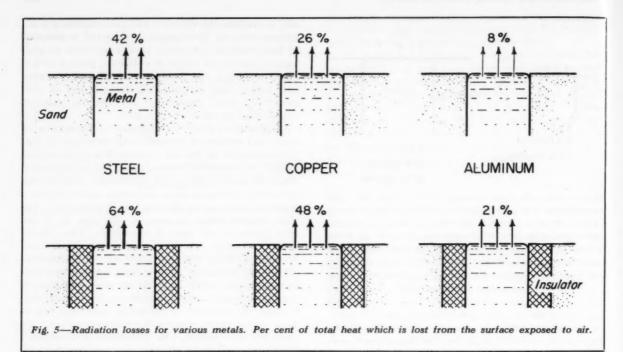
fect upon the relative solidification times is quitlarge; without radiation the ratio of times would be four-to-one, and with radiation less than three-to-one. The obvious extension of this line of reasoning is that for very large, open risers, practically all of the heat will be lost through the top of the riser, unless means are taken to prevent it.

# Freezing Temperature, Insulation, Radiation Shielding

The calculations outlined in the previous section have been repeated to include copper and aluminum, and to evaluate the effects of radiation shields and insulating sleeves. The insulating material is assumed to have thermal properties similar to those of gypsum; to accent the effect of radiation, perfect shielding is used (no radiation losses). The 4-in. x 4-in. riser is used, and the insulating sleeve is over an inch thick; i.e., thick enough so any backing material would have no effect on heat flow.

Solidification times for different combinations of metal and method of protection are presented in Fig. 4. Comparison of the first two columns shows the influence of shielding alone upon the solidification times of steel, copper, and aluminum. Due to the differences in temperature, shielding exerts almost a threefold effect upon the time for freezing steel, less than twofold effect for copper, and only a 16 per cent change in the case of aluminum. The general trend is exactly reversed when insulation is used without shielding (column 3, Fig. 4). The advantage gained by insulating a steel riser without top protection is very slight; the freezing time for copper is increased twofold, and for aluminum is nearly tripled. Clearly, the influence of radiation upon heat loss at high temperatures cannot be overemphasized. To complete the picture, the possibility of combining insulation and shielding must be considered; these results are presented in column 4, Fig. 4. The increase in solidification times over those obtained by either insulating or shielding alone are quite drastic, even for aluminum. The times in column 4 may be high by about 10 to 15 per cent, since all heat is assumed to have escaped through a good insulator; perfect shielding becomes less possible as solidification





times become longer. Column 4 may be regarded as the upper limit or goal of non-exothermic riser treatments.

Figure 5 is another method of showing the relative effects of insulation upon radiation and conduction. In any system heat will tend to flow along the path of least resistance, as does electricity or fluid under pressure. If restraint is imposed over one part of the system, more heat will flow through another part. Figure 4 indicates this behavior for insulation without shielding. Here, instead of solidification time, the per cent of total heat which leaves by radiation, is presented. Insulation increases solidification time by detouring some of the heat from the sides to the top of the riser. In other words, Fig. 5 shows that insulating the sides of risers causes more heat to be lost through the top by radiation, than would be the case if no insulation were used. Thus, shielding is even more important in insulated risers than when they are uninsulated. This also helps explain the relatively very long solidification times given in column 4, Fig. 4, where risers were both insulated and shielded.

The values (per cents or times) for insulated steel are enclosed in parentheses, since they are of comparative value only. Gypsum cannot withstand the temperature of molten steel, and no equally good substitute has been developed. An insulator under these conditions might be any fine-grained refractory material with low volumetric specific heat (Appendix II).

#### Materials for Riser Treatment

1. Inert Materials. This category comprises non-reactive insulators and radiation shields; these may be used in molded form or as loose powders. Fine grain size and open packing are the main physical characteristics which make a refractory aggregate an insula-

tor. Some materials have fine cellular structures which make them excellent for high temperatures.

Many materials (including sand) may act as radiation shields. A light cellular opaque substance is the most desirable, however.

- 2. Carbonaceous Combustible Materials. a. Cellulose Base—Some organic materials, when burned in air, produce ash which has an open structure and make an ideal radiation shield on top of risers. The heating value of these organics is probably quite small, since the reaction product is a gas which carries away most of the heat; but enough heat is usually produced to compensate for specific heat of the material, so little or none is drawn from the riser for this purpose.
- b. Amorphous Carbon or Graphite. The heating value of carbon is also limited by gaseous reaction products. Most of these materials are probably not as effective as the cellulose base in eliminating radiation because the open structure is often absent, and carbon is a fairly good conductor of heat; also oxygen supply from the air is constricted by dense packing, and burning is often incomplete. In the case of steel, there is risk of carbon pick-up in the casting.

Charcoal is different than most other combustible, carbon-base materials, in that it has an open structure; for this reason, charcoal is very effective as a riser compound.

3. Exothermic Materials. a. Producing No Metal and Containing No Carbon. Most materials in this category consist of oxidizable, granulated metals, which burn slowly when heated in the presence of air or other oxidizing agents. The temperatures and reaction rates are relatively high, but should not be so high as to cause mechanical breakdown of the base structure when used in the molded form; the product may be a solid, or a slag. These substances may be used

either in a molded sleeve surrounding the riser, or in the loose powdered condition on top of the riser.

b. Metal-Producing Type. Metal-producing exothermic materials make use of the Goldschmidt, or thermit-type, reaction, whereby a metal oxide is reduced by a finely divided reducing agent such as aluminum. The temperature of the liquid metal product is very high, and so is the reaction rate; the reaction actually adds heat to the riser, since the reaction product is hotter than the riser-metal itself. Metal-producing materials can only be used on the top of the risers, and because of the large quantity of heat evolved are particularly advantageous on short risers; this saves metal, reduces side-losses, and sets up favorable thermal gradients in the casting. There is some loss of heat during the initial flash of the reaction, but a slag cover is produced which acts as a good radiation shield. The liquid product formed insures good thermal content with the riser, and the composition of the product is not usually critical, since the high temperature causes it to "float" rather than mix with the casting.

4. Externally Applied Heat (Arc, Gas, or Induction). This method of keeping a riser open is similar in principle to exothermic treatment. Because of the fixtures required, the only probable commercial application would be the repeated manufacture of very large castings.

# Summary

The foregoing is an introductory analysis of riser treatment designed to

(1) develop an appreciation of how heat is lost from, or conserved in, risers,

(2) show the mechanism by which heat is lost as it varies with riser size and metal composition, and

(3) emphasize that there is no single panacea for all the foundrymen's risering problems. Each general type of material used for riser treatment has an important application, but no single type has an allenveloping application.

It is not intended that this do more than reflect the trend of experimental work now in progress.

# APPENDIX I

Carslaw and Jaeger² present an expression for the heat flux into the region bounded internally by a circular cylinder for the case where the region is initially at uniform temperature  $T_1$  and the interior surface is suddenly raised to temperature  $T_2$  and held constant:

$$\frac{b}{A} = \frac{4K(T_2 - T_1)}{\pi^2 R} \int_{0}^{\infty} \frac{e^{-a\theta u^2}}{u[J_0^2(Ru) + Y_0^2(Ru)]}$$

where R = radius

$$\alpha = \text{thermal diffusivity} = \frac{K}{e \, C_p}$$

 $\theta = time$ 

Jo and Yo are Bessel functions of the first and sec-

ond kind, respectively, of order zero. Jaeger and Clarke³ have tabulated numerical values of the above expression. Using the numerical values, and a process of graphical integration, a value for *n* may be obtained for use in the following approximation for the heat flux:

$$\frac{\mathbf{q}}{\mathbf{A}} = \mathbf{K} \left( \mathbf{T}_2 - \mathbf{T}_1 \right) \left( \frac{\mathbf{M}}{\mathbf{R}} + \frac{1}{\sqrt{\pi^a \theta}} \right) \tag{2}$$

To obtain the total rate of heat loss, radiation is added:

$$q = AcK (T_2 - T_1) \left( \frac{M}{R} + \frac{1}{\sqrt{\pi a \theta}} \right) + q_r$$

where  $A_c$  represents the area of the sides or sand surface of the riser, and  $q_r$  is the radiation heat loss rate. Integration then leads to:

$$Q = A_{c}K \left(T_{2} - T_{1}\right) \left(\frac{M}{R} \theta + \frac{2\sqrt{\theta}}{\sqrt{nq}}\right) + q_{r} \theta$$
(4)

where Q represents the heat content of the riser. Equation (4) may be solved for the time, which is the only unknown.  $q_r$  is evaluated by means of the radiation equation presented earlier. When the time has been determined, the last term in Equation (4) represents the total heat lost by radiation.

For solving problems on solidification of metals in refractory molding materials (poor conductors), the value of n may usually be taken as 0.47.

Figure 2, illustrating the effect of size, represents the first and second terms of Equation (3) plotted individually as a function of time.

# APPENDIX II

# Thermal Properties of Aggregate Molding Materials

There is some profit in considering what properties to look for in a material which might make it a

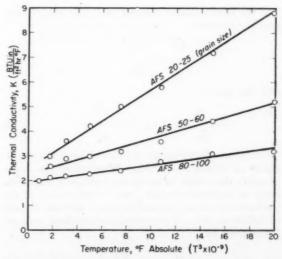


Fig. 6—Thermal conductivity of dry sand vs temperature (Data from Lucks¹)

good insulator upon exposure to molten metal. Reference to standard engineering tables of thermal properties can be quite misleading when extended to high temperatures. For example, McAdams gives the room temperature values for gypsum and dry sand:

Thermal Conductivity, 
$$\frac{BTU}{(Hr)(Ft)(^{\circ}F)}$$
 Gypsum, molded and dry,  $k=0.25$  Dry Sand,  $k=0.19$ 

At room temperature the heat is carried through an aggregate principally by conduction through the grain and through the air spaces. However, at high temperature the thermal conductivity of almost any aggregate will increase, as the mechanism of radiation from grain to grain becomes important. One may consider that the process of heat flow through a particulate medium, such as sand, may be represented by superposing two mechanisms: (1) True conduction by air and grains, and (2) regularly interrupted radiation. The radiation component may be approximated by considering the radiant flow of heat through a laminated medium, in a direction normal to the laminae. McAdams gives an expression for the radiant transfer of heat between two flat parallel sheets at temperatures T1 and T2:

$$\frac{q_r}{A} = C_1 \ (T_2^4 - T_1^4) \tag{1}$$

where C₁ is a constant.

If there is an entire series of regularly spaced sheets which are very thin, opaque, and closely spaced, Equation (1) will apply to any two consecutive sheets, and the difference ( $\Delta$  T) between T₁ and T₂ will be small, so that:

$$\frac{q_r}{A} = 4 C_1 T^3 \Delta T \qquad (2)$$

If the interlamellar distance is l, the temperature gradient becomes:

$$\frac{\Delta T}{1} = \frac{dT}{dx}$$

and this gives:

$$\frac{q_r}{A} = 4 C_1 1 T^3 \frac{dT}{dx}$$
 (3)

Equation (3) represents the flow of heat by a mechanism of regularly interrupted radiation. In order to apply this result to an air-sand mixture, radiation must be superposed upon "true" conduction. If the conductivity of the air-sand mixture, without the aid of radiation, is  $k_o$ , the conduction heat flow component is given by

$$\frac{q_c}{A} = K_o \frac{dT}{dx} \tag{4}$$

The total heat flow is obtained by adding radiation and conduction:

$$\frac{q}{A} = \frac{q_r}{A} + \frac{q_c}{A} = (4C_1 1 T^3 + K_0) \frac{dT}{dx}$$
 (5)

Thus the effective conductivity of dry sand has the form:

$$K = 4 C_1 1 T^3 + K_0 (6)$$

In Fig. 6 the thermal conductivities of three sands, determined by Lucks and co-workers,⁴ are plotted as a function of T³, illustrating the general validity of Equation (6). The quantity *l*, defined above as the interlameller spacing of the sheets, is a measure of the grain size of the material. It may be seen that the slopes of the lines in Fig. 6 are in direct relation to

the grain size.

This treatment shows the increase to be expected in thermal conductivity with increasing temperature and grain size. More important, it indicates that for materials of very fine grain size, thermal conductivity may remain almost constant with varying temperatures—in other words, very fine materials may be expected to behave like insulators compared to coarse sand when used at very high temperatures. This is a partial explanation for the behavior of gypsum as an insulator. However, thermal conductivity alone will not completely describe the behavior of a material under transient heat flow conditions.

In addition, a good riser insulator will have a low volumetric specific heat. Most refractory oxide minerals have roughly the same volumetric specific heat in the non-porous condition. However, both thermal conductivity and volumetric specific heat are reduced by open or loose packing.

#### **Bibliography**

1. W. H. McAdams, Heat Transmission, McGraw Hill Book Co., Inc., New York, 1942.

2. H. S. Carslaw and J. C. Jaeger, Conduction of Heat in Solids, Oxford University Press, London, 1947.

3. Jaeger and Clarke, Proc. Roy. Soc., Edinburgh, vol. 61 (1942), p. 229.

4. L. F. Lucks, C. L. Linebrink and K. L. Johnson, "Thermal Conductivities of Three Sands," Transactions, A.F.S., vol. 55, p. 62 (1947).

# Acknowledgment

The authors are indebted to George E. Schmidt, Jr., Research Assistant, Metallurgy Dept., M.I.T., for preparing the illustrations used in this paper, and to others of our foundry colleagues for assistance of various sorts.

#### DISCUSSION

Chairman: H. A. SCHWARTZ, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: W. S. PELLINI, Naval Research Laboratory, Washington, D. C.

Recorder: J. B. CAINE, Consultant, Cincinnati.

V. PASCHKIS (Written Discussion): 1 From the text it would seem as if the authors disregard in the mathematical treatment the existence of heat of solidification. If this is true it would distort the results.

¹ Columbia University, New York

Referring to appendix 1, equation 1, this equation holds for the following case

The riser is held at constant temperature; what is the rate of heat flow into the sand? In order to evaluate equation 1 the thermal diffusivity has to be introduced. This does not include in any place the heat of fusion. The latter acts in the same way as a temporary increase of specific heat. Since the increase is only for a limited time it would not be permissible to merely decrease the diffusivity.

The authors introduce different values for emissivity for the various materials; but they take an average value for the conductivity and the specific heat of the sand. As may be seen from the report on the writer on solidification of aluminum castings (Transactions, A.F.S., vol. 56, p. 366, 1948) the average sand conductivity in case of aluminum castings is only half the value of the average sand conductivity in case of a steel casting.

In Fig. 4 the authors indicate that they have investigated the limiting case of a "perfect shield." Introducing the shield produces on the top of the riser a similar condition as prevailing on the sides: thermal capacity has to be taken into consideration in addition to the thermal resistance. It would be possible without special effort to consider a shield more practical than the "perfect" one.

Summarizing, it seems as if the approach of the authors would be promising if they could offer proof that the various simplifications which they introduce do not create serious errors in the practical conclusions. Without such proof it seems to be dangerous to present the conclusions for the practical foundryman who may not readily be in a position to gage the significance of the various simplifications.

C. M. ADAMS, JR. AND H. F. TAYLOR (Authors' reply to Dr. Paschkis): The authors wish to thank Dr. Paschkis for his observations on the analysis.

With regard to the heat of solidification, in Appendix I, Q is defined as "the heat content of the riser" which, of course, includes the heat of solidification, as well as the superheat.

Most experimental investigations of thermal conditions in sand castings indicate that mold surface temperatures are substantially constant during solidification. (e. g. see Bishop, Brandt, and Pellini, "Solidification of Steel from Sand Walls," TRANS-ACTIONS, A.F.S., 1951). Fluctuations of more than 5 per cent seldom occur. This observation might also be predicted on theoretical grounds.

Concerning the notation in Appendix I, all symbols are defined in the text.

The thermal properties which are adopted for the calculations were obtained by interpreting the solidification times found for pure copper, aluminum, and steel in sand and gypsum molds (experimental work now in progress). Since the temperature of the mold, during solidification, ranges from room temperature almost to the freezing temperature of the metal, these figures represent "average" values, strictly applicable only at some un-known intermediate temperature. The values for the sand lead to correct solidification times, and are fairly reliable. The gypsum figures are more questionable since data obtained thus far have been erratic.

The general conclusions regarding the relative influences of riser size, temperature, and method of treatment do not depend upon the accuracy of the thermal data which have been used. The authors attach little significance to the absolute solidification times obtained in the examples chosen. However, the relative or comparative values should be of some aid in helping the foundryman select or reject various devices for riser treatment by examining his particular application more closely.

CHAIRMAN SCHWARTZ: It would seem that the term "equivalent conductivity" would be more suited to the conditions of this

paper than the more general term "conductivity.

Mr. TAYLOR: The authors are agreement with the suggestion. J. GROTT: 1 Production experience with highly exothermic pipe eliminators bears out the authors' theory that top and side losses are additive. Instructions for these pipe eliminators only consider cylindrical risers. The amounts recommended in these instructions can be converted to other shapes by first, using an equivalent amount per square inch of top surface and then increasing this amount by a ratio of surface area of the sides of the riser used to that of the surface area of the sides of a cylindrical

DR. PASCHKIS: One possible serious error may arise from the authors' simplified calculations, in that they use an average value for sand conductivity. This value changes for different metals. The sand conductivity with aluminum is about half that of steel.

MR. TAYLOR: The temperature influence on sand conductivity should not cause an error of more than 5 to 10 per cent for the different metals.

H. H. HARRIS: 3 Induction heating of risers has one serious drawback, the riser freezes over and shuts off atmospheric pressure. Exothermic pipe eliminators are the best. Gypsum causes trouble if it gets mixed with the sand.

An efficient riser collar can be made of a roll of toilet tissues. Simply cut out the center with a sharp knife to the form of a truncated cone, paint with shellac to hold it together and coat with silica wash. The heat of combustion of the paper retards heat transfer and the roll of paper is an excellent insulator, equal to a material with a density of only 10 lb/cu ft. The conical shape of the riser allows for unhindered contraction.

Another efficient pipe eliminator can be made of a mixture of ground corn cobs, fuel oil and carbon dust. Such a material has been used to cover low carbon stainless risers with no trouble due to carbon pick-up on remelting. The best solution to many problems is gating and slow pouring; many gates to distribute the metal uniformly. This has proved successful in casting high alloy lead and cyanide pots. Risers are harmful in that they produce adverse temperature effects.

G. E. Dalbey: 4 Perlite sleeves have been effective in reducing the freezing time of manganese bronze propellers. Similar risers on the same casting remained at the liquidus for 20 min when in contact with sand, 50 min when in contact with a perlite

Unitcast Corp., Toledo, Ohio.
 General Alloys, Inc., Boston, Mass.
 Mare Island Naval Ship Yard, Vallejo, Calif.

# MELTING IN A HIGH SPEED REVERBERATORY FURNACE

By

# R. G. Carlson* and W. B. Scott*

The review of one foundry's experience in melting in one particular type of furnace is presented in an effort to add to the general understanding of the merits of this particular type furnace. The furnaces used were operated as seemed best suited to this foundry's needs with melt time, economy, and melt quality being the factors used to evaluate the turnace operation.

The furnace of this reference is that relatively small reverberatory furnace commonly known as a Stroman or Sklenar furnace. The specific features of this furnace that make it a distinct type are the relatively high rate of fuel input, and the charging arrangement whereby the charge material is preheated as it passed down the stack. The practice reviewed in this article is limited to the common copper-base alloys.

The schematic diagram of Fig. 1 illustrates the basic features of the furnace construction. Originally, the furnaces were equipped with a recuperator as shown in this diagram but this was later discarded, easing the maintenance man's problem at an undetermined loss in efficiency. Figure 2 shows one such unit of 600-lb rated capacity. Figure 3 shows a larger unit of 2000-lb capacity.

Strictly speaking, from an operating standpoint, melting the nominal brasses and bronzes in this type furnace presents few problems that are not already known to the working foundryman. Because of the fast melting rates, little of the charge is subjected to 2 "soaking time" as the entire bath is brought up to pouring temperature. By shortening this "soaking time," the gassed metal bugaboo of melting is minimized. Furthermore, this fast melting enables the melting department to approach the production rates required of the modernized foundry with its mechanized molding, sand handling, and shake-out systems. In return for this increase in output, the furnace operator is presented with problems of refractory maintenance and the operating control that is part and parcel of the high heat input.

Furnace atmosphere control is still important even though the high melt rates do minimize soaking. Under the conditions of direct contact of combustion atmosphere and melt, considerable care must always be taken. For the high quality metal required for the authors' shop production, we have settled on a slightly oxidizing furnace atmosphere to the extent of 1/2 per cent excess oxygen.

# Sampling Furnace Atmosphere

When the furnaces were first installed, an exhaustive study was made of the furnace atmospheres. On the 2000-lb furnace, sampling tubes were installed as shown in the diagram of Fig. 1. These sampling tubes were so arranged that each of the six locations could be checked individually but simultaneously. The actual sampling tube was a 3/8-in. quartz tube inserted through a hole drilled in the furnace wall. Secondary air was excluded from the sampling spot by asbestos packed around each tube. Sampling position No. 1, No. 2 and No. 3 (not shown) were located in line with the three burners and projected into the stack. Positions 4 and 5 were directly in the middle of the furnace front to rear and placed behind the throat arch and between the third and fourth roof arches respectively. Position 6 was at the back of the furnace approximately 12 in. in front of the burners. The gas analysis equipment was carefully

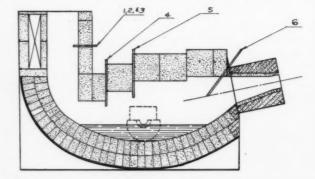


Fig. 1—Cross-sectional view of melting furnace showing basic features of furnace construction.

^{*}National Bearing Div., American Brake Shoe Co., Meadville, Pa.



Fig. 2-Showing 600-lb rated capacity furnace.

calibrated before the test operation. In theory, sampling position 1, 2 and 3 should have given the most consistent results because of their location. Actually, the quartz tubes in these locations suffered continual plugging by refractory slag and bumping by the charge. In practice, it was found that by taking readings when the melt was almost up to tap temperatures, each of the six sampling tubes could be made to give approximately the same gas analysis indicating that the overall furnace atmosphere was uniform. A cold charge in the stack could be expected to change the equilibrium ratios of the combustion gases. It was also indicated that the turbulator on each of the three burners broke up practically the entire air and oil mixture within 12 in. of the burner tips. Knowing now that the overall furnace atmosphere would be uniform as the temperatures throughout the furnace became uniform, one would expect that one long sampling tube inserted at the pouring spout and aligned separately with each burner would give an accurate reading for control purposes.

While the excess air is a general rule, an exception is made for the melting of high copper heats, i.e., 99 per cent plus copper. For this type melt, and partly due to the higher pouring temperatures required, the ½ per cent excess air atmosphere developed a troublesome excess slag. Moreover, the molten metal in the ladle had a "wild," rolling appearance which no practical deoxidizer could adequately control. A slightly reducing atmosphere on this type

heat has proven more practical, giving a cleaner furnace as well as better castings.

The operation of the furnace under these oxidizing conditions produces some slag, as must be expected. Excess slag will tend to insulate the bath lengthening the melt time and increasing costs. To minimize this effect, each heat should be carefully skimmed before tapping to eliminate the slag buildup from heat to heat and thus tend to maintain a clean furnace for subsequent charges. The furnace can be skimmed through the pouring spout if the furnace is lined properly. If through a mechanical failure in the oil pressure or because of the inevitable human element, the furnace is operated on a highly oxidizing atmosphere, or a heat is allowed to soak after normal tap time even with correct fuel ratios, a heavy metal oxide will build up on the bottom of the furnace.

A measure of the heat input of the furnace may be indicated by a melt time of 20 min on the 600-lb unit or 35 min on the larger 2000-lb unit in use at the authors' shop. This time corresponds to a fuel consumption of 48 gal of oil per hour for the larger furnace. The oil used is Grade 2.

Because of this high heat input, a high melt loss might at first seem probable. However, actual melt losses have been found to be comparably low. On a wide range of alloys from high tin gear bronze to a high lead-bearing metal such as AAR 503 medium and including hydraulic brasses such as Navy G, the authors have checked quantitative losses of 3/4 per cent to 21/2 per cent. The overall average is well below 2 per cent. On a nominal 85-5-5-5 heat in the 600-lb furnace the authors have found zinc losses to run from 0.35 per cent to 0.65 per cent which must be



Fig. 3-Showing 2000-lb rated capacity furnace.

considered low for an open-flame type furnace. A qualitative loss on an alloy approximating 80-10-10 will average 1 per cent lead loss, 1/4 per cent tin loss, with a quantitative loss of 11/2 per cent. These figures are based on charges of clean gates or ingots. Somewhat higher figures will be encountered when melting borings and turnings. The furnace has not appeared practical with the higher zinc alloys (above 5 per cent zinc) although the authors' experience has been purposely limited. Experimentally, the authors have tried a 90,000 psi tensile manganese bronze with zinc approximately 24 per cent. The zinc losses amounted to reduction of more than 51/2 per cent zinc content.

The refractory linings available for this type furnace still have much room for improvement. Many types of brick and ramming materials have been advocated and still have to be tried. Currently, the authors are getting better than average results with super duty, double burned Missouri fire clay brick. Spalling of this extremely hard brick in the roof arches was at first a serious problem. This trouble has been practically eliminated by the furnace manufacturer who has incorporated tension springs on the arch frames to insure constant, but not excessive, loading at all times. The weakest spot of the refractory lining, remains the throat arch, i.e., first arch next to the stack. This arch takes the brunt of the flame impingement plus the abrasion of heavy scrap charges. To try to overcome this problem spot, three different silicon carbide ramming mixtures have been tried. The authors have also tried two single piece silicon carbide sections, silicon carbide face brick with clay backup brick, and six different clay brick on this particular arch. The results were all comparable. The silicon carbide in any of the forms tried will take the flame impingement but the abrasive abuse of heavy scrap charges equalizes the life of any of these materials.

To summarize findings, the authors would say that this type of furnace is fast and efficient. Atmosphere control is in line with general foundry practice with other types of open flame melting units. However, the refractory material available for lining is subjected to most severe operating conditions and presents a major problem of maintenance that has yet to be solved in a satisfactory manner.

#### DISCUSSION

Chairman: W. A. MADER, Oberdorfer Foundries, Inc., Syracuse, N. Y.

Co-Chairman: B. W. SCHAFER, Kuhlman Electric Co., Bay City, Mich.

Recorder: C. P. KOTOWICZ, Ampco Metal, Inc., Milwaukee. CHESTER PETZKE: 1 The Strohman furnace is not adaptable to aluminum bronze melting because of melt losses being too high. MR. Scott: We experienced the same high lesses in the remelt

of borings and spillings in this type of furnace.

CHAIRMAN MADER: We had several reverberatory furnaces but encountered difficulty in melting aluminum bronze alloys in this type of furnace. We prefer this type of furnace for melting 85-5-5-5 and 80-10-10 alloys.

R. A. COLTON: 2 Did you arrive at oxidizing by experiments? Mr. Scott: We were not too concerned in general practice if it was either oxidizing or a reducing atmosphere. The furnaces were started every morning from a cold furnace.

L. H. King: 3 Have you melted high-nickel alloys in this type of furnace?

Mr. Scott: No, we melted 85-5-5-5 and 83-7-7-3 type alloys. MEMBER: Has anyone melted iron successfully in this type of

furnace? MEMBER: To my knowledge iron was not melted in this type of furnace successfully even though the manufacturers claim it can be done. The biggest problem encountered was the difficulty

in eliminating slag.

J. G. Winger: Our findings resulting from experiments parallel those of the authors. The slag usually takes place on the bottom of the furnace. High-nickel alloys can be melted in this furnace with good results, but conditions must be ideal.

Foundry Supt., Ampso Metal, Inc., Milwaukee.
 Research Metallurgist, American Smelting & Refining Co., Barber, N. J.
 President, King Foundry, Inc., Minneapolis.
 Foundry Supt., Reda Pump Co., Bartlesville, Okla.

# EFFECT OF PHOSPHORUS CONTENT ON GRAPHITIZATION RATE AND MECHANICAL PROPERTIES OF BLACK-HEART MALLEABLE IRON*

By J. E. Rehder**

#### ABSTRACT

The literature on the effect of phosphorus in black-heart malleable iron is reviewed and summarized, and new data are provided. With increasing phosphorus content, the ultimate strength, yield strength, and hardness increase, and elongation decreases. The ratio of yield to ultimate strength increases with phosphorus content, and the untouched and notched impact strengths are little affected up to 0.15 per cent phosphorus. The notched impact strength of heat-treated (quenched and drawn) malleable iron decreases with increase in phosphorus content.

The principal disadvantage of phosphorus in malleable iron would seem to be its action as a retardant of anneal, each doubling of the phosphorus content increasing the time necessary for first and second stage annealing by about one half and one third, respectively.

#### Introduction

Phosphorus is present in all commercial malleable irons, and is derived from the ores used in making pig iron. It is little changed in amount by the melting practices commonly used, and the only practicable way to control phosphorus content in malleable iron has been to regulate the amount of phosphorus in the raw materials.

It has been known for many years that the presence of more than about 0.25 per cent phosphorus seriously embrittles malleable iron in both the white iron and annealed state, and the practice has been to keep phosphorus content below 0.20 per cent, usually about 0.15 per cent. This was done by specifying that the pig iron contain not over a certain maximum content of phosphorus, and this is one of the important criteria of "malleable grade" pig iron.

With increasing use of duplex melting wherein relatively large proportions of steel scrap, low in phosphorus, are used in the charge, phosphorus content of the finished iron of about 0.07 per cent became common, and the questions of the effects of phosphorus in malleable iron became more important, since some control could now be had by additions of phosphorus. The work described below is an attempt to evaluate the effects of phosphorus content

in malleable iron, and to determine whether there is an optimum phosphorus content,

#### Previous Work

There has been relatively little data reported in the literature on quantitative effects of phosphorus content on graphitization rate and mechanical properties of blackheart malleable iron, and it has apparently been assumed that phosphorus has an embrittling and hardening effect on mechanical properties, with little effect on annealability or graphitization rate.

In 1922 Schwartz¹ stated that in amounts up to 0.15 per cent, phosphorus had no effect on mechanical properties, but that in amounts above 0.25 per cent the phosphorus was insoluble in the ferrite and there was then an (unspecified) harmful effect. It was further stated that phosphorus had no effect on the graphitizing process. Bean² found that phosphorus content was a factor in susceptibility of malleable iron to galvanizing embrittlement.

Kikuta³ studied the effect of phosphorus on graphitization rate of black-heart malleable iron, and concluded that phosphorus accelerates graphitization except in amounts of the order of 0.70 per cent. However, on reassessment of his complete data on phosphorus, this conclusion would not appear to be justified. Fielden⁴ in 1942 stated that higher phosphorus contents in cast iron resulted in deeper chill, but with no effect on graphitization rate. Dastur and Cohen⁵ showed that phosphorus in cast iron inhibits graphitization, but to a lesser extent than does chromium.

Schwartz⁶ in a summary paper states that phosphorus in malleable iron increases the ultimate and yield strengths with little effect on elongation. Evans⁷ has shown recently in a detailed and comprehensive study of the fluidity of cast irons that phosphorus is one half as effective in promoting fluidity, as is an equal amount of carbon.

Rehder⁸ stated that phosphorus inhibits graphitization and decreases impact strength in malleable iron. In the same paper the effects of phosphorus content on the mechanical properties of nodular iron were examined, and it was shown that phosphorus in

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^{**} Foundry Engineer, Physical Metallurgy Div., Mines Branch, Dept. of Mines and Technical Surveys, Ottawa, Ontario, Canada.

nodular iron increased the yield, ultimate strength and hardness slightly, and decreased the elongation

and impact strength appreciably.

Some of the work reported in the literature on the effect of phosphorus content on the mechanical properties of low-carbon steels will be of interest. Piggott[®] stated that phosphorus increased the ultimate strength of steel by about 1,000 psi for each increase of 0.01 per cent phosphorus up to 0.40 per cent, but with increased brittleness. Motok¹⁰ in a study of pure iron-phosphorus alloys found that the solid solubility of phosphorus in pure ferrite was over one per cent, and that phosphorus increased the hardness of ferrite in a fairly regular manner.

To summarize, it would appear generally agreed that phosphorus inhibits graphitization to some extent, and that it increases the strength and hardness of annealed malleable iron with uncertain effects on

elongation and impact strength.

# **Experimental Method**

A 500-lb heat of iron was made up from pig iron and steel scrap, with small amounts of lump ferromanganese and ferrosilicon and a few ounces of sulphur to bring the iron within normal commercial ranges of composition. The heat was melted in an induction furnace, heated to 1571 C (2860 F), cooled to 1555 C (2830 F) with power off, and tapped in five taps into a preheated 100-lb pouring ladle. The temperature of the iron in the furnace was maintained at 1555 C (2830 F), and pouring was done with promptness to avoid too long a holding period of the last iron in the furnace.

Additions of crushed 25 per cent ferrophosphorus were made to the ladle while filling, and after the full ladle was stirred and weighed, the iron was poured into the test bar molds. Tapping and pouring temperatures, and delay time, were essentially constant for each tap. A wedge sample was poured in a split copper mold from each ladle, and the wedges were eventually pulverized to provide samples for chemical analysis. Four increasing additions of phosphorus content were made, to provide five levels of phosphorus content.

One test bar casting from each phosphorus level was reserved and cut into smaller specimens in the white iron state by means of an abrasive cut-off wheel. These specimens were used for determining the times necessary for first and second stage graphitization, and the location of the top and bottom of the critical temperature range on cooling, by methods previously described by the writer.^{11,12} All of the remaining castings were then fully annealed, those of

TABLE 1-CHEMICAL COMPOSITION

		T	est Specim	ens	
	A-1	A-2	A-3	A-4	A-5
Total carbon, %	2.35				2.35
Silicon, %	1.31	1.32	1.31	1.32	1.31
Manganese, %	0.43		-		0.43
Sulphur, %	0.170	_			0.165
Phosphorus, %	0.020	0.049	0.065	0.134	0.410
Chromium, %	0.025				0.026

each phosphorus content being given appropriate annealing times according to those determined on the reserved samples.

Two thirds of the annealed castings were then reheated to 816 C (1500 F), held for ½ hr, and half of them quenched in oil, and half air-cooled or normalized. The quenched and the normalized castings were then drawn or tempered for 3 hr at 621 C (1150 F). There resulted from the above treatments three sets of test castings; one fully annealed, one oil-quenched and drawn, and one air-cooled or normalized and drawn; each set containing castings of five different phosphorus contents, but all of nearly identical chemical composition otherwise. After heat treatment the risers were cut off, Brinell hardness was determined, and the appropriate test bars machined.

The test casting used, a Y-block casting as described by Flinn¹³ and made in a baked oil-sand core, requires description. The standard ASTM tensile test bar for malleable iron, A47-48, which is specified to be tested in the unmachined condition, was not used because of the effect of variables in gating and risering, and especially in surface condition as affected by the grain size of molding sand used.

# Test Casting Described

In the present work the effect of phosphorus content on the iron itself was being sought, and other possible variables were eliminated as much as possible. The test bar used was a single-leg keel block or Y-block, which produced a usable portion ½ in. thick, about 2 in. high, and 6 in. long as-cast after cutting off the riser. These coupons were completely sound throughout, and would represent a section from a well-fed malleable iron casting of ½ in. section thickness.

Full-sized Charpy impact test specimens were, of course, obtainable from the coupons, but a standard 0.505 in. diameter tensile specimen was not obtainable. The tensile test bar used was \( \frac{5}{16} \)-in. diameter in the gage length portion, with a gage length of 2 in. The ends were threaded to \( \frac{1}{2} \)-in. diameter, and the over-all length was \( 5\frac{1}{2} \) in. All test bars used in the present work were, of course, machined from the coupons after annealing or heat treatment.

It is realized that the mechanical test results obtained from tensile test bars are affected by the diameter of the reduced section as well as by the ratio of gage length to cross-section area, and this fact renders it unwise to make direct comparison of the tensile test results obtained in the present work with those obtained from test bars of standard dimensions. However, the internal consistency should be unaffected,

TABLE 2-GRAPHITIZATION TIMES

Iron	Phosphorus Content, %		Time Necessary at 718 C (1325 F) Second Stage, hr	No. of Nodules per sq mm Sample
A-1	0.020	4.0	23	53
A-2	0.049	5.0	18	85
A-3	0.064	5.5	19	54
A-4	0.134	9.75	25	13
A-5	0.410	18.0	35	_

i.e., the changes in tensile properties with change of phosphorus content to be noted in the present work are real. The impact test results reported below should, of course, be directly comparable with other work since they are ASTM standard Charpy specimens.

From each lot of coupons, at various phosphorus levels and either annealed, normalized, or oil-quenched and drawn, two tensile specimens, four unnotched Charpy bars and four V-notched Charpy bars were machined. In addition, from each set of fully annealed specimens an extra set of Charpy impact test pieces was machined, and these were subjected to a typical galvanizing time-temperature cycle, using lead instead of zinc to prevent contamination of the samples.

Since the blocks which were fully annealed were cooled in air from 649 C (1200 F), and hence should not be subject to galvanizing embrittlement, the impact test results should show the efficacy of air-cool as a means of preventing galvanizing embrittlement at various phosphorus contents.

# Test Results

The chemical analyses of the irons obtained are given in Table 1. The liquid shrinkage of the Y-block heads was essentially uniform throughout the series, and no differences in visual character or handling of the metal were noticed as phosphorus content was increased. The as-cast white iron microstructures for four of the phosphorus levels are shown in Figs. 1 to 4, inclusive.

The results of annealing rate or graphitization time determinations are given in Table 2 and Fig. 5, along with nodule counts made on sections cut from broken fully annealed tensile test bars.

The results of the critical temperature range determinations, made at a cooling rate of 4 C (7 F) per hour, are given in Table 3 and Fig. 6.

Mechanical test results, including Brinell hardness, are given in Tables 4, 5, and 6, and in Figs. 7 to 10, inclusive. Small quantities, little more than traces,

Fig. 5—Effect of phosphorus content on time necessary for graphitization.

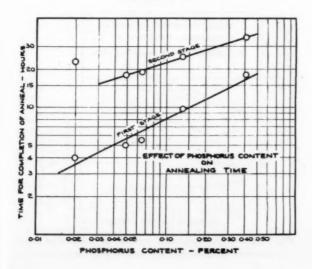


Fig. 1—White cast iron, 0.020 per cent phosphorus. X100. Etched, 2 per cent nital.

Fig. 2—White cast iron, 0.049 per cent phosphorus. X100. Etched, 2 per cent nital.

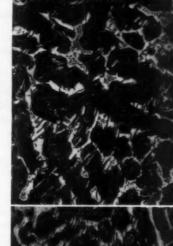


Fig. 3—White cast iron, 0.134 per cent phosphorus. X100. Etched, 2 per cent nital.

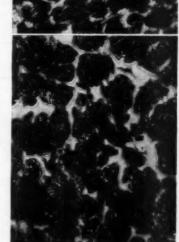


Fig. 4—White cast iron, 0.410 per cent phosphorus. X100. Etched, 2 per cent nital.

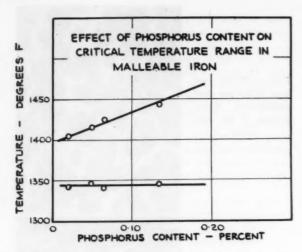


Fig. 6—Effect of phosphorus content on critical temperature range of a malleable iron.

of primary graphite were found to be present in the A-5 series (0.410 per cent phosphorus), and mechanical properties were not determined due to the uncertain effect of the small quantity of mottle. This mottle was probably the result of inoculation by the relatively heavy addition of ferrophosphorus, but is no indication of graphitization rate since even crushed ferro-chromium, as a ladle addition, will increase mottling tendency.

In Table 7 is shown the effects of phosphorus content and heat treatment on the ratio of yield to ultimate tensile strength. Table 8 shows the data on asannealed and galvanized impact properties.

The microstructures of the fully annealed samples, and of one each of the heat-treated samples, are shown in Figs. 11 to 17, inclusive.

White Iron: The chemical analyses obtained are essentially constant with respect to elements other than

TABLE 3-CRITICAL TEMPÉRATURE RANGE

	Phosphorus	Top of Range		Bottom of Range	
Iron	Content, %	°C	° <b>F</b>	°C	°F
A-1	0.020	763	1405	728	1342
A-2	0.049	768	1415	730	1346
A-3	0.064	774	1425	727	1340
A-4	0.134	784	1443	729	1345

TABLE 4-MECHANICAL PROPERTIES-FULL ANNEAL

		Ultimate	Yield				Impact ngth
Phosphorus Content Iron %		Tensile Strength psi	Strength 0.2 % offset, psi	Elong.	BHN	Un- notched, i ft-lb	V- notched ft-lb
A-1	0.020	49,100	31,700	15.0			
		45,600	31,900	13.5	119	52	12.1
A-2	0.049	50,500	33,000				
		47,700	34,000	14.0	125	60	12.5
A-3	0.064	50,600	34,200	14.5			
		47,900	34,800	13.0	126	43	11.6
A-4	0.134	50,200	38,000	11.5			
		50,200	37,600	12.5	133	48	11.75
A-5	0.041				149	-	

phosphorus, and the test results obtained should, therefore, be the result of variations in phosphorus content.

The effect of phosphorus content on the white iron microstructures, shown in Figs. 1 to 4, inclusive, is to coarsen the primary carbide structure and to consolidate the eutectic-appearing ledeburite. In comparing irons A1 and A2 (0.020 and 0.049 per cent phosphorus, respectively), a small decrease in grain size is evident, but this is probably due to the effect of the ladle addition per se, rather than that of increased phosphorus content alone.

Graphitization Rates: In Table 2 and Fig. 5 it is evident that the presence of phosphorus acts as a carbide stabilizer and increases the annealing time necessary. Approximately, each doubling of the phosphorus content increases the first stage annealing time necessary by about one half, and the second stage annealing time by about one third. It will be noted that the time for second-stage annealing of sample A-1 is not included in the line drawn for second-stage annealing in Fig. 5, and was omitted because A-1 was the only sample without ladle addition and had a low nodule count compared with sample A-2.

It is possible that a small ladle addition of 0.01 per

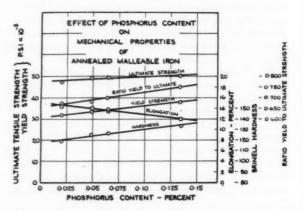


Fig. 7—Effect of phosphorus content on mechanical properties of fully annealed malleable iron.

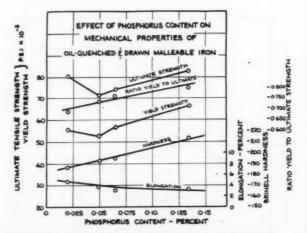


Fig. 8—Effect of phosphorus content on mechanical properties of oil-quenched and drawn malleable iron.

cent phosphorus as crushed ferro-phosphorus would be beneficial commercially in decreasing second stage annealing time, but with the disadvantage that the phosphorus content would slowly increase through the remelt until the initial advantage was lost. There is also the slight mottling tendency of any ladle addition to be considered.

The low nodule count at higher phosphorus levels would appear to be due to phosphorus content, and is probably related to as-cast grain size.

Critical Temperature Range: The top of the critical temperature range increases rapidly with phosphorus

Table 5-Mechanical Properties-Oil-Quenched and Drawn

		Ultimate	Yield			Charpy Stre	Impact ngth
P	hosphorus Content %	Tensile Strength psi	Strength 0.2 % offset, psi	Elong.	BHN	Un- otched, i ft-lb	V- notched ft-lb
A-1	0.020	81,600	56,000	5.5			
		79,200	54,100	4.0	187	32	9.3
A-2	0.049	71,400	51,200	3.5			
		72,300	54,100	4.0	193	31	7.5
A-3	0.064	70.700	52,200	3.5			
		77,500	61,100	2.5	195	35	7.5
A-4	0.134	81,600	67,500	3.5			
		83,700	64,600	3.0	213	27	5.9

Table 6-Mechanical Properties-Air-Cooled and Drawn

		Ultimate	Yield			Charpy Stre	Impact ngth
P	hosphorus Content %	Tensile Strength psi	Strength 0.2 % offset, psi	Elong.	BHN	Un- otched, n ft-lb	
A-1	0.020	70,400	41,600	5.5			
		65,100	40,000	4.5	169	25	
A-2	0.049	79,200	48,700	3.5	187	26	_
A-3	0.064	69,600	43,400	4.0			
		71,500	44,800	5.0	179	28	
A-4	0.134	79,400	49,900	3.5			
		80,800	52,200	4.0	197	22	
* 7	This set of	bars was	damaged	in mac	hining.		

TABLE 7-YIELD: ULTIMATE STRENGTH RATIO

Iron	Phosphorus Content, %	Fully Annealed	Oil-Quenched and Drawn	Normalized and Drawn
A-1	0.020	0.672	0.685	0.602
A-2	0.049	0.682	0.734	0.615
A-3	0.064	0.700	0.764	0.625
A-4	0.134	0.754	0.800	0.637

TABLE 8-IMPACT STRENGTH, AS ANNEALED AND GALVANIZED Air-Cooled from 649 C (1200 F)

		Str	y Impact rength, ched, ft-lb	Charpy Impact Strength, V-notched, ft-lb		
lron	Phosphorus Content, %	As Annealed	Galvanized	As Annealed	Galvanized	
A-1	0.020	52	50	12.1	12.0	
A-2	0.049	60	54	12.5	12.3	
A-3	0.064	43	59	11.6	12.3	
A-4	0.134	48	51	11.75	12.6	

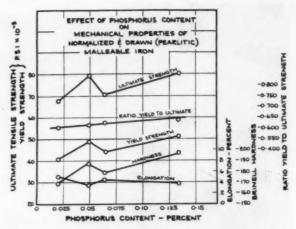


Fig. 9—Effect of phosphorus content on mechanical properties of normalized and drawn malleable iron.

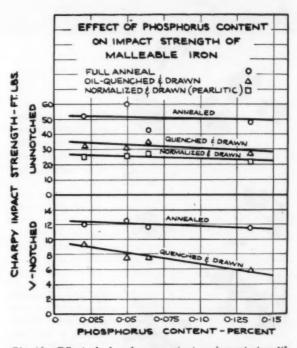


Fig. 10—Effect of phosphorus content on impact strength of annealed and of heat-treated malleable iron.

content, each increase of 0.01 per cent phosphorus content increasing the top of the range by 2 C (3 F) for these irons. There is, however, apparently little effect on the location of the bottom of the range. The practical significance of this is that with higher phosphorus content malleable irons, slow cooling for second-stage annealing must start from a higher temperature, and the total cooling time is therefore longer.

Mechanical Properties: The data presented in Table 4 and Fig. 7 show that with increasing phosphorus content the strength and hardness increase and elongation decreases in direct proportion, at least up to about 0.15 per cent phosphorus content. This in-

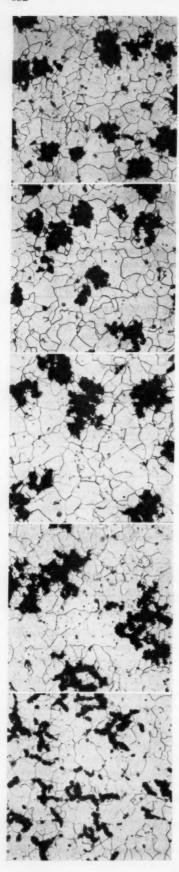


Fig. 11 — Annealed iron, 0.020 per cent phosphorus. X100. Etched, 6 per cent nital.

Fig. 12 — Annealed iron 0.049 per cent phosphorus. X100. Etched, 6 per cent nital.

Fig. 13 — Annealed iron, 0.064 per cent phosphorus. X100. Etched, 6 per cent nital.

Fig. 14 — Annealed iron, 0.134 per cent phosphorus. X100. Etched, 6 per cent nital.

Fig. 15 — Annealed iron, 0.410 per cent phosphorus. X100. Etched, 6 per cent nital.

crease in strength and hardness is equal to, or a little greater than, that contributed by the same amount of silicon. The ratio of yield to ultimate strength increases with phosphorus content. The impact strength of malleable iron, either unnotched or notched, is apparently little affected by phosphorus content up to about 0.15 per cent, as can be seen in Fig. 10.

The data on mechanical properties of oil-quenched and drawn malleable iron, given in Table 5 and Figs. 8 and 10, are less regular than those for annealed iron, but it is clear again that phosphorus increases strength and hardness, and decreases elongation. It should be noted that the ratio of yield to ultimate strength is highest for the quenched and drawn state. There is little effect of phosphorus content on unnotched impact strength, but the notched impact strength is decreased by phosphorus.

In the normalized and drawn (pearlitic malleable) state, the effects of phosphorus content are similar to those in the quenched and drawn state, except that the strength values are at a lower level for similar hardness. It is apparent that the best combination of mechanical properties in malleable iron is obtained in the quenched and drawn state, which is also true of nodular iron.

As will be seen in Table 8, phosphorus contents up to about 0.15 per cent have negligible effect on the brittleness, or impact strength, of annealed malleable iron that has been air cooled from about 649 C (1200 F) before galvanizing, indicating that this treatment is effective as a prevention of galvanizing embrittlement.

Microstructures: Phosphorus has a marked effect on the shape of temper carbon nodules obtained on annealing, as is evident in Figs. 11 to 15, inclusive, the nodules becoming progressively less well-formed as phosphorus content increases, until at 0.41 per cent the nodules are more nearly coarse flakes. The mechanical properties, especially ductility and impact strength, of the A-5 iron containing 0.41 per cent phosphorus are likely to be very poor. No evidence of phosphorus concentrations or of decomposed phosphide eutectic, even at 0.41 per cent phosphorus, was detected in the fully annealed irons.

The microstructures of the normalized and drawn, and quenched and drawn, samples in Figs. 16 and 17, respectively, are typical and what one would expect from the treatment.

# Summary

Phosphorus is an unavoidable constituent in malleable irons, but the amount is under some control especially where the iron is duplex melted with considerable amounts of steel scrap in the charge. There are some who advocate adding phosphorus in such latter case, and so a summary of advantages and disadvantages of low phosphorus content (0.08 per cent) vs. higher phosphorus content (0.16 per cent) may be of interest.

First, however, the question of molten metal fluidity should be discussed since this is one effect of phosphorus supposedly well known. The effect of phosphorus in promoting fluidity is, however, gen-

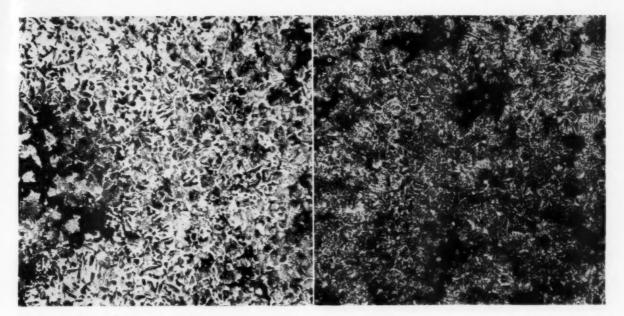


Fig. 16—Normalized and drawn malleable iron, 0.064 per cent phosphorus. X100. Etched, 2 per cent nital.

Fig. 17—Oil-quenched and drawn malleable iron, 0.064 per cent phosphorus. X100. Etched, 2 per cent nital.

erally over-estimated, and Evans⁷ has shown recently that phosphorus is about one half as effective in promoting fluidity as is an equal amount of carbon.

In comparing phosphorus contents of 0.08 and 0.16 per cent, the change in fluidity corresponds to that caused by a change of 0.04 to 0.08 per cent carbon, and this means closer control of carbon content than most foundries are fortunate enough to achieve. It is believed, therefore, that the effect of the mentioned changes in phosphorus content on fluidity are negligible and can be ignored.

The chief advantage of lower phosphorus content is in decreased annealing time and better impact properties on heat-treated malleable iron, and possibly less susceptibility to galvanizing embrittlement in annealed irons not suitably treated. Higher phosphorus content will improve strength, at some loss in ductility, but silicon is nearly as effective in this direction and improves annealability. Therefore, in general there would seem to be little justification in adjusting phosphorus content unless a specific end was in view, with corresponding disadvantages recognized.

#### Conclusions

From the data presented above, the following conclusions may be drawn:

- 1. Phosphorus acts as a carbide stabilizer in malleable iron.
- 2. As phosphorus content increases, the temper carbon nodule shape becomes less regular, approaching flake form at 0.40 per cent phosphorus.
- 3. Phosphorus increases the top of the critical temperature range on cooling, with little effect on the location of the bottom of the range.
- 4. As phosphorus content increases, the strength and hardness of the malleable iron increase, both in

the fully annealed and in the heat-treated states. The elongation decreases and the ratio of yield to ultimate strength increases with increasing phosphorus content.

- 5. The unnotched and notched impact strength of malleable iron is little affected by phosphorus content up to about 0.15 per cent phosphorus. The notched impact strength of heat-treated malleable iron is, however, adversely affected by phosphorus.
- 6. An air-cool after annealing, from about 649 C (1200 F), is effective in preventing galvanizing embrittlement up to 0.15 per cent phosphorus.
- 7. The best combination of mechanical properties, in the higher strength field, is obtained with quenched and drawn structures.

# **Bibliography**

- Schwartz, American Malleable Iron, Penton Publishing Co., Cleveland, 1922.
- 2. Bean, *Transactions*, American Institute of Mining and Metallurgical Engineers, vol. 59 (1923), p. 895.
- 3. Kikuta, Science Reports, Tohoku Imperial University, Sendai, Japan, vol. 15, no. 2 (1926), p. 115.
  - 4. Fielden, Iron and Steel, vol. 16 (1942), p. 214.
- 5. Dastur and Cohen, Transactions, American Foundrymen's Society, vol. 50 (1942), p. 895.
- Swartz, Transactions, American Foundrymen's Society, vol.
   (1946), p. 101.
- 7. Evans, Journal of Research and Development, British Cast Iron Research Association, vol. 4, no. 2 (Oct. 1951), p. 87.
- 8. Rehder, Transactions, American Foundrymen's Society, vol. 59 (1951), p. 501.
  - 9. Piggott, Iron and Steel, vol. 20, p. 309.
  - 10. Motok, Iron Age, vol. 146, (1940), no. 18, p. 27.
- 11. Rehder, Transactions, American Foundrymen's Society, vol. 55 (1948), p. 138.
- 12. Rehder, Transactions, American Foundrymen's Society, vol. 57 (1949), p. 173.
- 13. Flinn, Transactions, American Foundrymen's Society, vol. 58 (1950), p. 153.

#### DISCUSSION

Chairman: W. M. Albrecht, Chain Belt Co., Milwaukee. Co-Chairman: P. F. Ulmer, Link-Belt Co., Indianapolis. Recorder: L. G. OSBORNE, Lakeside Malleable Castings Co.,

Racine, Wis.

W. D. McMillan (Written Discussion): 1 In reviewing this paper, the first reaction is that too little attention has been paid to the phosphorus content of malleable iron. Perhaps this is because there was little variation in the phosphorus content with the conventional cold melt batch operation using the "malleable grade" of pig with a fairly uniform phosphorus content of 0.13 to 0.18 per cent. There was little that could be done about lowering the phosphorus and complications developed with phosphorus over 0.20 per cent.

It has become generally accepted that when the phosphorus is low, a condition that has come as the result of a high steel charge in duplexing, a little more phosphorus is needed to bring the yield strength safely above the minimum required to meet ASTM Specifications. I believe that this is true, particularly with a long anneal. Also that with a short anneal and air cooling from 1300 F there is less or no need for phosphorus above 0.10 in order to obtain satisfactory yield strength. It is true also that with the higher silicon duplex iron (1.20 to 1.40 per cent Siyield strength requirements can be met with a phosphorus con-

tent of 0.08 per cent.

It has been assumed by some that since air furnace iron runs 0.13 to 0.15 per cent phosphorus, that it is necessary to bring up the duplexed iron normally running 0.08 to 0.10 per cent phosphorus to about the same percentage in order to get the iron to

run properly.

I am pleased to see Mr. Rehder's comparative evaluation of phosphorus and carbon with respect to fluidity in the percentages which usually obtain in malleable iron and to see his recorded observations that the effect of a few points of phos-

phorus are negligible and can be ignored.

Mr. Rehder's observations with regard to the morphology of the temper carbon nodules are interesting. Certainly there is a distinct difference in the character of the nodule with increased phosphorus content. Iron 4a with a phosphorus of 0.134 per cent is fairly representative of most cold melt air furnace metal with respect to phosphorus content. It is also quite typical with respect to graphite nodule shape.

Our experience has been that with a phosphorus of 0.10 or less we may expect more compact nodules than with a phosphorus of 0.14 per cent, other factors being the same.

This paper shows that with a phosphorus of 0.02 the first stage time required is less than half that required with a 0.134 phosphorus. Since it is hardly possible commercially to obtain a phosphorus less than 0.05 per cent, it is necessary to set the annealing cycle on the basis of the highest phosphorus normally encountered. The annealing practice is not actually penalized by a 0.134 phosphorus. It is just a condition with which we are obliged to contend. No annealing cycle is successful when set up for the most favorable conditions. Provisions must be made for conditions of the poorest anealability.

Mr. Rehder has a good thought provoking paper and I hope he continues his interest in malleable iron and that he will continue to bring the fruits of his studies to A.F.S. literature.

RICHARD SCHNEIDEWIND (Written Discussion): ² Mr. Rehder has again presented a carefully worked out presentation on the influence of another factor in malleable iron. The paper is so com-

plete that there can be few comments on it.

The question is raised with regard to the nodule counts in Table 2, whether a satisfactory method of determination has been worked out. All of us simply count the temper carbon spots and count them as nodules. The table indicates the count as 53, 85, 54, 13, and no count respectively with increasing phosphorus. Photomicrographs 1 to 3 show decreasing grain size which in the absence of alloying effect would lead to the expectation of increasing nodule count.

Photomicrographs 11 to 15 might bear out this view if the temper carbon spots are examined closely. Figure 15 especially appears to have many nodules which have grown together in a

chain-like fashion.

MR. REHDER (Reply to Messrs, McMillan and Schneidewind): The comments of Mr. McMillan and Dr. Schneidewind are much appreciated, and as usual both contribute tangibly to the value

of the paper.

Mr. McMillan makes the point that phosphorus contributes to yield strength, which is mentioned in the paper, and it is true that phosphorus can be useful in this direction. However, it is accompanied by an appreciable decrease in annealability, whereas silicon, which has at least as good an effect on yield strength, helps the annealing process. The choice is then between the carbide-stabilizing effect of phosphorus and the potential motiling tendency of too high silicon content. In view of the facts that the effects of retarding elements such as phosphorus, manganese, chromium, and other trace elements such as tin are apparently additive; and that raw material supplies are becoming increasingly contaminated, it would seem advisable to proceed cautiously in the use of phosphorus and to use silicon as an aid to yield strength wherever possible.

This is, of course, accompanied by some risk, but so is the use of phosphorus, and improved annealability is an end which is not lightly to be ignored. One of the objects of doing the work described in the paper was to obtain the data which would

enable the relative factors involved to be weighed.

Dr. Schneidewind's points on nodule count are well taken, but it is probably mutually realized that nodule counts in malleable iron, at least with presently known methods of counting, are relative only, and different observers will probably obtain different counts on the same sample.

With respect to grain size, unfortunately grain size determinations by Dr. Schneidewind's excellent method described before this Society some time ago, were not made. Visual estimation over the whole area of each sample did not show large variations in grain size.

Plant Metallurgist, International Harvester Co., McCormick Works, Chicago.
 Prof. of Me. Engr., University of Michigan, Ann Arbor.

# EFFECT OF HYDROGEN ON GRAPHITIZATION

By

# F. Brown* and M. F. Hawkes**

Graphitization of Cementite. In one of the earliest studies on the effect of hydrogen on graphitization, Wüst and Sudhoff¹ in 1910 reported a most spectacular observation on the subject: that the process of graphitization of cementite was reversed by adding hydrogen. To do this they first completely graphitized white iron castings (of about 2.79 per cent C and 0.32 per cent Si) by heating them in vacuum in the first stage malleablization range, allowing the reaction

cementite  $\rightarrow$  graphite + austenite (I) to take place. They then divided these specimens into two groups, heating one group in purified nitrogen and the other in purified hydrogen for various times (5 and 10 hr) at temperatures of 1742 to 1832 F (950 to 1000 C).

The specimens heated in hydrogen showed an increase in combined carbon of 0.12 to 0.90 per cent, at the expense of the temper carbon. That this was indeed accomplished by the specific effect of hydrogen and not alone by pressure changes is attested by the observation that those specimens reheated in nitrogen did not show this reversion. The account of the heat treatment is so sketchy that the question arises whether the effect may have stemmed from oscillating temperature, a question which unfortunately cannot be answered from the published account. Their experiment has not been successfully repeated to date.

Sawamura² studying the same reaction observed that it was retarded not only by molecular hydrogen, but also by ammonia and by methane. Schwartz, et al.³ observed a similar effect by acetylene which, like ammonia and methane, decomposes at malleablizing temperatures to yield hydrogen; the extent of the effect of hydrogen was found to depend upon its concentration in the iron. Apparently if the composition and annealing temperature are such that the graphitization rate is extremely fast (i.e., if the ce-

mentite is highly unstable), the addition of an atmosphere of hydrogen does not appreciably retard the reaction.⁴

Eutectic Graphite. When cast iron solidifies, the eutectic portion may freeze either by the stable (II) or by the mestastable (III) reactions:

Reaction (II) produces most of the flake graphite seen in gray irons; and according to the work of Eash and of others, the cementite which forms by reaction (III) may immediately decompose according to reaction (I), yielding ASTM type D graphite.

Honda and Murakami⁵ noted that by bubbling hydrogen through the melt immediately before casting, the stable eutectic reaction was inhibited, permitting alternatively reaction (III) to form primary carbides in the castings. Boyles⁶ in a systematic study of the effect of hydrogen on reaction (II) showed that the inhibiting effect was proportional to the pressure of the hydrogen, just as was true for graphitization of cementite.

Cowan⁷ showed that the stable eutectic reaction could be repressed not only by molecular hydrogen, but by water vapor as well, and he was able to correlate the depth of chill in a test coupon with the

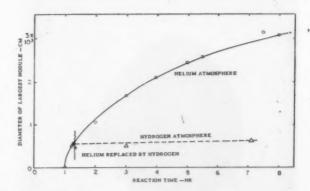


Fig. 1—Growth curves for graphite nodules in malleable iron reacted at 905 C in helium, and reacted partially in helium which was subsequently replaced by hydrogen.

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amount of moisture in the blast for the cupola in which the iron was melted.

The phenomenon of inverse chill has been considered by Zapffe and Phebus⁸ who proposed the explanation that hydrogen segregated to the center of the castings, repressed the stable eutectic reaction (II) and permitted instead the formation in the center of white iron by reaction (III). Vigorous exception was taken to the explanation, but in general the arguments in the exception do not appear to be entirely valid.

However, one of these arguments involves an observation which invites consideration here because a basic characteristic of the effects of alloying elements is involved. This observation was that when hydrogen was bubbled through molten cast iron before pouring, the graphite flakes in the castings became coarser. This actually is in keeping with the general observations of the effect of alloying additions on graphite formation both from the liquid and from the solid: that graphitizing elements increase the rate of nucleation of graphite faster than they do the rate of growth (leading to more and finer graphite particles), and that carbide stabilizers correspondingly decrease the rate of nucleation faster than they do the rate of growth (leading to fewer and coarser graphite particles).

Thus it would be expected that a limited amount of hydrogen would produce coarser flakes, just as a limited amount of sulphur does. With sufficiently high hydrogen contents or with sufficiently low basic graphitizing tendency (for example, with low silicon con-

Fig. 2—Eutectoid austenite in nodular iron partially transformed (20 min at 714C). 1000 X. Nital etch. (a) In helium: graphite, ferrite, austenite (martensite), some pearlite. (b) In hydrogen: graphite, pearlite, austenite (martensite), traces of ferrite. tent), graphite is not nucleated at all, and the iron solidifies white. It is pertinent to note that hydrogen is also found to cause the nodules in malleable iron to be larger in size and fewer in number.9

It might be a fair summary to say that probably both solidification rate and hydrogen are involved in the formation of inverse chill, and that of these the hydrogen content of the metal is the variable more readily placed under operational control.

Precipitation Along  $A_{\rm gr}$ . A further graphite-producing reaction—the precipitation of graphite along the  $A_{\rm gr}$  line on cooling—is also apparently inhibited by hydrogen, permitting instead the precipitation of cementite.¹⁰

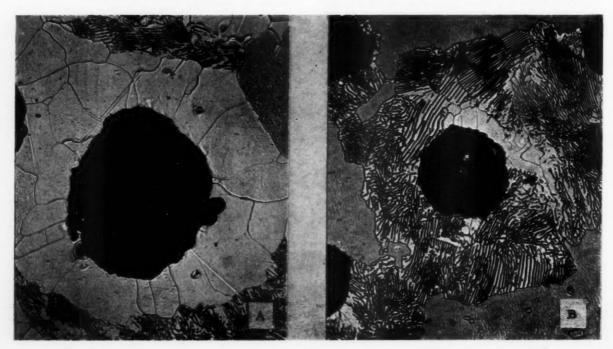
Eutectoid Transformation of Austenite. Houdremont and Heller¹³ in a broad study of hydrogen effects in steel and cast iron reported on a study of a gray iron of 2.86 per cent C and 2.1 per cent Si. One specimen of this iron was annealed at 1832 F (1000 C) for 6 hr, in air, and allowed to cool in the furnace. Other specimens were somewhat similarly treated (2 and 8 hr at 1832 F), except in wet and in dry hydrogen. The specimen annealed in air had a completely ferritic matrix. The matrix of those annealed in hydrogen, on the other hand, was either wholly or largely pearlitic. Surprisingly enough, the dry hydrogen was more effective in promoting the pearlitic matrix than was the wet hydrogen.

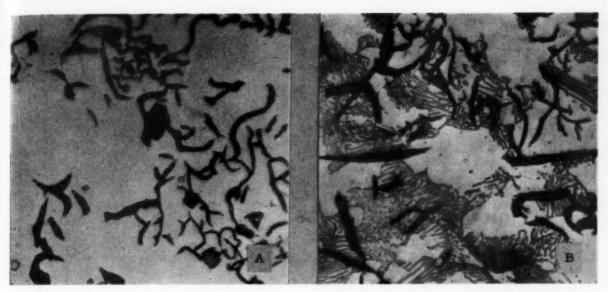
These results, though fragmentary, indicated that hydrogen inhibits the stable eutectoid transformation

 $\begin{array}{ccc} \text{austenite} & \rightarrow \text{graphite} & + \text{ ferrite} & (\text{IV}) \\ \text{and permits instead the metastable (pearlite) eutectoid reaction} \end{array}$ 

austenite → cementite + ferrite. (V)

The rate of transfer of carbon between graphite and matrix can be profoundly influenced by thermal history, probably because of altered interface condi-





tions, and such effects may have been active in these experiments, where holding times at the annealing temperature were not always the same, obscuring some of the results.

There are in the literature two further observations unrelated either to the above or to each other: first, Lorig and Samuels¹¹ observed a graphitization accelerating effect from adding hydrogen followed by prebaking (heating at subscritical temperatures prior to malleablizing) which removed the hydrogen. Although it has been reasoned that this accelerating effect took place by the formation by hydrogen of fissures which provided for easier nucleation of graphite, evidence now available indicates that this effect might better be assigned to an inherent maximum in rate of nucleation in the range at which prebaking was carried out,^{4,12} and it is debatable that hydrogen exerted any appreciable effect in this prebaking phenomenon.

Second, European literature on malleabilization frequently carries statements to the effect that hydrogen or water vapor increases the rate of malleabilization, a conclusion which may confuse the unwary; it will almost invariably develop that the reaction concerned is actually the decarburization of whiteheart malleable, not the simple decomposition of cementite to graphite and austenite.

#### **Experimental Work**

Gamma-Range Graphitization (first-stage malleablization). In endeavoring to analyze the mechanism by which hydrogen retards first stage malleablization, the rate at which individual nodules grow during this process was measured with the reaction being carried out in helium. Graphitization of another group of specimens was started in helium, after which the atmosphere was replaced by hydrogen. The graphitization thereupon stopped abruptly; not only were no more nodules nucleated, but those already formed ceased to grow (Fig. 1).

It was observed qualitatively that if the hydrogen is diluted with helium, the graphitization rate changes

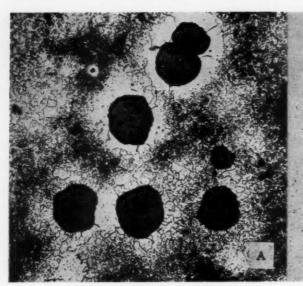
Fig. 3—Gray cast iron annealed at 900 C and furnacecooled. 1000 X. Nital-picral etch. (a) In helium: matrix entirely ferritic. (b) In hydrogen: matrix half ferrite, half pearlite.

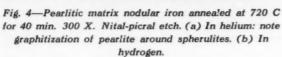
in proportion to the partial pressure of hydrogen. It was also noted that when the reaction is carried out in an atmosphere containing some hydrogen, the graphite nodules are more compact and are more nearly spherical when more hydrogen is present. It was incidentally observed that in white iron in which gamma-range graphitization was inhibited by an atmosphere of gaseous hydrogen, flushing with helium quickly acted to permit apparently normal graphitization to occur.

During studies of first stage malleablization carried out in molten lead, erratic graphitization followed the addition of protective charcoal to the surface of the lead. Subsequent experimentation demonstrated that hydrogen can rapidly diffuse through molten lead to dissolve in specimens immersed in it. Indications were that probably this hydrogen can be furnished by reduction from water vapor absorbed on the charcoal. Subsequent experimentation requiring molten reaction baths was conducted in a molten salt which was found to be neutral to the reaction.

It was not found possible to repeat Wüst and Sudhoff's observation using commercial blackheart malleable iron in one atmosphere of hydrogen.

Stable Eutectoid Reaction. For the study of this reaction (which is probably responsible for the matrix in most ferritic malleable, gray, and nodular irons), a specimen of nodular iron was austenitized in helium to give a eutectoid matrix. The specimen was then transferred to a helium filled reaction chamber at 1318 F (714 C) for 20 min, after which it was quenched. Another specimen of the same iron was treated identically except that it was austenitized and reacted in dry hydrogen. Figure 2 shows how the hydrogen has profoundly retarded the stable reaction (reaction IV), and permitted instead the metastable eutectoid reaction (reaction V) to form pearlite.





As a corollary of this it was observed that if a gray cast iron was furnace cooled from 1652 F (900 C) in helium, the matrix was entirely ferritic; but when it was given the same treatment in hydrogen, the matrix was about half pearlitic and half ferritic (Fig. 3). The same effect was observed in nodular iron similarly treated.

Subcritical Graphitization of Cementite. If cementite is held at subcritical temperatures, it may decompose by the alpha range graphitization reaction

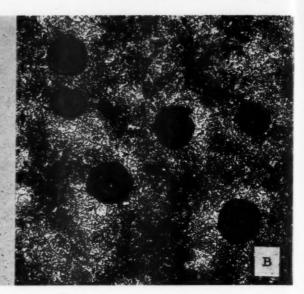
cementite → graphite + ferrite. (VI) This reaction frequently is the one used to study "second-stage malleablization," although evidence for the soundness of such an attack is yet to be published.

To study the effect of hydrogen on this reaction, specimens of the nodular iron were heat treated to produce a pearlitic matrix. Two specimens were reheated to 1328 F (720 C) for 40 min—one in helium and the other in hydrogen. The specimen treated in helium showed extensive graphitization of pearlite; the specimen treated in hydrogen showed a considerable tendency to spheroidize, but it did not graphitize perceptibly (Fig. 4).

#### Discussion

Hydrogen is thus observed to inhibit the formation of graphite (1) during solidification of the eutectic; (2) by gamma-range graphitization of cementite; (3) by rejection on cooling along the A_{gr} line; (4) by the decomposition of austenite by the stable eutectoid reaction, and (5) by the alpha-range graphitization of cementite. No accounts of studies of the effect on the other two graphitization reactions were found in the literature—the reaction to produce kish graphite and the almost trivial possibility of graphite precipitating along the ferrite-solvus line on cooling.

Although these first five reactions are always inhibited by hydrogen, they are not necessarily com-



pletely stopped by it. If sufficient instability exists because of temperature or composition, the effect of hydrogen may not be decisive in completely stopping the reaction.

Hydrogen for all these effects can be provided by many and perhaps all of the gases which contain the element—i.e., water vapor, ammonia, methane, acetylene, alcohol vapor, benzene and butane—and the effect is proportional to the concentration of hydrogen in the iron (as inferred for graphite formation from the liquid in the study of Cowan⁷ and for graphite formation in the solid in the study reported here). The effects appear to be at least largely removable from molten or solid metal by flushing with non-hydrogen-bearing gas.

Whatever may be the mechanism by which hydrogen exerts this carbide stabilizing effect, it does not similarly retard the diffusion of carbon; for it was observed in this study that in numerous white iron specimens being held at first stage malleablizing temperatures, graphitization could be entirely prevented by one atmosphere of hydrogen, but the carbon from carbides had diffused to the surface of the metal to effect a decarburization even more severe than when hydrogen was not present. It would be highly desirable, in formulating the mechanism of the carbide stablizing effect of hydrogen, to know its distribution in the various phases present, particularly in the carbides and in the graphite.

One might speculate on the possible application of the retarding effect of this element which can be added and removed so readily. If one should wish to spheroidize the carbides in a gray iron, or in a gray iron having a chill zone, for example, the possibility is suggested of carrying this out in a hydrogen-bearing atmosphere such as wet methane, which would inhibit graphitization but permit spheroidization without decarburization.

The primary graphite which sometimes forms in the center of large malleable castings, because of too much graphitizing tendency for the cooling rate afforded, might possibly be eliminated by introducing hydrogen in the molten metal before pouring; after the hydrogen had served its function to stabilize carbides during solidification, it could be removed from the solid metal by heating in a nonhydrogen-bearing atmosphere before or during malleablization.

#### References

1. F. Wüst and E. Sudhoff, "On the Effect of Hydrogen and Nitrogen on Malleable Cast Iron at Different Temperatures," Metallurgie, 7, 261 (1910).

2. H. Sawamura, quoted in "The Influence of Gases on Graphitization in Cast Iron," The Metallurgist, 6, 93 (June 27,

3. H. A. Schwartz, G. M. Guiler, and M. K. Barnett, "The Significance of Hydrogen in the Metallurgy of Malleable Cast Iron," Transactions, ASM, 28, 811 (1940).

4. Fritz Schulte, "Metallurgical Considerations on the Gaseous Annealing of Whiteheart Malleable," B.C.I.R.A. Journal of Research and Development, 3, 883 (1951).

5. K. Honda and T. Murakami, "On Graphitization of Fe-C Alloys," Journal, Iron and Steel Institute, 102, 11, 287 (1920).

 Alfred Boyles, "Influence of Atmosphere and Pressure on Structure of Iron-Carbon-Silicon Alloys," AIME Transactions, 135, 376 (1939).

7. R. J. Cowan, "Effect of Moisture in Blast Air," Iron Age, 150, Nov. 12, p. 43; Nov. 19, p. 46 (1942).

8. C. A. Zapffe and R. L. Phebus, "Cause and Cure of Inverse Chill and Hard Spots in Cast Iron," ASM *Transactions*, 41, 259 (1949).

9. H. G. Hall, "Graphitization in the Malleable Iron Process," Foundry, 77, Ncs. 6 and 7 (1949).

10. Fritz Schulte, private communication.

11. C. H. Lorig and M. L. Samuels, "Some Effects of Hydrogen on the Time of Malleablization," A.F.S. GRAPHITIZATION Symposium (1942).

12. C. O. Burgess, "Discussion to Reference 11."

13. E. Houdremont and P. A. Heller, "Hydrogen as an Alloying Element in Steel and Cast Iron," Stahl und Eisen, 61, 756 (1941).

#### DISCUSSION

Chairman: F. T. McGuire, Deere & Co., Moline, Ill.

Co-Chairman: W. A. HAMBLEY, Chas. A. Krause Milling Co., Birmingham, Mich.

Recorder: C. T. MAREK, Purdue University, Lafayette, Ind. A. L. DE Sy (Written Discussion): 1 In a recent paper dealing with the theory of graphite spherulite formation in nodular iron, we have considered the possible effect of gases; all known elements enhancing graphite spherulite formation have a relative low boiling temperature and are or may be present as minute gas bubbles in the melt.

Of course gas bubbles will preferentially be adsorbed on the surface of solid particles in suspension in the liquid iron. If such a particle is a graphite nucleus, this nucleus will be isolated or vitiated by the gaseous component or its condensation product,

thus forcing supercooling.

For considerations relative to the spherulitic graphite formation theory we refer to the paper already mentioned,* but in relation to the effect of hydrogen we suggest that one should think about graphite nuclei isolation by an adsorbed hydrogen

J. E. REHDER (Written Discussion): 2 This paper is a needed attempt to investigate more fully the fact that the presence of hydrogen in an annealing atmosphere decreases the rate of graphitization or annealing, but is disappointing in that it does not proceed far past the well-known fact that hydrogen does decrease graphitization rate.

There are one or two points mentioned in the paper on which the present writer is not clear. In Fig. 2 is shown a nodular iron which had been quenched from 20 min at 714 C (1318 F) after cooling from a higher temperature, and in which martensite

(previously austenite) is visible. Then in Fig. 4 is a nodular iron which had been held 40 min at 720 C (1328 F) and which shows no martensite nor is martensite mentioned in the text. Since the latter samples were held at 6 C (10 F) higher temperature, one would then expect to find martensite if the samples were similar. The only explanation that occurs is that the first samples were cooled rapidly to the holding temperature and thus undercooled, while the latter samples were not quenched from the holding temperature so that the presence of austenite would not have been noticed. Also the two samples may have been of different composition. Data on the cooling rate, quenching procedure, and chemical compositions of the irons involved would be appreciated.

The statement is made in the Discussion that in white iron annealed in hydrogen atmosphere "The carbon from the carbides had diffused to the surface of the metal to effect a decarburization even more severe than when hydrogen was not present." Could the authors please clarify this statement.

E. A. LORIA (Written Discussion): This is an excellent presentation of a metallographic study of the effects of hydrogen on graphitization. It is unfortunate that actual hydrogen analyses were not made in order to show the magnitude of the amounts necessary to inhibit the formation of graphite. Ordinarily, the amount of hydrogen dissolved in cast iron would be very low. In the parallel case of oxygen, there is meager proof in the literature that it is also a carbide stabilizer. When a vacuum fusion analysis is made on steel, oxygen is thought to come from oxygen dissolved in the iron, and oxygen resulting from the reduction of silicates and oxides. Cast irons, however, appear to be remarkably free of oxides and silicates, and inclusions of oxygen-bearing compounds are seldom, if ever, detected by the microscope. Thus it appears that the oxygen found in cast iron is mainly dissolved therein. If this is the case, the oxygen content of cast iron would be expected to be very low because of the appreciable amounts of carbon, silicon and manganese con-

A comprehensive study of the determination of oxygen, hydrogen and nitrogen by the vacuum-fusion method in gray irons melted in a 10-in. diam cupola under different operating condi-tions was recently completed. Several heats were made under normal cupola operating conditions and under unusual conditions intended to produce a high degree of oxidation in the cupola. The many hydrogen analyses reveal that, where the cupola operating variables produced a relatively high oxygen value, the hydrogen usually increased proportionately. Thus, for example, the values for normal cupola practice were in the 0.0016-0.0022 per cent oxygen and 0.00019-0.00022 per cent hydrogen range.** With relatively high oxqgen values of 0.0036-0.0068 per cent, the corresponding hydrogen values arose to 0.00031-0.00057 per cent. Machining tests on a constant pressure lathe on arbitration bars cast from the same metal as the gas determination pins produced poorer machinability ratings in the latter irons which would indicate carbide stabilization of the microstructure with an increase in the amounts of these dissolved gases. The nitrogen contents of all the irons were rather constant and since titanium nitride inclusions (as well as manganese sulphide) were observed in the microstructures, one would conclude that nitrogen had been made insoluble and was consequently an inactive factor.

The writer would like to know the section size of the gray irons annealed at 900 C and shown as Fig. 3, and the way these irons were melted. The major commercial implication, insofar as gray iron is concerned, is the effect of dissolved gases on microstructure during casting (or initial cooling to room temperature). Morrogh states that experiments have been made on melting iron and carbon in very high vacuum and then ironfree carbon was obtained but when traces of oxygen were present, white iron was obtained. Actually, there is no need to eliminate oxygen (and hydrogen) from the melt to make it solidify so that all one had was iron and carbon. All one has to do is to cool it sufficiently slow because even if all the gases

value, whichever is greater.

^{*} A. De Sy and J. Vidts: "Précisions et données complémentaires sur le mécanisme de formation des sphérolites de graphite," Fonderie Belge, March, 1952.

¹ Professor, Laboratorium Voor Metallurgie, Gent, Belgium.
² Foundry Engineer, Department of Mines & Technical Surveys, Ottawa, Ont., Canada.

^{*}H. W. Lownie, M. W. Mallett and E. A. Loria, "Determination of Oxygen in Iron in the Presence of Silicon and Carbon by the Vacuum Fusion method—Results on Cast Irons Melted in Small Cupola under Different Operating Conditions." To be published.

*Precision of vacuum-fusion analyses: 0.0004% Og, or ± 10% of reported value, whichever is greater. 0.00005% Hg, or ± 10% of reported

³ Senior Engr., Metallurgy, The Carborundum Co., Niagara Falls, N. Y.

from the melt were removed and then it were cooled sufficiently rapid, one would still get iron carbide (chill) structure. Then cooling rate (or section sensitivity) of the iron hinders graphite crystallization regardless of how well degasified the molten iron might be. However, there are limitations to slow cooling, particularly if the dissolved gases are high, as can be seen by the fact that the authors' irons were furnace cooled from the annealing temperature which would, of course, be a slow rate of cooling. Was hydrogen still introduced into the furnace during the cooling period?

For the record it might be well to include how the three types of iron studied by the authors were melted in order to determine what effect, if any, the melting medium would have on the initial microstructure and the consequent tendency for nucleation during subsequent graphitization by heat treatment in neutral and hydrogen atmospheres. Also what were the initial matrix structures in each case? In regard to melting, it can be mentioned that electric furnace and crucible-melted iron have a greater tendency for undercooled structures whereas cupola irons will show a coarser graphite structure owing to an inoculating effect obtained by the metal being in contact with an excess of graphitizing material (for example, good coke).

MR. BROWN (Author's Reply): Professor De Sy suggests the possibility that the mechanism of the effect of hydrogen on graphitization might be one of blanketing the graphite with a layer of adsorbed hydrogen which might inhibit transfer of carbon from matrix to graphite. Though this is a rather attractive speculation in some ways, it appears to be ruled out by the observation than a ferritic malleable iron, when its temperature is raised into the austenite range and is then cycled repeatedly over a range of 25 C for several hours, will experience a reversion of graphite to carbide if the atmosphere is hydrogen but not if it is helium. This indicates that the hydrogen does not impede the transfer of carbon from carbide to matrix as would be expected if the inhibiting effect were one of a gaseous barrier.

Mr. Rehder's perplexity over the fact that the specimen of Fig. 2 shows some austenite (now martensite) while that of Fig. 4A does not can be resolved by emphasizing the following:

that the specimen of Fig. 2 is one which was austenized (converted to graphite plus austenite of eutectoid composition), cooled rapidly to 714 C, and allowed to partially transform at that temperature before being quenched to room temperature. If it had been allowed to transform sufficiently longer it would have transformed completely to ferrite, graphite, and pearlite. In Fig. 4A, however, a specimen of the same iron was austenized as before, then quenched to a lower temperature (595 C) where it transformed completely to pearlite. There was no austenite remaining. It was then reheated to 720 C (which is too low for austenite to re-form) and the pearlite allowed to transform to graphite plus ferrite. The nodular iron used in this work contained about 3.51 C, 2.2 Si, 0.47 Mn, 1.07 Ni, and 0.12 P; the nodulizing agent was Mg.

Decarburization of white cast iron being annealed is a process of diffusion of carbon from carbides in the iron through austenite (or sometimes through ferrite) to the gas-metal interface. Graphitization involves diffusion of carbon from carbide to graphite. The diffusion in the decarburizing process is not retarded—it is actually accelerated—by the presence of hydrogen, particularly if the hydrogen is wet; but the diffusion in the graphitization process is severely retarded. This rules out an effect on the mobility of the carbon atom through the austenite lattice.

In answer to Mr. Loria, the specimens of Fig. 3 were cubes of about 1 cm; melting data are unavailable except that the specimens of Fig. 1 were melted in an induction furnace. In annealing and slowly cooling specimens such as those of Fig. 3B, hydrogen at one atmosphere pressure was continuously passed through the furnace during the entire heat treatment. Hydrogen analyses of the metal were not made in any of the experiments but the amount of hydrogen introduced was controlled by controlling the pressure of hydrogen over the specimen (usually at one atmosphere). The initial matrix structures of the iron for Fig. 1 was pearlitic and for the others ferritic plus pearlitic. In all instances comparisons were made only on specimens which were from the same heat and initially of the same structure.

# HEAT TREATMENT OF MAGNESIUM ALLOY CASTINGS

# By

# E. M. Gingerich*

#### ABSTRACT

During the past ten years, a number of papers have been presented which covered the subject of heat treatment of magnesium alloy castings. A review of this literature reveals that, while the various papers covered the subject of heat treatment, practically none of them covered all phases in very much detail. Some emphasized certain phases and passed lightly over others.

The purpose of this paper is to attempt to present a more complete picture of the important phases of commercial heat treatment of magnesium alloy castings, particularly to those in the magnesium industry not too well versed in the subject and to those in other industries interested in knowing the similarities and differences in principle and practice between the heat treatment of magnesium and the heat treatment of other metals or alloys with which they are familiar.

Magnesium alloy sand and permanent mold castings are heat treated to obtain a greater range of physical and mechanical properties or characteristics than can be obtained in the as-cast condition. The types of heat treatment commonly employed in commercial practice are: -T4-solution heat treatment; -T5-artificial aging treatment; -T6-solution followed by artificial aging treatment; and -T7-solution followed by stabilizing treatment.

The process of heat treatment will affect the mechanical properties—tensile and yield strengths, elongation or ductility, hardness and resistance to impact—in varying degrees depending upon the type of heat treatment employed. In addition to this, certain heat treatments are employed to relieve casting or quenching stresses and to obtain dimensional stability by reducing the tendency to grow during service at elevated temperatures.

The solution heat treatment designated as T4 consists of heating the castings in the range of 650-1050 F followed by a suitable quench. Solution heat treatment will improve tensile strength, elongation or ductility, and resistance to impact. Solution heat treatment will slightly reduce yield strength and hardness.

Solution followed by an artificial aging treatment, designated as -T6, consists of a -T4 solution treatment followed by heating in the range of 300-500 F for specified periods of time. This treatment will

produce an increase in yield strength and hardness. It will have little effect on tensile strength but will reduce ductility and resistance to impact. With selection of suitable aging temperature, this treatment will also effect partial stress relief and decrease the tendency toward growth in service for certain alloys.

The development of desired mechanical properties should not be the only consideration in establishing the time and temperature for the artificial aging treatment. Other aims should be:

 Relief of quenching stresses to prevent warpage of the castings during machining.

2. Removal of an appreciable portion of the potential growth in castings subject to elevated temperatures in service.

3. Selection of time and temperature suitable for more than one alloy to expedite aging and thereby reduce costs.

Artificial aging alone or solution followed by a stabilizing treatment, designated as -T5 and -T7, consists of heating castings in the as-cast and -T4 condition, respectively, in the range of 425-550 F. The -T5 treatment has very little effect on the original mechanical properties in the as-cast condition, but the -T7 treatment generally produces tensile properties in the range of the -T6 temper. The main purpose of these treatments is to obtain stress relief and freedom from growth in service at elevated temperatures.

#### Principles of Heat Treatment

The principles of the heat treatment of cast magnesium alloys are generally similar to those of other non-ferrous alloys in that they involve the solution and subsequent precipitation of one or more hardening constituents.

The structure of commercial magnesium alloys in cast form is well known from the work of various investigators. According to an excellent paper by J. B. Hess and P. F. George, the structure of commercial magnesium alloy castings consists of a solid solution matrix surrounded by an intergranular material which acts like a eutectic but crystallizes as a compound.

Depending upon the composition of the alloy, the

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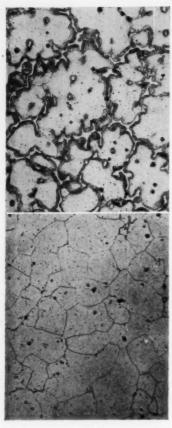


Fig. 1—AZ92 alloy, as-cast condition. Glycol etch. 100X.

Fig. 2—AZ92 alloy, —T4 condition. Glycol etch. 100X.

solid solution will contain aluminum or aluminum and zinc (as the solute elements), and the intergranular material will be  $Mg_3Al_2$  or the  $\beta$  constituent of Mg, Al and Zn. More recent work has defined the two constituents as  $Mg_{17}Al_{12}$  and  $Mg_3Al_2Zn_3$ , respectively.

During a high temperature or solution heat treatment the cored solid solution becomes more homogeneous and the grain boundary constituent is taken into solution. By prolonged heating it is possible to convert all the commercial casting alloys to homogeneous solid solutions with the exception of the manganese constituent and impurities such as Mg₂Si which are insoluble.

However, no attempt is made in commercial practice to attain complete solution because the resulting small improvement in mechanical properties does not justify the increased cost of doing so. The effect of incomplete solution of constituent is not too drastic on the mechanical properties in the —T4 condition. It is estimated that only 85 per cent of the constituent in solution will produce nearly maximum properties.

As in the case of other non-ferrous alloys, factors such as grain size, section thicknesses and character of charge will affect solution treatment procedures, particularly with respect to time required to attain satisfactory solution. Coarse grained material will respond less rapidly to the solution of constituent. The constituent occurs in larger particles which require more time for solution. The distance over which diffusion

must take place to attain complete homogeneity of the solid solution phase is greater in larger grains than in small grains.

The high temperature solution treatment is followed immediately by a quench to maintain the solid solution structure. Unlike aluminum alloys, the rate of precipitation of the solute elements is relatively slow and, for this reason, the quench need not be drastic. It is, therefore, sufficient in commercial practice to quench in still air or in a stream or streams of moving air, as generated by industrial fans.

Fan quenching is necessary to retain -T4 properties, but not if castings are to be subsequently converted to the -T6 or -T7 condition. Quenching in water is said to improve tensile and yield strength of casting alloys on subsequent aging. This practice, however, carries with it the possibility of cracking and distortion and the introduction of internal stresses.

The completeness of solution heat treatment is determined by the amount of constituent remaining out of solution. This may be done by comparing a suitably polished and etched specimen with a standard rating chart.² This chart rates the amount of constituent out of solution from 0 with none to 10 representing the as-cast structure. The intermediate structures have more constituent out of solution as the numbers increase.

Figure 1 shows a typical as-cast structure of AZ92 alloy. Massive compound surrounded by lamellar constituent occurring by precipitation from solid solution during cooling is observed at the grain boundaries. In Fig. 2, illustrating a —T4 structure, all the lamellar constituent has been dissolved and most of the massive compound has been taken into solution.

# Structure Changes

Aging in the range of 300-450 F after a regular solution or -T4 treatment effects a partical decomposition of the solid solution, precipitating, in very finely divided particles, that amount of magnesium-aluminum or magnesium-aluminum-zinc constituent not soluble at the aging temperature. It is this change in structure which increases yield strength and hardness.

Discussing briefly the physical changes which take place during the artificial aging treatment, the precipitation which takes place is classed as grain boundary, lamellar or pearlitic, and general or Widmanstätten type.³ Of minor importance, grain boundary precipitation occurs quite early in the aging cycle and has the effect of more clearly outlining or revealing grain boundaries.

The lamellar or pearlitic type also starts early in the aging cycle. This type of precipitation apparently originates in nuclei in the grain boundaries and grows inward in patch-like formations resembling the pearlitic structure in steel. More than one patch or area may be found in a single grain. The area covered by this type of precipitate will vary and appears to depend upon the rate of cooling induced by the quench after solution heat treatment.

The general type of precipitation occurs later in the aging cycle than the other two types. It originates 18

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Fig. 3 (left)—AZ92 alloy, —T6 condition. Glycol etch. 100X.

Fig. 4 (below)—AZ 92 alloy, —T6 condition, showing general precipitation. Glycol etch. 500X.

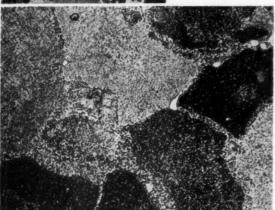




Fig. 5 (left)—AZ63 alloy, —T6 condition. Glycol etch. 100X.

Fig. 6 (below)—AZ 63 alloy, —T6 condition, showing lamellar and general precipitation. Glycol etch. 500X.



from an exceedingly fine dispersion of constituent throughout the grain and becomes visible under the microscope when the particles grow to sufficient size. This type of precipitation occurs throughout the grain except possibly where lamellar precipitation occurs. The particles are arranged in parallel planes with a single orientation and often have the appearance of the well known Widmanstätten structure. Long etching periods are sometimes necessary to reveal the general type of precipitation.

In the opinion of some investigators, maximum yield strength and hardness, particularly in the AZ92 alloy, are obtained when the minimum of lamellar

precipitate is produced.

Figures 3, 4, 5, and 6 show precipitated constituent after a -T6 heat treatment. Figures 3 and 4 illustrate the microstructure of AZ92 alloy at 100X and 500X, respectively. The precipitated constituent occurs generally throughout the grains.

Figures 5 and 6 represent AZ63 alloy at similar magnifications. These show, more clearly, individual patches of lamellar precipitate and a considerable amount of general or Widmanstätten precipitate.

A stabilizing treatment is essentially an artificial aging treatment performed at higher temperatures. The effect on the microstructure is similar in that the same types of precipitation occur as in the lower temperature ranges. The difference lies in the larger amounts and coarser particles of precipitate formed at the higher temperatures.

It has been stated that one of the functions of the stabilizing treatment is to reduce the tendency of castings to grow during service at elevated temperature. It is also sometimes called a "degrowthing" or growth

removal treatment.

Growth in castings is essentially a change in volume resulting from the precipitation and coalescence of constituent precipitated from solid solution. Thus the purpose of a stabilizing treatment is to complete precipitation before machining so that only a negligible amount of growth will take place during any subsequent heating in service. The use of higher temperature in stabilizing or degrowthing treatments is primarily to expedite the process.

A great deal of work has been done by various investigators to determine the amount of growth that may occur in magnesium alloys. As would be expected, the amount of potential growth is greatest in the -T4 or solution treated condition. Increasing the quenching rate increases the amount of possible growth. Artificial aging and stabilizing treatments remove progressively larger amounts of growth.

#### Heat Treat Furnaces

Furnaces for heat treating magnesium alloy castings must be so designed and constructed as to provide uniform temperatures throughout the furnace as attained by a high rate of air circulation. They must be equipped with thermocouples and controlling equipment of such accuracy that no areas deviate from the nominal operating temperature by more than 10 F.

The furnaces must be tight enough to maintain a protective atmosphere within and fitted with an easily observed and carefully controlled method of introducing that atmosphere. They must be provided with sufficient thermocouples to continuously and completely survey furnace temperatures. should be equipped with an alarm system to call attention to any increases in furnace temperatures which may lead to fires. The source of heat must be well shielded to avoid local overheating of parts of the charge by radiation.

Furnaces may be continuous or batch type. A continuous furnace may be desirable when one alloy and one heat treatment are used. A batch-type furnace is preferred for a jobbing foundry because it offers more flexibility in economically handling a number of alloys and heat treatments. Furnaces may be heated

indirect gas-fired furnaces are preferred.

# **Heat Treatment Hazards**

electrically or indirectly by gas or oil. Electric and

The quality of heat-treated castings, with particular reference to mechanical properties, is affected by several metallurgical phenomena, more or less peculiar to magnesium alloys, which may occur at the heattreating temperature. One of these is surface oxidation. This is the result of the reaction of magnesium with the normal atmosphere, air, at heat-treating temperatures and its occurrence is generally attributed to some deficiency in the protective atmosphere of the furnace.

Surface oxidation is generally evident as a gray, black powder on the surface of castings. Also, craters and voids may be present on the surface, and the latter may extend into the interior of the casting. The effect of such oxidation may lead to local weakening of the metal structure and cause serious reduction in mechanical properties and serviceability of castings. In extreme cases it can lead to fires in heat-treating

Figure 7 shows a severe example of surface oxidation in AZ92 alloy. Pits may be observed on the surface of the specimen, and grain boundary voids under the surface.

The most effective means of eliminating oxidation has been by the use of a protective atmosphere in the heat-treating furnace. The most commonly used protective atmosphere is produced by passing a continuous stream of sulphur dioxide gas into the furnace and maintaining a concentration of 0.5 to 1 per cent

in the furnace atmosphere.

Sulphur dioxide atmospheres may also be obtained from sulphur or iron pyrites. These materials are somewhat cheaper than compressed sulphur dioxide gas but are more difficult to handle and control. Sulphur can be used to generate sulphur vapor in a separate burner discharging into the furnace, or placed in a steel container inside the furnace where a slow evolution of sulphur dioxide gas will take place as the sulphur vapor burns. Pyrites is used in the latter manner.

Magnesium castings may also be successfully heat treated in an atmosphere containing carbon dioxide. This is advantageous where it is necessary to heat treat magnesium and aluminum castings in the same furnace. The use of carbon dioxide gas is more ex-

pensive than sulphur dioxide because a much higher concentration must be maintained to produce a protective atmosphere. Its use appears to be advantageous only under conditions mentioned above except that it does eliminate the fume problem.

Some consideration has been given and work done on other gases such as helium, argon, hydrogen, hydrocarbons and carbon monoxide as protective atmospheres.4 None of these gases has been considered to be satisfactory. Helium and argon are too costly and do not act as inhibitors of oxidation but function merely by displacing air in the furnace. Hydrogen, methane or other hydrocarbon gases and carbon monoxide may be sufficiently protective but they are difficult to handle and are dangerous.

The solution heat treatment of magnesium castings in direct-fired oil or gas furnaces appears to be of relatively little commercial importance, probably because it does not provide as much protection against oxidation as do inderct-fired fuel or electric furnaces with sulphur dioxide atmospheres. With this type of furnace, heat treating is done directly in the products of combustion. When this method is used the hot combustion gases must be diluted to approx. 30 per cent with air to provide a suitable temperature.

One undesirable component of combustion gases is water vapor which tends to promote oxidation.5 Dehydration of the gases would, of course, remove the water vapor. Since water vapor is present, the addition of sulphur dioxide for protection against oxidation is not desirable because of the formation of corrosive sulphurous acid fumes.

# Other Methods Less Commonly Used

Heat treatment in a vacuum furnace is ideal from the standpoint of protection since all harmful gases are eliminated. Its use is limited, however, because of non-uniformity of temperature distribution. The modern circulating air furnace is much preferred because of simplicity of construction, uniformity and accuracy of control, and ease of operation.

Heat treatment of castings in salt baths has not found much favor, although good temperature control can be maintained and atmospheric considerations are eliminated. Salt mixtures can be explosive in contact with magnesium, particularly if they contain nitrates, nitrites, or cyanides. Mixtures of molten chromates were used in Germany for several years. The principal objections to salt baths are loss of salts because of carry-over to the quenching tanks and the difficulty in later washing the solidified fused salts from the casting.

Other phenomena which may appear during heat treatment are eutectic or grain boundary melting and grain growth or germination. These are controlled or eliminated by employing specific heat-treating

practices.

Eutectic melting is caused by heat treating at too high a temperature or by heating too quickly to the heat-treating temperature. The Mg-Al-Zn alloys contain two eutectics, one of which melts at about 660 F and the other at about 780 F. When the zinc content is above 2 per cent, it is often necessary to hold the castings at 650 F until the lower melting eutectic

dissolves in the solid solution matrix. The temperature can then be raised to 770 F to cause solution of the higher melting eutectic which is hardly affected

by the 650 F treatment.

Eutectic melting in itself, unless unusually severe, does not adversely affect mechanical properties. However, eutectic melting is sometimes accompanied by intergranular oxidation causing voids which lower mechanical properties. The common practice in avoiding eutectic melting is to slowly raise the temperature of the castings being heat treated through the lower part of the temperature range to permit diffusion of the low-melting-point eutectic into the solid solution matrix, and then raise the temperature to not higher than 10 F under the melting point of the least fusible eutectic present in the alloy.

It has also been found (U. S. Patent No's. 2,185,452 and 2,185,453) 6.7 that the addition of about 0.05 per cent calcium eliminates the need for the gradual increase in heat-treating temperature so that the final heat-treating temperatures of Mg-6Al-3Zn and Mg-9Al-2Zn alloys (ASTM designations AZ63 and AZ92, respectively) can be raised 10-20 F, and the holding time for the former can be greatly shortened. The increase in temperature permissible when calcium is present makes it possible to save several hours in the

heat-treat cycle.

Figure 8 shows eutectic melting in AZ92 alloy. Areas of fused constituent may be observed in the center of the micrograph. The extreme angularity of these areas is an indication of considerable local fusion.

#### Occurrence of Coarse Grains

Germination is the tendency of some magnesium alloys to develop large grains when held at temperatures at or near the heat treating temperature. This is generally detected on machined surfaces of castings.

The enlarged grains usually occur in irregular shaped areas surrounded by normally fine-grained areas. Germination seriously affects the mechanical properties of castings. Tests on bars machined from fine-grained and coarse-grained areas in the same castings show at least a 50 per cent reduction in tensile and yield strengths in the coarse-grained areas.

Germination may occur in sections of castings in which the following conditions exist during heat

treatment:

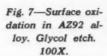
1. Grain size is very fine.

2. Internal stresses and stress gradients are present although the magnitude of these stresses appears to be quite small. Chilled areas are particularly prone to germination. Even a saw-cut will sometimes start germination.

3. Time at heat-treating temperature is sufficient to effect complete solution of grain boundary constituent. Germination apparently does not proceed until the grain boundaries are essentially free of interfering material.

4. The heat treating temperature must be above 730 F.

One positive method of preventing germination is the interrupted solution heat treatment described in U. S. Patent No. 2,389,583.8 In this method, the total



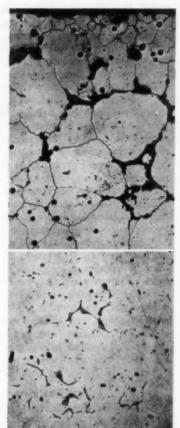


Fig. 8 — Eutectic melting in AZ92 alloy. Glycol etch. 100X.

time at heat treating temperature is broken up into several periods with intervening cooling periods.

For example, a charge of castings given an interrupted heat treatment is first heated for 5 hr at 775 F, then cooled to under 500 F and raised to 775 F for 5 hr. This is followed by another cooling cycle and a final heating period of 5 hours at 775 F for a total of 15 hr at temperature. The charge is then regularly quenched.

The interrupted cycle heat treatment is based on the conception that germination cannot proceed without a gradient in energy level within the casting. The volume change resulting from alternate solution and precipitation relaxes the lattice and eliminates

gradients in the stress or energy level.

The first cycle of short-time solution and precipitation is most effective in preventing germination. Succeeding solution periods may be carried out for longer periods of time before growth takes place until, finally, grain growth tendency no longer exists at the end of the last solution period. This, as stated before, is believed due to a relaxation of the internal stresses which promote germination brought about by the volume changes accompanying the cycles of solution and precipitation.

It may be possible to control germination to some extent by modifying the factors which appear to promote germination. That is, production practices could be adopted which would produce coarser grained castings and reduce the amount and degree of chilling, and thus minimize stresses in the castings. Such expedients are not desirable because they are unreliable in effect and not in the direction of

producing high-strength castings.

The alloys particularly susceptible to germination are those relatively high in aluminum and low in zinc. These are the A10, AZ92, and AZ91C alloys. It must be kept in mind that germination in these alloys is not general but originates in critical conditions of grain size and casting stresses. Heat treatments to prevent germination are applied only when necessary, and may have to be varied to conform to the characteristics of individual castings.

Figure 9 shows germination in a AZ92 alloy casting. The irregular area of large grains is surrounded

by normally fine-grained areas.

A heat treating hazard of a non-metallurgical nature is warpage. Magnesium alloy castings lose much of their rigidity at solution heat treating temperatures and may warp, sag, or otherwise deviate from the original casting shape or dimensions. The tendency to warp or distort is influenced to a large extent by size, shape, and section thickness. This may be eliminated, in many instances, by the proper supporting and placement of castings in the heat-treating rack, but sometimes special jigs or supports are required to maintain the proper casting shape.

#### **Heat Treatment Practices**

The discussion of heat treatments up to this point has dealt only in principles and little reference has been made, specifically, to times and temperatures. As with other groups of non-ferrous alloys, chemical composition of the magnesium alloys governs heat-treating temperatures. Size of castings, size of the charge, section thickness and grain size of castings largely determine heat-treating times.

In common with other non-ferrous alloys, the heattreating temperatures selected for magnesium alloys are those just below the temperatures at which eutectic melting is encountered under commercial heattreating conditions. These have been established generally by experiment until the proper temperature

has been determined.

The heat treating practices for the alloys described in this paper are discussed under the headings of the principal types of heat treatments, namely, solution, solution followed by artificial aging, and stabilizing. A tabulation of typical practice for these heat treatments is given in Table 1.

#### Solution Heat Treatments

The current standard heat-treatable castings are as follows:

Composition		ASTM	Dow	Alcoa	
Al, %	Zn, %	Designation	Designation Designation		
10.0		A 10	G	AM240	
9.0	2.0	AZ92	C	AM260	
8.7	0.7	AZ91C	AZ91C	<b>AMA263</b>	
6.0	3.0	AZ63	H	AM265	

Typical schedules for the solution heat treatment of each of these alloys are:

A10 Alloy: Sand castings are solution heat treated for 18 to 24 hr at 780 F, followed by the quench in moving air. Calcium, approximately 0.05 per cent, is added to the molten metal by some foundries to aid in heat treatment. With this calcium present the soaking temperature can be raised 15–20 F. The same increase in soaking temperature applies in the case of the other alloys when calcium is added. The net effect of 0.05 per cent calcium is to raise the heat-treating temperature and reduce the time required at temperature.

It is preferred practice to start a heat treatment charge of castings with the temperature of the furnace no more than 500 F. It is also desirable to increase the temperature of the charge at a gradual and uniform rate, through about a 2-hr period, to the soaking temperature. The soaking periods for the various alloys indicated herein are preceded by such a pre-heating period.

Since permanent mold castings in the A10 alloy are quite susceptible to germination, an interrupted heat treatment is generally used. Typical schedules consist of soaking 2–3 hr at 785 F, cooling to 300–400 F and reheating to 785 F, and holding at this temperature for 7 to 12 hr.

AZ92 Alloy: A typical solution heat treating sched-



Fig. 9—Germination in AZ92 alloy casting.

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TABLE 1-TYPICAL HEAT TREATING SCHEDULES FOR MAGNESIUM CASTING ALLOYS

Alloy		on Heat ent (T4) ¹	Solutio	g After on Heat ent (T6)	Solution	izing After on Heat ent (T7)	Stab Treatme	ilizing ent (T2)
	Time, hr	Temp., F	Time, hr	Temp., F	Time, hr	Temp., F	Time, hr	Temp., F
A-10	18-24	780-800	10-12	325-400	4	500	_	_
AZ92	18-222	760-775	5-15	400-425	4	500	_	_
AZ91C	18-20°	775-790	3-5	400-425	4	500		-
AZ63	10-153	720-740	5-15	400-425	4	500	_	
ZK61	2	930	48-72	300	*	****	_	_
ZK51	-	_					12	350
EZ33	-	_	_	Millions.	_	0.000	12	350
EK30A	16	1050	16	400	-	_	-	-

Castings are brought up to holding temperatures at uniform rate of rise in approximately 2 hr before treatment at times and temperatures shown.

2. The step or interrupted cycle heat treatment described in the text may be necessary to prevent germination.

This treatment is usually preceded by heating to 640 F in
 hr, holding for 2 hr, heating to heat-treating temperature in
 hr to prevent eutectic melting.

ule for this alloy, without calcium, is 18–22 hr at 760 F, followed by a standard quench when germination is not a problem. There are some variations in procedure when germination occurs, determined apparently by specific conditions.

In one procedure there are soaking periods of 5, 6½ and 10½ hr at 760 F. In the two intervening cooling periods the castings are cooled rapidly to 650 F, then to 450 F in 2 hr, then raised to 780 F in 2 hr. Another variation consists of cooling the castings to 200 F outside the furnace, which is maintained at the soaking temperature. The variations in practices where germination must be prevented are usually developed to do a tedious but necessary job in the most convenient and economical way.

Soaking temperatures are, of course, 15-20 F higher when calcium is added to the metal. Soaking times are also generally shorter.

AZ91C Alloy: The solution heat treatment procedures for this alloy are quite similar to those of the AZ92 alloy except that the soaking temperatures can be slightly higher because of lower zinc content.

In one procedure there are three soaking periods of 5, 6 and 9 hr at 775 F. The intervening cooling periods consist of cooling rapidly to 650 F, then slowly in 2 hr to 450 F, then reheating in 2 hr to the soaking temperature of 775 F. When the alloy contains calcium, the soaking temperature is increased to 790 F and the periods shortened somewhat.

AZ63 Alloy: This alloy is not susceptible to germination and it is not necessary to employ an interrupted treatment. However, the alloy is prone to eutectic melting, and a different type of solution heat treatment must be employed to counteract this characteristic. The conventional method is to employ a slow heating rate through the lower part of the temperature range to permit diffusion of the low-melting-point eutectic—the Mg-Al-Zn constituent.

# Solution Followed by Artificial Aging Heat Treatments

In one procedure, in which slow heating is emphasized, the charge is heated to 640 F in 2 hr, soaked at this temperature for 2 hr, then heated to 720 F in 2 hr and soaked 15 hr before quenching.

Artificial aging treatments following the solution treatments are less complicated and generally more uniform throughout the industry. Typical artificial aging treatments with respect to time and temperature are included in the table below:

Alloy	Time, hr	Temperature, F
A10	10–12	325-400
AZ92	5–15	400-425
AZ91C	3-5	400-425
AZ63	5–15	400-425

Normal atmospheres are used in aging furnaces. Temperatures are too low to require the protection of sulphur dioxide.

It is possible to employ numerous combinations of time and temperature in the aging treatments. Less time is required at the higher temperatures to complete aging. Somewhat higher elongations are obtained but with some reduction in yield strength. Longer times and higher temperatures are used when partial stress relief and growth removal are desired. The degree of solution of constituent is reported to influence the effect of aging. Maximum properties are developed in the —T6 condition when practically all constituent is dissolved in the solution treatment.

# Stabilizing Treatment

The schedules for stabilizing heat treatments to provide practically complete stress relief and freedom from growth show little variation in commercial practices. This is generally 4 hr at 500 F for castings either in the as-cast or -T4 condition.

#### Magnesium-Rare Earth-Zirconium Alloys

This comparitively new series of magnesium-base alloys has come into use over the past few years. These alloys are used in castings for service at temperatures up to 600 F. The optimum composition has not yet been agreed upon by either foundries or aircraft manufacturers. For the purposes of this paper only two alloys will be considered, each representative of several in each of two general groups. All alloys of this series have essentially the same high-temperature properties.

One group of alloys contains zirconium in about the maximum amount that can be alloyed, and is used after a low-temperature age of about 12 hr at 350 F. EZ33 alloy, nominal composition 3 per cent rare earths, 2.5 per cent zinc, and 0.7 per cent zir-

conium, is typical of this group.

The second group of alloys contains zirconium in lesser amount than EZ33, and is used after a solution heat treatment of 16 hr at about 1050 F followed by a 16-hr age at about 400 F. EK30A alloy, nominal composition 3 per cent rare earths and 0.3 per cent zirconium, is typical of this group. The furnace atmosphere is maintained at about 2 per cent SO2 during the solution heat treatment of this alloy.

# Magnesium-Zinc-Zirconium Alloys

This series of alloys has been used experimentally to give a combination of high yield strength and high elongation after a low-temperature-aging treatment. Alloy ZK51, with which most work has been done both in this country and abroad, nominal composition 4.5 per cent zinc, 0.7 per cent zirconium, is used after a 12-hr age at 350 F.

# Fires During Heat Treatment

A hazard encountered in the heat treatment of magnesium alloy castings is the danger of fires. In general, good operating practices employed for satisfactory heat treatment will also serve as fire-prevention practices. Two sets of furnace temperature controls, operating independently of one another, should be provided. Controls and other furnace equipment should be checked frequently to maintain at all times the recommended temperaure ranges.

Sulphur dioxide atmospheres should be used. Furnaces should be kept clean and free from iron scale. Each furnace charge should consist of castings of one alloy composition only. These should be free from fines or dust. Water vapor is extremely hazardous and, for this reason, water should never be used as a bearing coolant in recirculating fans installed inside

heat-treating furnaces.

In spite of precautions taken in heat treating, fires can and do occur. Well-known methods have been developed to combat such fires. These include the use of materials such as G-1 powder and soapstone. The use of these materials requires the rack of burning materials to be outside the furnace or just inside the door so that the materials can be effectively applied. A fire can be fought with the furnace door open but without moving any racks by the use of a special G-1 pump with which a stream of powder may be sprayed to any part of the furnace.

A method of fighting a fire entirely within the furnace and without opening the door consists of burning lubricating oil within the furnace to reduce the rate of burning. This method is intended to minimize damage to the furnace rather than save castings.

A new method has been developed which is effective in extinguishing fires.10 This method makes use of compressed boron trichloride gas. The gas is introduced into a small opening in the furnace through a pipe and tubing leading from the containing cylinder. The action of the gas is to form a protective film over the burning metal which quickly checks a vigorous fire. The gas flow is continued until the furnace temperature is about 700 F and the fire is extinguished.

An alternate method is to use boron trichloride until the vigorous fire is checked. The rack can then be withdrawn from the furnace and the remaining fire extinguished quickly and without discomfort by G-1 powder.

#### References

1. J. B. Hess and P. F. George, "The Metallography of Commercial Magnesium Alloys," American Society for Metals Convention (1942).

2. P. F. George, "A Numerical Rating Method for the Routine Metallographic Examination of Commercial Magnesium

Alloys," ASTM Bulletin (1944).
3. A. E. Flanigan, I. I. Cornet, R. Hultgren, J. T. Lapsley, and J. E. Dorn, "Factors Involved in Heat-treating a Magnesium Alloy," AIME Technical Publication No. 2282 (1947).

4. C. E. Nelson, "Atmospheric Control in the Heat Treatment of Magnesium Products," American Society for Metals Convention (1941).

5. R. F. Thomson, D. B. Burks, and W. E. Jominy, "Effects of Water Vapor on Magnesium Alloys During Heat Treatment," Metal Progress, July, 1946, pp. 67-71.

6. R. T. Wood, U. S. Patent 2,185,452 (Jan. 2, 1940). 7. R. T. Wood, U. S. Patent 2,185,453 (Jan. 2, 1940).

8. R. T. Wood and M. E. Gantz, U. S. Patent 2,389,583 (Nov. 20, 1945).

9. D. Burks, R. F. Thomson, and W. E. Jominy, "Heat Treatment of Magnesium," A.F.A. Transactions, vol. 53, pp. 253-257 (1945).

10. "Extinguishing Magnesium Fires in Heat Treating Furnaces with Boron Trichloride," Dow Technical Memorandum No. 17a (1950).

#### DISCUSSION

Chairman: A. CRISTELLO, American Light Alloys, Inc., Little

Co-Chairman: H. E. Elliott, Dow Chemical Co., Bay City, Mich.

Recorder: A. CRISTELLO.

P. A. FISHER (Written Discussion): 1 I have been interested to note the details given in this paper of the heat treatment cycles for the T4 condition and partictularly the emphasis on the necessity to avoid germination. In Great Britain the two alloys of this type now universally employed contain 8 per cent and 91/2 per cent aluminum respectively, both containing 0.5 per cent zinc. The two alloys are known as A8 and AZ91. Alloy AZ63, although used prior to 1939, is not now used mainly owing to its increased tendency to microporosity as compared with A8 and AZ91.

With these two alloys in Great Britain, germination in both sand and permanent mold castings is almost unknown, and the treatment to obtain the T4 condition consists simply of heating the castings for 8 hr at 725 F followed by 16 hr at 770-790 F. In this connection, I think the figure given in this paper for the melting point of the magnesium-aluminum eutectic is a little low. The melting point of this eutectic is, I believe, 815 F.

It is not clear from this paper whether germination has, in the U.S.A., been found of practical difficulty in the case of AZ91C, although the rather complicated heat treatment cycle mentioned in connection with this alloy tends to suggest that some difficulty is experienced. If this is so it is rather difficult to explain in view of our experience with virtually the same alloy-it is noted however that the nominal zinc content of AZ91C is slightly higher than that of the British version.

The effect of calcium has also been investigated by my Company and we have confirmed that it enables slightly higher temperatures to be used for solution treatment. If too much calcium is added however the rate of diffusion decreases so that no benefit is gained. A serious disadvantage is the fact that distortion of castings increases considerably if the higher solution treatment temperatures are employed and this, coupled with the fact that calcium additions are somewhat difficult to control is a production foundry, has not encouraged its use in the U.K.

Regarding the generation of SO2 in electric resistance heat treatment furnaces, it has been our experience that the use of

¹ Metallurgist, Magnesium Elektron Ltd., Manchester, England,

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pure sulphur frequently leads to breakdown of the heating elements unless these are totally enclosed. In U.K. foundries, it is usual to generate the sulphur from pyrites and undue difficulty with heating elements does not seem to arise in this case.

With reference to the section of this paper dealing with zirconium alloys, the mention is of course quite brief but, I think, gives the wrong impression of the present development of these alloys. Both the magnesium-rare earth-zirconium and the magnesium-zinc-zirconium alloys have now been in commercial use in Britain for five years, and many thousands of castings, amounting to hundreds of tons in weight in each alloy, have

been produced and used.

It is also incorrect to state that the optimum composition of the magnesium-rare earth-zirconium alloys has not yet been agreed. While the compositions selected from any alloy system are subject to modification as a result of research and experience, two definite and by now well known alloys in this particular system were introduced to the British aircraft industry some five years ago and have been accepted as a very useful addition to the range of ultra light alloys ever since. They are in fact now regarded as a standard and routine production line, by the foundries engaged in their manufacture. These alloys have also been available to the U.S. industry for nearly three years.

In this class of alloys the use of the solution treatment at 1050 F is to improve the low room temperature tensile properties obtained when the zirconium content is limited to about 0.25 per cent. However, considerable care must be taken in applying this treatment since, if the temperature is too low, the creep strength

of the alloys is seriously affected.

The use of such a high temperature obviously involves considerable risk of distortion of the castings and may be very expensive. In the alloys developed by my Company, solution treatment is avoided. The higher level of room temperature tensile properties can readily be achieved by the use of a higher zirconium content and the heat treatment then merely consists of a low temperature stabilizing treatment in which the temperature involved is not very critical and no danger of distortion is present.

With regard to the magnesium-zirconium-zinc alloys, here again the use of the alloy which you term ZK51 is, in the U.K., far beyond the experimental stage. Many thousands of castings, again amounting to many hundreds of tons in weight, have been made in this alloy over the past five years and, as in the case of the alloys containing rare earth metals, the use of this alloy is a normal operation in the foundries engaged in its manufacture. As a point of interest, the British "Comet" aircraft, which it is hoped will commence the first pure jet commercial passenger service, contains over 1,000 magnesium alloy castings, the majority of which are in Z5Z, the alloy which is of course known in U.S.A. as ZK51.

It should also be mentioned that J. W. Meier of the Canadian Bureau of Mines has proposed the use of a magnesium-zinc-zirconium alloy containing 6 per cent zinc which is used in the solution treated condition followed by artificial aging. As in the case of the alloy containing rare earth metals, the temperature required for solution treatment is very high (930 F) and the dangers of distortion of castings are correspondingly also high.

There are, in addition, various other difficulties involved in the commercial production of this alloy which, in our opinion, render it less attractive to the founder than Z5Z (or, in American parlance, ZK51), in spite of the outstanding properties which can be obtained with some types of test bar.

To conclude therefore, I would like to say that in both classes of zirconium alloys there are available well known and tried alloys. In both classes the alloys have been particularly designed to present the least difficulty to the founder so that the foundry industry can offer to its customers a sound and practical proposition. The zirconium alloys developed by my Company are now being cast in eight different countries, and their use, as I have mentioned earlier, is far beyond the experimental stage and

daily growing in importance and quantity.

J. W. MEIER (Written Discussion): A recapitulation of proper heat treating techniques for magnesium alloy castings is very valuable at this time of considerably increased interest in various applications of magnesium products, and the author should be congratulated for this fine compilation. My only comment would be to ask for a correction of the heat treating cycle for alloy ZK61 given in Table 1. Similarly, as for other alloys, there is a range of temperatures and times to be used depending on casting size and shape, as well as local heating facilities. The range for alloy ZK61 castings was established* as follows:

Solution heat treatment 1-5 hr at 800-930F Aging 24-48 hr at 265-300 F.

There is also a question of definitions: Would it be not more correct to call the heat treatment used for alloys ZK51 and EZ33 an aging treatment instead oi "stabilizing" as used in Table 1, which is not consistent with the authors' definition of the stabilizing treatment.

C. E. Nelson: I would like to point out that germination does not occur at all on the AZ63 alloy, but has been observed on the AZ92 alloys. Fortunately, this has never been a serious problem in actual practice because the occurrences of germination are rare and, in any case, the so-called "interrupted" type heat treatments for AZ92 alloy, some of which are described in this paper, will entirely prevent germination.

One published work (Peters, Busk and Elliot) shows that the occurrence of germination is related, among other things, to alloy composition and the data show that the AZ91 type composition is near the borderline and is not as likely to

germinate as AZ92 (higher Al and Zn) alloy.

M. E. Brooks: * Use of CO₂ instead of SO₂ in heat treat furnaces is of great interest to many foundries running both aluminum and magnesium alloys. Can you give details about the latest tests on the use of CO₂ including the concentration used in the furnace atmosphere?

MR. GINGERICH: There had been some recent work done with CO₂ in a production size furnace but we do not have in-

formation on the results of the test.

² Canadian Department of Mines & Technical Surveys, Qttawa, Ont., Canada.
⁴ J. W. Meier and M. W. Martinson, "Development of High-Strength Magnesium Casting Alloy ZK61," A.F.S. Transactions, vol. 58, pp. 742-751 (1950).

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# EFFECT OF PERCENTAGE OF NODULAR GRAPHITE ON CERTAIN MECHANICAL PROPERTIES OF MAGNESIUM-TREATED CAST IRON

By

R. W. Lindsay* and A. Shames**

#### ABSTRACT

A series of melts was prepared by induction melting, and was treated in such a way that the graphite structures in the resulting castings varied from all flakes at the one extreme to practically all nodules at the other extreme. The variation in graphite structure was produced by adding increasing quantities of magnesium to a base iron, which could be nodularized by sufficient magnesium addition. A double-treatment technique was used; the magnesium was added as an iron-silicon-copper magnesium alloy followed by a ferrosilicon addition. Microexamination of the castings showed the percentages of nodules in the various structures to be 0, 19, 55, 81 and 92 per cent.

Two standard 0.505-in, diameter tensile bars and eight R. R. Moore fatigue specimens with 0.300-in, minimum diameter were prepared from each casting, and hence from each structure of the various nodule contents. These test specimens were used to determine the ultimate strength and elongation in tension, the endurance limit for un-notched specimens, and the endurance ratio.

The ultimate strength, elongation, and endurance limit increase with increasing percentage of nodules in the structure. The greatest improvement in these properties results as up to 50 to 60 per cent of the graphite in the structure assumes a spherulitic shape. It is to be noted that the graphite which does not exist in spherulitic form in these magnesium-treated irons will be in the form of very stubby flakes. The endurance ratio decreases with increasing percentage of nodules in the structure.

These results are of importance industrially, since they indicate that high values of certain mechanical properties comparable with those for a fully nodular iron may be obtained with a substantial amount of stubby flake graphite and irregularly-shaped nodules present in the structure.

It is well known that if sufficient magnesium is retained in an iron of proper gray iron composition, a nodular iron will result in the as-cast condition. The usual limits set for residual magnesium are 0.04-0.10 per cent. Morrogh† described the manner in which the mechanical properties of the treated iron may vary with the quantity of nodular graphite in the structure. Furthermore, it would be of interest to determine whether or not a critical percentage of nodules might result in development of properties approaching those of the fully nodular iron. These points formed the subjects of the investigation to be

described in this paper. Tensile and fatigue properties of the treated iron in the as-cast condition were used as criteria for the change in properties with quantity of nodule formation.

# **Experimental Procedure**

All heats were made in a high-frequency induction furnace of 17-lb capacity. Charges were made up of a low-sulphur, low-phosphorus pig iron, and a low-carbon steel. The molds were constructed of a fine bank sand mixed into 2 per cent of oil binder, 1 per cent of cereal, and 4 per cent of water. These molds were baked for 1 hr at 400 F. The mold cavity was  $2\frac{1}{2} \times 2\frac{1}{2}$  in. at the bottom,  $3 \times 3$  in. at the top, and  $8\frac{1}{2}$  in. long.

A superheating temperature of 2800 F was used in all heats, and the magnesium addition was made at about 2650 F followed closely by a ferrosilicon addition. The magnesium alloy used to make the additions contained 10.20 per cent magnesium, 6.87 per cent copper, 34.32 per cent silicon, and the remainder iron. The metal was poured directly into the mold shortly after the ferrosilicon addition at a temperature of about 2600 F. The molds were top poured, and the only feeding action was the result of the increase in section size from bottom to top of mold. Each casting included a definite shrink, but it was easily possible to avoid this in sectioning the casting into test bar blanks.

The percentage of nodules was varied by introducing different amounts of the magnesium alloy into the molten metal. Pertinent data are given in Table I as regards the compositions of these irons.

Specimens for metallographic examination and estimation of percentage of nodules were taken from several locations in each casting. Blanks were cut from each casting for the preparation of two standard 0.505-in. diameter tensile bars and eight R. R. Moore fatigue specimens with 0.300-in, minimum diameter. After final machining, the fatigue specimens were ground and then polished with 00 emery paper.

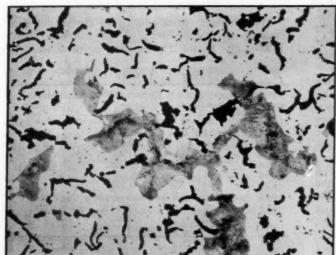
Associate Professor of Metallurgy, The Pennsylvania State College, State College, Pa.

^{••} Research Engineer, North American Aviation Inc., Columbus, Ohio; formerly Metallurgist, Battelle Memorial Institute, Columbus, Ohio.

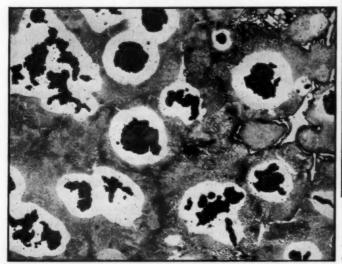
[†] H. Morrogh, "The Influence of Some Residual Elements and Their Neutralization in Magnesium-Treated Nodular Cast Iron," A.F.S. Transactions, vol. 60, pp. 439-452 (1952).



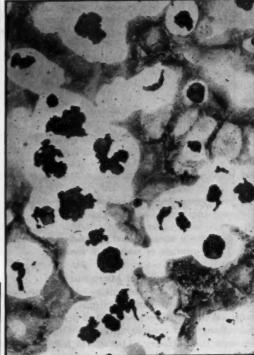
(a) Melt No. 14. Nodules, 0%.



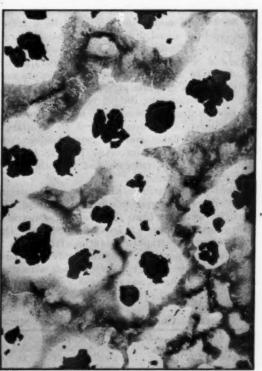
(b) Melt No. 13. Nodules, 19%.



(c) Melt No. 11. Nodules, 55%.



(d) Melt No. 7. Nodules, 81%.



(e) Melt No. 6. Nodules, 92%.

Fig. 1—Photomicrographs (a, b, c, d, e) show microstructures of the various castings used in the investigation. Picral etch. X100.

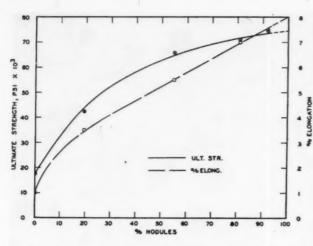


Fig. 2—Tensile strength and elongation values vs. percentage of nodules in structure.

# Discussion of Microstructures

Estimation of the percentage of nodules was accomplished by counting the individual nodules and flakes on unetched sections at 100 magnification without taking into consideration any difference in size between the particles. The size of a flake or nodule as seen through the microscope depends not only upon the actual size of the flake or nodule, but also upon the manner in which the plane of polish intersects the individual graphite particles. What appears to be a small nodule may actually be a large one that has been sliced through at a relatively large distance from a diameter parallel to the polished section.

As can be seen from the photomicrographs in Fig. 1, true flake and true nodular graphite are not the only shapes of graphite present. In order to simplify the task of estimating the percentages of nodules and flakes, the graphite was classified into two groups only. Such classification required an arbitrary division between flakes and the more compact type of graphite called nodules. Any particle having one dimension at least twice that of the other was considered as a flake.

Many of the particles placed into the category of nodular graphite were not truly spherulitic. These particles were far from the common conception of flakes and did approach a spherulitic shape. On the other hand, the graphite particles classified as flakes were usually not what are commonly considered as flakes. In general, except for the melt with no magnesium addition, the so-called flakes were shorter and more stubby than ordinary flakes. A gray cast iron

TABLE 1—CHEMICAL COMPOSITION OF MELTS

	Composition, %					
Melt No.	6	7	11	13	14	
Total Carbon	3.23	3.20	3.26	3.36	3.43	
Silicon	2.54	2.33	2.32	2.01	1.53	
Sulphur	0.026				0.030	
Phosphorus	0.030			-	0.026	
Manganese	0.22				0.22	
Carbon Equivalent	4.09	3.99	4.04	4.04	3.95	
Per Cent Nedules	92	81	55	19	0	

with the graphite entirely in the form of these stubby flakes would undoubtedly have higher strength properties than a similar iron with graphite in the normal flake form.

It is concluded then from examination of the microstructure, that the magnesium addition, even though small, affects a major portion of the graphite in the irons studied and not only that portion which appears as nodules.

Several areas were counted and re-counted in order to determine the reproducibility of the method of estimation. In no case did the difference between estimations, either on the same specimen or different specimens from the casting of any given iron, exceed 3 per cent nodular graphite.

The appearance of the matrix was fairly constant from heat to heat; it consisted of lamellar pearlite plus small patches of carbide with rims of ferrite enveloping the graphite. Typical photomicrographs of the matrices occurring in the test specimens are shown in Fig. 1.

#### Discussion of Tensile Results

The excellent reproducibility of the tensile data indicates the homogeneity of the materials tested (Table 2). Figure 2 shows the ultimate strength

TABLE 2-TENSILE DATA

Specimen	Nodules, %	Ultimate Strength, psi	Elongation, % in 2 in.
14A	0	17,600	
14B	0	17,400	1.0
13A	19	43,000	3.5
13B	19	41,700	3.5
11A	55	65,500	5.5
11B	55	66,000	
7A	81	71,100	7.0
7B	Unsound	specimen-not tested	
6A	92	74,100	7.5
6 <b>B</b>	92	74,600	7.5
• Fracture	d outside gage	e length.	

values plotted again the percentage of nodules. As can be seen, the ultimate strength increases with increasing percentage of nodules. The curve rises rapidly at first and then levels off. Considering the point at 50 per cent nodules as approximately the end of the rapidly-rising part of the curve and the beginning of the slowly-rising part, it can be seen that the gain in strength in the initial portion is about four times as great as in the final portion.

The reason for this difference can be found by examining the photomicrographs. The greatest change in the shape of the graphite (and also in the tensile strength) appears between the iron with zero per cent and the iron with 19 per cent nodules. The increased strength in the initial portion of the curve appears to be due not so much to the actual production of nodular graphite, but to the formation of stubby flakes instead of ordinary flakes.

A graphite particle in cast iron acts as a notch to concentrate stress at the root of the notch. The stress concentration becomes more pronounced as the angle of the notch becomes more acute. Most of the stubby flakes in the iron with 19 per cent nodules would, therefore, result in a decrease in stress concentration

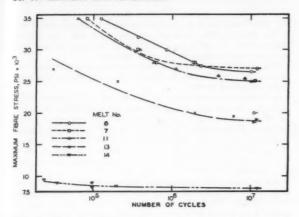


Fig. 3—Stress-cycles curves for the various irons used in the investigation.

at the edges of the particles as compared to ordinary flakes; not as sharp a decrease as if the particles were spherulitic, but nevertheless a decrease. Also, less interruption of the steel-like matrix will occur if the graphite is in the shape of stubby flakes or nodules, because of their greater compactness (lower ratio of surface area to volume) as compared to ordinary flakes.

An examination of the curve of percentage elongation vs. percentage of nodules in Fig. 2 reveals a similar situation. These data are not reliable as the strength data, since precise elongation measurements were difficult because of the poor fit after some brittle fractures. For example, it is well known that gray cast iron has an elongation close to zero, although the value in Table 2 is given as 1 per cent.

## Discussion of Fatigue Data

Table 3 is a summary of the results obtained in the fatigue tests. These data are shown in graphical form in Fig. 3. The general trend of the endurance limit—the limiting stress below which the iron will withstand an indefinitely large number of cycles without fracture—can be seen to increase as the graphite particles become more stubby and as more nodules are formed. In nearly all cases an iron with a higher percentage of nodular graphite will withstand a greater number of cycles to failure at any stress level than an iron with a lower percentage of nodular graphite.

Figure 4 is a curve of endurance limit vs. percentage of nodules. The similarity between the fatigue

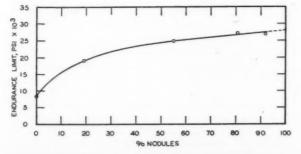


Fig. 4—Endurance limit values vs. percentage of nodules in structure.

and tensile curves is readily apparent and the same analysis as before can be applied.

The endurance ratio, which is the ratio of the endurance limit for completely reversed bending to the static ultimate strength, was calculated for each iron and the values are shown in Table 4. The endurance ratio definitely decreases with increasing percentage of nodular graphite. No explanation can be offered for this behavior.

#### Conclusions

1. A curve relating percentage of nodules and static strength properties reveals that the greatest improvement in ultimate strength and ductility occurs from zero per cent to about 50 per cent nodules. This behavior is a result of the presence of stubby flakes and nodular graphite instead of normal flakes.

2. The endurance limit for completely reversed bending varies with the amount of nodular graphite formation in much the same way as does the static ultimate tensile strength. The behavior can be explained in the same manner.

3. The endurance ratio decreases with nodule for-

4. High values in ultimate tensile strength, ductility, and endurance limit comparable with those for a fully nodular iron may be obtained with a substantial amount of flake graphite present and with irregularly-shaped nodules.

TABLE 3-FATIGUE DATA

Specimen	Maximum Fibre Stress (psi	i) Number of Cycles
14A	23,000	0
14B	10,000	6,000
14G	9,500	24,000
14D	9,000	35,000
14E	9,000	97,000
14F	8,500	199,000
14H	8,250	97,000
14C	8,000	11,949,000*
13A	35,000	2,000
13E	27,000	31,000
13B	25,000	202,000
13C	20,000	1,953,000
13G	19,500	6,043,000
13F	19,000	11,491,000*
13D	18,000	10,781,000*
11A	35,000	64.000
11B	30,000	388,000
11D	28,000	581,000
11E	27,000	1,135,000
11F	26,000	3,888,000
11G	25,500	8,422,000
11C	25,000	10,000,000*
7A	35,000	83,000
7B	30,000	369,000
7D	28,000	635,000
7G	28,000	1,950,000
7F	27,500	2.325.000
7E	27,000	12.284.000*
7C	25,000	11,245,000*
6A F	ractured before 50,000-psi stress	was completely applied
6B	35,000	125,000
6E	32,000	372,000
<b>6F</b>	30,000	838,000
6G	28,000	1,847,000
6H	26,500	10,269,000*
6C	25,000	11,339,000*
6D	20,000	11,092,000*
• Specia	men did not fracture.	

TABLE 4-ENDURANCE RATIOS OF MELTS

211212 2 2					
Melt No.	6	7	11	13	14
Endurance Ratio	0.36	0.38	0.38	0.45	0.46

# Acknowledgment

The authors wish to express their appreciation to The Pennsylvania State College for permission to publish this work, which formed the subject of a thesis written by Alvin Shames in partial fulfillment for the degree of Master of Science in Metallurgy.

The pig iron was supplied by the Bethlehem Steel Co., and the magnesium alloy was supplied by the Vanadium Corporation of America. Thanks are also extended to D. E. Krause, Executive Director of the Gray Iron Research Institute, Inc., Columbus, Ohio, for advice given in the early stages of the investigation, and to C. T. Greenidge of Battelle Memorial Institute, Columbus, Ohio.

# DISCUSSION

Chairman: J. E. REHDER, Department of Mines & Technical Surveys, Ottawa, Ont., Canada.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

Recorder: C. F. WALTON, Case Institute of Technology, Cleve-

R. A. FLINN: 1 Physical properties are strongly influenced by the difference in the matrix structure. Photomicrograph B (Fig. 1) appears to have about 60 per cent ferrite whereas Photomicrograph C (Fig. 1) has only 20 per cent ferrite.

The addition of data on Brinell hardness and combined carbon would be helpful.

Section size has a definite effect on the percentage of nodular and flake graphite.

All of the castings could have been made ferritic before testing. MR. LINDSAY: It is agreed that there is a distinct difference in the appearance of the matrix in melt number 13 as compared to the other matrices. It might be said that it was anticipated that questions such as those posed by Dr. Flinn would arise. It is true that the influence of he graphite shapes presented here would be more truly reflected in a constant (ferritic) matrix. If the matrices were made pearlitic, then the tensile strength value would be increased and the elongation value would be decreased to a greater extent in melt 13 (and to a lesser degree in melts 7 and 6) as compared with melts 14 and

11. This would serve to change the shape of the curves in Fig. 2, but would hardly invalidate the conclusion that "the greatest improvement in ultimate strength and ductility occurs from zero per cent to about 50 per cent nodules."

The addition of data on Brinell hardness and combined carbon would merely serve to confirm that the matrix difference under discussion is real.

It is common knowledge that section size has a definite effect upon the percentage of nodular and flake graphite. It would be our contention that the property differences caused by the variation in percentage of nodular graphite would be found regardless of whether the variation was produced by changes in the nodularizing treatment or by changes in section size.

J. T. MACKENZIE: 8 What was the magnesium content and the modulus of elasticity of the material?

MR. SHAMES: The magnesium contents of the irons were not determined. If they had been, they would have been reported in the paper. Our main concern was in producing a range of graphite structures with increasing percentage of nodules. This was done by increasing the amount of magnesium addition agent, with no magnesium being added to melt 14. It would be expected that an increasing amount of magnesium would be found to have been retained with increasing amount of addition

GOSTA VENNERHOLM: 3 Tensile strength alone is not a sound basis for evaluation of a material. The shape of spherulites can vary the modulus of elasticity from 16 million to 28 million.

RICHARD SCHNEIDEWIND: 4 The difference in silicon content (Table 1) could have a profound influence on the physical properties. The copper content would also be involved. The effect would be present even if the samples had been annealed.

MR. LINDSAY: The comments by Mr. Vennerholm and Dr. Schneidewind were taken as not requiring a direct reply. It was interesting, but not unexpected, to learn of the difference in modulus of elasticity caused by variation in shape of the spherulites. It was known from some unpublished work in which one of the authors was involved that increased flake size in a constant matrix caused a definite decrease in modulus. This would lead to the expectation that lack of perfection of spherulitic shape would cause a decrease in modulus.

It is agreed that the differences in silicon and copper contents would have a bearing on the properties. The authors used the iron-copper-silicon-magnesium addition agent only because we were informed that its reaction would be least violent. This was an important consideration from the safety standpoint, since the melts were made in an area to which many people had access while melting was being done.

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# QUALITY CONTROL IN A MALLEABLE IRON FOUNDRY

By

# E. F. Price* and O. K. Hunsaker*

Statistical quality control, when employed in foundry operations, has many applications which will reduce scrap and rework and at the same time produce a high quality casting. This is accomplished by controlling each phase of foundry operations and each step in the production of the casting through the use of control charts. When each operation which goes into making a casting is properly controlled, then the finished product is of good quality, and produced with a minimum of scrap and cost. It is in this field of scrap reduction and customer rejects that quality control becomes a profitable tool for the foundry.

# Principles of Quality Control

Three general types of charts are used in the application of quality control to foundry operations. The first of these is the frequency distribution curve, or the histogram. In this case the dimension, or unit of measurement, is plotted against the number of occurrences, or frequency. Figure 1 shows such a frequency distribution for a 0.52 in. dimension boss of a casting. The normal curve is bell-shaped; if it takes any other form it is abnormal. The amount of variation from the average, or center, of the curve (i.e., the width of the bell) is known as the dispersion.

This dispersion is best calculated as the standard deviation, which is the root-mean-square deviation from the average, and is called sigma  $(\sigma)$ :

$$\sigma \; = \; \sqrt{\frac{\Sigma X^2}{n} \; - \; \overline{X}^2}$$

where X is the individual measurement,

 $\overline{X}$  is the average of the group,

n is the number of readings in the group. As shown on the curve,  $\pm 1\sigma$  represents 68.26 per cent of the area of the bell;  $\pm 2\sigma$  represents 95.46 per cent of the area of the bell; and  $\pm 3\sigma$  represents 99.73 per

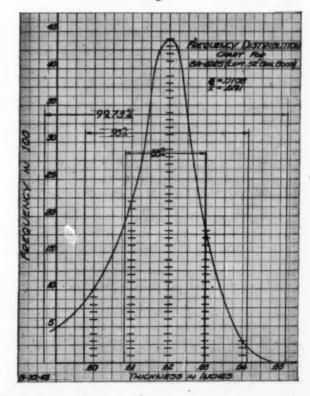
cent of the area of the bell.

These frequency distributions are useful in deter-

mining whether a dimension of a group of castings is normal, and whether the  $\pm 3\sigma$  limits (i.e., 99.73 per cent of the castings) are within the customer's specification.

The second type of chart used is the  $\overline{X}$  and R chart. In this case, the  $3\sigma$  limits mentioned above are converted mathematically and applied as upper and lower control limits. Figure 2 shows such a chart. Each of the individual points on the chart is an average of a group of readings. (In this case the size of the sample group is five.) This average is known as  $\overline{X}$ . The difference between the highest and lowest reading of the group is called the range (R), and is plotted as a point on the range chart. The size of

Fig. 1



^{*}Chief Inspector and Assistant Superintendent, respectively, Dayton Malleable Iron Co., Ironton Plant, Ironton, Ohio.

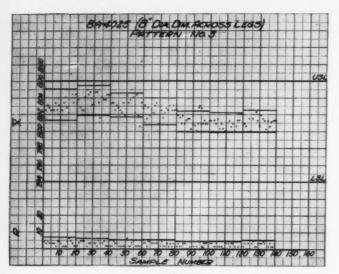


Fig. 2

the sample group can be varied according to the need. It should be noted that Fig. 2 is a combination of two charts, i.e.,  $\overline{\lambda}$  chart and R chart, and plotted as shown because they are interdependent.

The upper and lower control limits are calculated for every group of 20 points plotted on the  $\overline{X}$  chart. These control limits may be calculated for any group size, or they may be calculated as permanent control lines, but they are most conveniently calculated as shown in Fig. 2, in groups of 20, as will be discussed later. In the calculation of the control limits the average  $\overline{X}$  for the 20 sample groups is calculated and called  $\overline{X}$ . The average range for the 20 sample groups is calculated and called  $\overline{X}$ . The following formulae are then used:

 $\overline{X}+A_2\overline{R}$  = the upper control limit (U.C.L.) for  $\overline{X}$   $\overline{\overline{X}}-A_2\overline{R}$  = the lower control limit (L.C.L.) for  $\overline{X}$  $D_4\overline{R}$  = the upper control limit for the range

 $D_3\overline{R}$  = the lower control limit for the range

The constants A₂, D₃, and D₄ used in the formulae above vary according to the size of the original sample groups. For the original sample size of five pieces shown in Fig. 2 the value of these constants is:

 $A_2 = 0.58$   $D_3 = 0.00$   $D_4 = 2.11$ 

The physical and dimensional specifications of the castings are based on the customer's requirements. When these upper and lower specification limits are plotted on the chart, the chart is complete. The dimension is said to be in control if the points lie between the upper and lower control limits, and the parts will meet specifications if the upper and lower control limits lie between the upper and lower specification limits.

The third type of control chart is the fraction defective or (p) chart. In using this type of chart the fraction of defective parts in a sample group is plotted on a chart as shown in Fig. 3. The fraction

defective (p) is calculated as the ratio of defective pieces found to the total number of pieces inspected. This type of chart can be used to record a wide variety of data economically. It can be used to record data collected in normal inspection methods, or in evaluating data that can only be observed as attributes. Its best use in foundry operations is in evaluating data obtained from the use of go and no-go gages, and casting defects found in visual inspection. The fraction defective may be converted to per cent defective by multipling by 100, i.e., 100p.

Figure 3 is a typical (p) chart in which the fraction defective of the pins and bushings of a set of cope and drag flasks is recorded. The control limits are shown as heavy lines. They are calculated with

the following formulae:

Upper control limit for p (U.C.L.p)= $\overline{p}+3$   $\sqrt{\frac{\overline{p}(1-\overline{p})}{n}}$ 

Lower control limit for p (L.C.L.p)= $\overline{p}$ -3  $\sqrt{\frac{\overline{p}(1-\overline{p})}{n}}$ 

p is the fraction defective (ratio of defective parts to the total number inspected).

p is the average fraction defective for the period for which control limits are calculated. (p is calculated as the total defective divided by the total quantity inspected).

n is the size of the sample (i.e., the number of pieces in each sample lot).

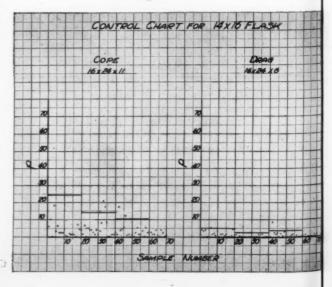
The three general types of charts mentioned above will cover most of the applications of quality control to foundry operations.

# Control of Raw Materials

Control of raw materials is the first step in the production of quality castings. All raw materials should be purchased from reliable firms, and a system of checking the incoming material to specifications is essential.

Foundry sand, whether ordered for the molding





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sand system or for the coreroom, should be purchased to some specified grain distribution. When sand is received, a random sample is taken from the car, and a grain distribution is made in the sand laboratory. Figure 4 shows such a distribution made on incoming sand. The actual grain distribution is not so important as the maintaining of a constant grain distribution from one shipment to another. Too much variation in grain distribution between shipments makes subsequent sand control impossible. When sufficient data are accumulated to show that a supplier consistently meets specifications, then the number of checks made on his sand may be reduced.

In order to control pig iron, a sample is taken from each car of pig and silvery iron upon arrival and sent to the laboratory for an analysis for the following: chromium, manganese, silicon, phosphorus, carbon, and sulphur. This serves as a double check since the vendor also supplies an analysis with each shipment. If the sample meets specification the car is unloaded.

If there is an element which is outside of specification, that shipment is kept apart and fed into the charge in small quantities as the analysis of the melt permits.

Scrap iron is difficult to control because of its inconsistent nature. Good scrap should meet chemical specifications, be of proper size for use and reasonably free of rust. In selecting his sample for chemical analysis the yard foreman must be familiar with the type of scrap he has been receiving, and must select pieces with which he is not familiar for special analysis. If the car contains a uniform load, such as a load of rail, three or four pieces are sufficient for a sample. If the load is mixed as many as 12 to 15 samples may be needed.

As in the case of pig iron, any part of a shipment that is outside of specifications, but usable, is separated and fed into the charge in small amounts when analysis of the melt permits. The supplier is then notified, so that this type of scrap will be eliminated on future shipments. If the scrap is not usable it is returned to the supplier.

Coal and coke are ordered to chemical and physical specifications, and samples are taken periodically to insure that these specifications are being met. They are checked chemically for sulphur, fixed carbon, ash, and volatile. In addition to the chemical analysis the coke is checked for size and cell space. All the coke is screened for size.

The limestone is also sampled for a chemical analysis. The following is a typical analysis of the stone used:

	Per Cent		Per Cent
SiO ₂	0.94		
Al ₂ O ₃	0.34		
$Fe_2O_3$	0.78		
CaO	48.95	CaCO _s	87.34
MgO	5.07	MgCO ₂	10.60
ignition loss	43.92		
(CO ₂ 43.92)			

The limestone is selected for its purity. Stones high in silicon oxides, aluminum oxides, and iron oxides are rejected.

# **Melting Control**

In melting operations, the quality control charts make maximum use of the data available to the melting foreman. When the chemical analysis, slag color, weather conditions, sprue fracture, and charging data are all plotted together in an orderly fashion, they present to the melting crew a clear-cut picture of what is taking place in their operations.

Figures 5 and 6 show the melting charts for April 27, 1950. These charts cover a duplex melting operation of two cupolas and one air furnace. Figure 5 is the furnace control chart on which is plotted the carbon, silicon, temperature, and allied data.

The carbon content is checked every half-hour, and the results are plotted on the charts as shown in Fig. 5. Although this is not an  $\overline{X}$  and R chart, each group

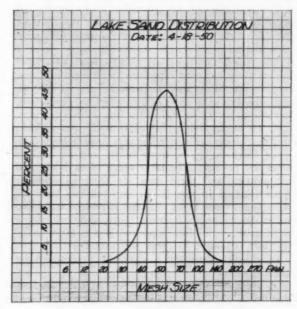


Fig. 4

of 10 successive readings is averaged for  $\overline{X}$  and R. The upper and lower control limits are calculated for each group of 20 readings of  $\overline{X}$  and R, and plotted on these charts as a guide for the operators.

The silicon content is checked every hour and plotted as shown on the chart. The upper and lower control limits are calculated in the same manner as for carbon above.

The manganese (bottom of Fig. 5) is checked every hour and the control limits are calculated in the same manner as for the carbon and silicon.

The temperature of the air furnace is plotted as shown in Fig. 5. The lines at 2785 F and 2795 F are not control limits, but specification limits established from experience.

The chart shown in Fig. 5 contains some additional information: the silvery iron additions to the air furnace are shown just above the silicon chart. The spiegel additions are shown just above the manganese. The sprue fracture is shown at the top of the chart, and the melting data is recorded on the margin.

Figure 6 is the cupola chart covering the same period as the furnace chart of Fig. 5. This is not a true quality control chart in the sense described earlier in the article. It does assist in the control of the melting operations, however, and in that respect it is a control chart.

In this chart the slag color is plotted against time for both cupolas. The slag color is listed as light green, dark green, and black. By running several hundred slag analyses taken from various conditions of melting it was possible to determine an approx. relation between color and chemical composition.

The normal slag color was found to be light green; increased impurities, such as iron oxide due to oxidizing conditions, caused the slag to turn dark. There are other factors in the melting operation, however, which affect the slag color, making it impossible to detect an oxidizing slag by color alone. When all other factors remain constant, it is not the color of the slag, but a change in the color which is indicative

The cupola control chart also contains additional information: barometric pressure, weather conditions, relative humidity, and temperature are all plotted, not only because of their effect on melting operations, but also to serve as a record for future investigations.

As an additional means of control, sprue blocks and test blocks are poured. Every hour a large 3-in. sprue block, a 11/2-in. sprue block, and an analysis * block are poured. The large sprue block is checked for mottles, and these data are recorded opposite sprue fracture in Fig. 5. Twenty minutes later another small sprue block (11/2-in.) is poured and broken to determine its fracture. Forty minutes later a third small sprue block is poured, and another analysis block is poured to check for carbon.

In addition, three test blocks are poured which are numbered for the hour poured, and put into the anneal ovens ahead of the castings. When annealed, they are checked for Brinell hardness number and microstructure as a preview of what the iron to fol-

low will be like after annealing.

In the melting operations for April 27, 1950, the slag color darkened between 10:00 a.m. and 11:00 a.m. and, since there was no apparent reason for the change, a 200-lb coke rebed was charged at 10:45 a.m. The carbon, silicon, and temperature started dropping just after 11:00 a.m., and went out of control until after 1:00 p.m. when the coke rebed began to take effect. By watching the slag and temperature closely, the foreman was able to rebed his cupolas in time to prevent a serious oxidizing condition. In general, these charts show very good control and maximum use of statistical quality control.

#### Coremaking

Since many of the critical dimensions of a casting are made by the core, a statistical check of the core box is the most direct approach to their control. The wear of a core box is very gradual, and for that reason the box is measured only once each day that it is in operation. Figure 7 is a typical control chart for a critical dimension made in a core box. This is not an X and R chart, but a simple plot of each day's measurement with upper and lower specification limits drawn on the chart as a guide.

The box is a three-part core box, and the dimension shown in Fig. 7 is made in each of the two bottom parts. The dimension for each part is plotted with a different symbol. In most cases a dimension made in a core box can only wear larger. Therefore the dimension should be made to the lower specification limit to allow for maximum wear before repair or replacement is necessary. In Fig. 7 the box was made to the lower specification limit as shown in the first eleven readings.

The dimension on the castings, however, did not conform to specifications, and it was necessary to alter the core box as shown by the 12th reading. There was no further change until the 73rd reading when the core box began to show wear. When a box begins to show wear it should be closely watched. No action is needed, however, until the upper specification limit is reached. The 65th reading shows a sudden drop which would indicate that metal had been added to the box. This discrepancy, however, was found to be an error on the part of the inspector.

In addition to measuring the core box, the inspector examines the box for bad blow tubes, worn or loose pins and bushings, and other defects.

After each operation such as blowing, baking, cleaning, and pasting, a statistical sample is removed from the rack and checked by the inspector. The foreman is notified immediately of any discrepancies so that corrective action can be taken.

After the cores are baked, a statistical sample is taken from the rack and checked for hardness. The results are plotted on an X and R chart. If the hardness of a core goes out of control, the foreman is notified so that he can check the sand mix and ovens.

## Molding Operations

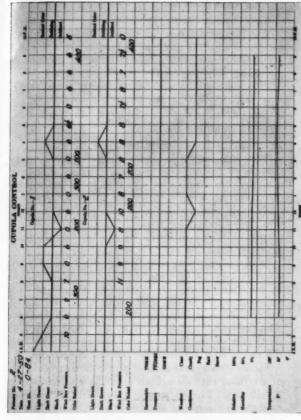
Quality castings can be made only from quality pattern equipment. Before the pattern equipment is put into production it must be made to the blueprint and properly gated to produce a sound casting. Sufficient samples must be made to check both of these conditions. When this has been accomplished, quality control is a good tool to use in keeping the equipment in condition.

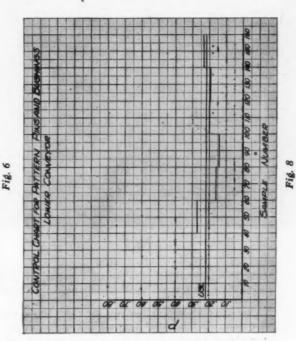
The pins and bushings on the pattern plate probably receive more wear than any other part of the pattern equipment and, because they control cope and drag shift, they are checked four times a day, and the results are plotted on a (p) chart. Figure 8 is a (p) chart showing the fraction defective of pins and bushings for patterns on the lower conveyor. There are five cope and drag stations on this conveyer line, making a total of 20 pins and bushings, which is the

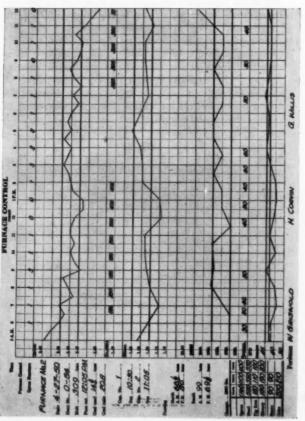
basis for calculation.

The pins and bushings are checked with a go and no-go gage, and the fraction defective plotted as shown. The foreman on this line is then notified, so that the defective pin or bushing can be replaced. When this chart was started, the fraction defective was as high as 0.80, and there was very little improvement for 60 sample checks. This condition was due to patterns being changed, and the new patterns

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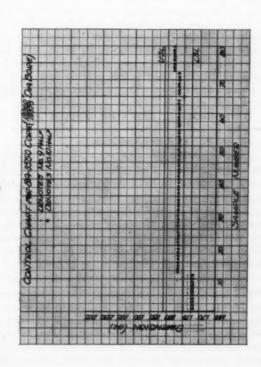


Fig. 7

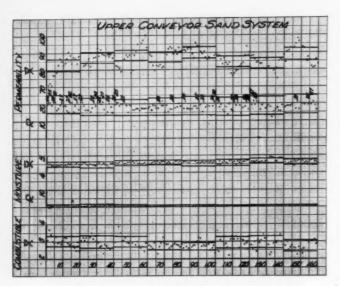


Fig. 9

being put into operation with defective pins and bushings. This was corrected by checking the equipment thoroughly before putting it in operation.

After the 60th sample number the chart shows very good control. The solid line at 0.20 was arbitrarily chosen as the upper specification limit, and the later sample groups show that control can be maintained below this level.

When the inspector is checking the pins and bushings, he also checks the equipment for loose risers, loose core prints, loose patterns, and visible defects. The foreman is immediately notified of any defects.

The flasks are also checked for defective pins and bushings. Figure 3 is a (p) chart for defective pins and bushings for one of the 16 by 24 in. tight flask sets. Since there are 75 sets of flasks as compared to only one pattern, there is less wear on flask pins and bushings, and therefore they are checked only once a week.

When these (p) charts are properly used, and prompt action is taken to correct defects, then the scrap loss due to shift is reduced.

The molding sand is checked for permeability, moisture, bond and combustibles. A sample is taken every 15 min and the results plotted on an  $\overline{X}$  and R chart. Each group of four is averaged for  $\overline{X}$  and R, and the upper and lower control limits are calculated for every 20 groups. The only exception is the check for combustibles which is made four times a day. Figure 9 is the control chart for permeability, moisture and combustibles.

The permeability is kept in control between 80 and 100. The  $\overline{X}$  chart for permeability shows several points out of control. This variation in  $\overline{X}$  is due to improper use of the blower in removing the fines in the system. This condition was gradually improved, and fairly good control was maintained between the 160th and 180th samples. Subsequent charts show even further improvement. As a further aid in the control of permeability, a screen analysis of the

foundry sand is plotted, and the frequency distribution carefully watched for variations.

The  $\overline{X}$  and R chart for moisture in Fig. 9 shows good control. The moisture is held between 4 and 5 per cent because of moisture loss due to the heat of the sand.

The combustibles are held between 3 and 4 per cent. The  $\overline{X}$  chart in Fig. 9 shows several points out of control because of insufficient mulling capacity to properly condition the sand.

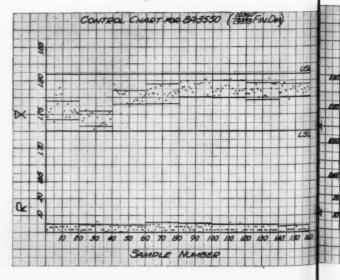
#### Hard Iron

Hard iron inspection and quality control serve the purpose of finding defects in the castings as soon as possible, thus eliminating the expense of subsequent operations on scrap castings. A statistical sample is taken from the production of each pattern in operation, and checked for critical dimensions, core shift, pattern shift and core location, and is magnetic particle inspected for invisible defects.

Critical dimensions are checked closely so that any discrepancies can be corrected before large quantities of defective castings are produced. All critical core dimensions are checked in the hard iron, even though the core boxes are checked for the same dimensions. This serves a double purpose, since the core may settle and become out of round, or the box may have too much tape, resulting in oversize dimensions.

Figure 10 is a control chart which shows an example of a critical core dimension which is also checked in a core box. The dimension checked is a cored diameter which must be not less than 1.725 in. because an arbor must pass through this diameter in the first machining operation. The heavy lines indicate the upper and lower specification limits, and the light lines indicate the upper and lower control limits. Starting with the 40th reading, each successive set of control limits shows that the dimension is getting larger. This actually reflects the core box wear. No action is required until the upper control line reaches the upper specification line. Then the box must be either repaired or replaced.





The calculation of the upper and lower control limits for every group of 20 points on a control chart gives a clear picture of what is happening in an operation. For instance, the upper control limit reached the upper specification limit between the 80th and 100th sample groups, showing that the box needed repair. The box was repaired and the result can be seen in the subsequent control limits.

Figure 11 is the control chart for another critical core dimension which represents the distance between the turning-radius stops of a steering gear housing. The customer set the upper and lower specification limits as shown, but asked that the dimension be held

as close as possible.

The 13th sample is out of control on both the  $\overline{X}$  and range charts due to crushed cores. It can also be noticed that there is a tendency for the dimension to grow. This is due to the fact that the two stops are in both halves of the core box, and the tape used to prevent blow-out sometimes gums up, causing the two halves of the box to spread. The control limits for samples between 140 and 160 show a slight drop because a new box was put into operation.

Figure 12 is an X and R chart for the 1.00 in. width dimension of a bearing cap. This dimension is controlled by the pattern and the molding operation. The specification limits are established by the customer, who coins this dimension in his first operation. This dimension went out of control on the 19th sample, and finally went out of specification on the 21st sample due to swelled castings.

In checking for the reason behind this condition, it was found that the patterns were to size and the sand was in control. The real cause was insufficient jolting on the part of the molder in an effort to increase his piece work. The pattern was made undersize on this dimension to compensate for the swelling effect, starting with the 33rd sample.

Figure 13 is an  $\overline{X}$  and R chart for pattern shift. In this case the upper specification limit is set at 0.035 in. This chart shows very good control for cope and

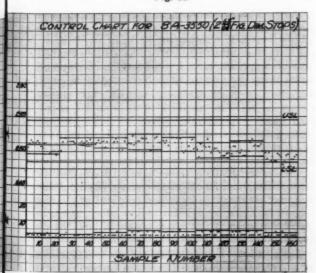


Fig. 11

CONTROL CARPY FOR 15355 (1 DIM)

Fig. 12

drag shift. By carefully checking for shift in the hard iron, costly losses due to shift can be greatly reduced. Since a quality control check was started on the pins and bushings, not one casting made on the tight flask line has been scrapped for cope and drag shift due to defective pins and bushings.

Every day a sample of five castings is taken from each pattern in operation, and checked for invisible defects on the magnetic particle inspection machine. This gives a fairly good check for hot tears, shrinkage, fine sprue cracks, etc., which would normally pass visual inspection. If defects are found, scrap castings are broken to determine the exact nature of the defect so that immediate correction can be made. In addition, scrap castings are broken in an effort to locate any internal defects in the casting.

At the end of each shift a special form is filled out listing each part number and the percentage of scrap, and listing the cause of such scrap. These forms are sent to the foremen so that action can be taken to reduce the scrap on the following shift.

# Annealing Control

Before each heat of iron is out of the annealing ovens the test bars for that heat, which were poured hourly, have been annealed. Every bar is tested for Brinell hardness number, and samples are cut for microscopic examination. If the bars do not anneal properly, the oven cycles can be adjusted to accomodate the iron, or the iron separated for special annealing.

As the castings come out of the anneal ovens, five pieces are taken from each push for a Brinell hardness test. Figure 14 is an  $\overline{X}$  and R chart for Brinell hardness for oven number one. The upper specification limit of 149 BHN is shown as a heavy line. There are a few points which are over the specification limit, and this iron was separated and re-annealed.

Figure 15 is a hardness chart for the same oven, but the data are plotted for a later period. Whereas the control in both charts is good, the iron shown in Fig. 15 is much softer. This was brought about by closer melting control and closer control over the annealing

There is a test bar in each push of iron which is broken to observe the fracture of the iron. These bars are placed in a rack for reference purposes. If the bars from one of the ovens begin to show bright spots when the bars from the other ovens look good, then that oven is checked to determine the cause for improper control. As an additional measure of control, the temperature from each zone in the oven is plotted periodically on a chart as a check on the annealing cycle.

The tensile bars, which are poured with every heat are annealed along with the castings, and after they are pulled, the data are plotted on an X and R chart.

Figure 16 is the  $\overline{X}$  and R chart for the percentage of elongation. The iron is grade 32510 malleable, which calls for a minimum of 10 per cent elongation, and therefore the lower specification limit is shown on the 10 per cent line. As an assurance that the iron will always meet specifications, the lower control limit is held at about 12.5 per cent. Samples 54 and 82 are out of control due to defective tensile bars.

The ultimate strength and yield point data are also plotted on X and R charts.

# Soft Iron Control

In the soft iron department the gates are sheared and ground, and the castings cleaned. Since many gates appear on chucking or locating surfaces or machining surfaces, they must be removed as closely as possible. A sample is taken of the castings at each grinder or shearing press to be checked for proper gate and fin removal. The results are recorded on a standard form and plotted on a (p) chart. The foreman is notified immediately of any discrepancies and given a copy of the inspector's report. Gates which are not located on control surfaces are merely checked for appearance.

In the finishing department the castings are die-

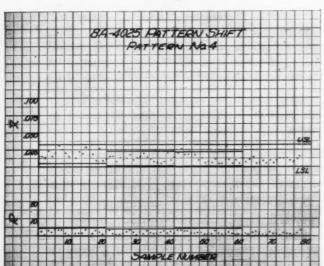
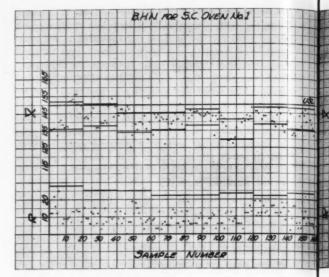


Fig. 13

QUALITY CONTROL IN A MALLEABLE IRON FOUNDRY



straightened, given a final inspection, tested for leaks, and painted. As the castings are being straightened, a sample is taken for checking all critical die-controlled dimensions. These dimensions are plotted on an  $\overline{X}$ and R chart as a check on the die. When the chart begins to show lack of control or failure of the dimension to meet specification, then the die must be rebuilt. Figure 17 is an X and R chart for a locating boss of a casting. The initial upper control limit lies outside of the upper specification limit because of improper pressure on the die.

Finishing Department Control

After the 10th sample this condition was corrected. The upper control limit for the 100th to the 120th samples lies above the upper specification limit because of excess metal due to a raised dowel pin in the pattern which was located in the center of the boss. This condition was corrected, but these castings were mixed in the system, which caused the control limits to rise between the 100th and 120th samples. This necessitated a 100 per cent gaging of all of these castings and the excess metal was ground off the boss before shipment.

Figure 18 illustrates die wear. The first two samples are out of specification. The die was badly worn and a new insert put into the die, which resulted in a sudden drop in  $\overline{X}$  at the third sample. The change shown in the 106th reading resulted from a new pattern which was put into operation. These castings were put through the die separately, resulting in a sharp rise, as compared to the gradual rise shown in

The control limits were held along the lower specification limits on the old pattern at the request of the customer because this dimension is coined in the first operation. As a result of an engineering change to the new pattern, it was not necessary to hold the dimension to the lower specification limit.

Figure 2 is an  $\overline{X}$  and R chart for the 8-in. diameter of the pedestal legs of a differential carrier. This is

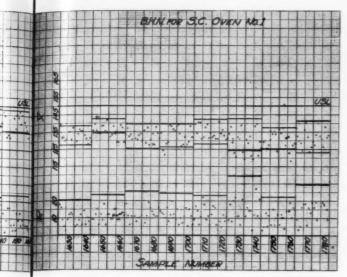


Fig. 15

an unfinished dimension. The customer, however, takes a clearance cut for assembly purposes. This is an intermittent cut which is hard on the tools, and for this reason an effort was made to tighten the control limits and hold the dimension as close as possible. By removing some metal from the pattern and adjusting the die, the diameter was brought under closer control starting with the 80th sample reading. This change resulted in increased tool life in the customer's plant.

Before a new pattern is placed in production, about 100 samples are poured for a survey of critical dimensions, and for each dimension a frequency distribution is plotted. Figure 1 is a frequency distribution for the 0.52-in. dimension boss of a casting. The rough casting measures 0.62 in. because of 0.100 in. for finish.

In this case the distribution is good and the three  $\sigma$  limits will fall within the specification limits. When these surveys show that the dimensions are correct, the patterns are put into use for production purposes. These surveys are also helpful in checking for discrepancies the customer might find, for checking casting weights, and in determining the amount of metal to be added or removed from a pattern to make the casting dimension correct.

Figure 19 is a control chart for core location. The casting is a differential gear case in which the cored dimensions must be held in close relationship to the pattern. When the castings are ready to ship they are given an outgoing quality check by gaging a statistical sample from each box.

The gage is made to the customer's specification, with steps to show the actual position of the core in relation to the flange. These steps are numbered, and Fig. 19 is a histogram-type chart in which the gage steps are plotted against the frequency in each sample group. The upper and lower specifications are shown as solid lines. If there are castings outside of the specifications, all of the castings in that box are checked with the gage. Charts of this type are particularly

well adapted to step gages, just as (p) charts are well adapted for use with go and no-go gages.

In addition to checking a statistical sample for specified dimensions as an outgoing quality check, the samples are also inspected for visible defects. This procedure serves as a final assurance that all operations have been performed on the castings, and reduces scrap in the customer's plant.

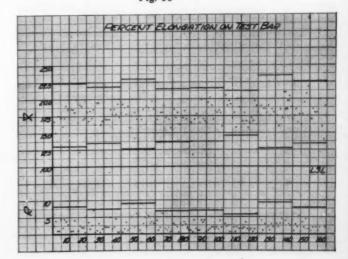
Differential carriers and castings of that type are all tested for leaks under air pressure. The castings are put under pressure and lowered under water so that the leaks are visible and readily located. A (p) chart is used to record the leakers. If one of the (p) charts shows a sudden increase in leakers, a survey is made to determine the cause and the exact location of the leak. Another item which must be carefully watched is scrap. The total scrap is averaged weekly and plotted on a (p) chart. If there is an increase in the scrap, a survey is made to determine the reason, and corrective action is taken immediately.

This same procedure is followed in controlling rework and welding. The success of the quality control program depends upon producing a higher quality casting with a lower scrap and rework, and for that reason it is imperative that any increase in scrap or rework be corrected immediately. A (p) chart is also used to record the scrap returns from the customer's plant, and a survey made to investigate any increase in scrap shown on this chart.

A large percentage of the castings must be painted to the customer's specification. This is done by dipping the castings in a paint tank by means of a conveyor. The paint in the tanks is checked periodically for viscosity and solids content. The viscosity is checked by means of a cup-type viscosimeter and plotted on an  $\overline{X}$  and R chart. With this procedure the paint is always maintained to the proper specifications.

As a further means of control the paint is checked for viscosity and solids content upon being received from the supplier, and the data recorded according to the supplier's batch number. The thinner is checked for evaporation rate when it is received. If a ship-

Fig. 16



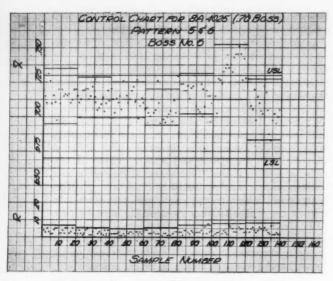


Fig. 17

ment of paint does not meet specifications, the supplier is notified so that proper action can be taken.

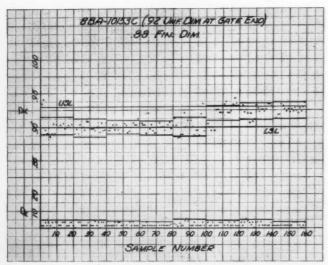
As a final check on the casting, a routine layout is made of all dimensions. At least one casting from each pattern in production should be sent to layout every week for a complete check.

## Conclusion

Statistical quality control is not new, and its success in manufacturing operations is well known. Therefore, its value in foundry applications should come as no surprise. Quality control is not a cure-all which will make good operations out of bad operations, or good castings out of bad castings, but the charts will indicate when action should be taken to correct an operation or repair a piece of equipment. The charts merely indicate, they do not do the work of correcting.

The success of the quality control program depends upon four factors:





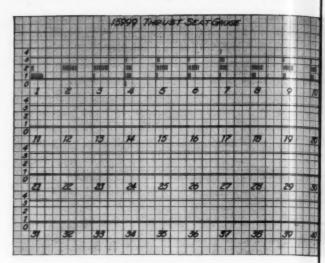


Fig. 19

1. Collecting the proper data.

- 2. Relaying the data to the proper authority for action.
- 3. Prompt action as a result of this information.

4. Close cooperation among all members of the organization.

It is also necessary to have a clear understanding of the customers' requirements as well as a thorough knowledge of foundry problems, so that the charts can be used to the maximum advantage.

#### References

- 1. E. L. Grant, Statistical Quality Control, McGraw-Hill Book Co., Inc., New York (1946).
- 2. Frederick C. Mills, Statistical Methods, Henry Holt and Co., New York (1938).
- 3. Edward S. Smith, Control Charts, McGraw-Hill Book Co., Inc., New York (1947).

# DISCUSSION

Chairman: A. A. Evans, International Harvester Co., Indianapolis.

Co-Chairman: GEO. VER BEKE, John Deere Malleable Works, East Moline, Ill.

Recorder: H. H. JOHNSON, National Malleable & Steel Castings Co., Sharon, Pa.

T. L. Nuzum (Written Discussion): 1 In reading this paper, or any article dealing with Statistical Quality Control, I believe that the first impression is the necessity of teamwork of all departments working toward the ultimate goal. The best quality attainable in the finished product should be sold to the customer.

This can be malleable castings or any other manufactured article.

Therefore, management must be sold on the benefits to be derived from this program before it is started and see that interdepartmental cooperation is given to the Quality Control Department for satisfactory results. Good cores set in defective molds or defective cores set in good molds will both produce bad castings. Control charts showing proper evaluation of sand mixing, core making practices, and molding practices will point the finger to the cause of defective castings.

Quality Control is not new, but Statistical Quality Control as we think of it today is new to many organizations. Control bands set up from practical experience in each of these fields and

¹ Pattern Engineer, Vendor Contact Repr., Salisbury Axle Works, Daga Corp., Fort Wayne, Ind.

the operation held within these control bands will produce the desired results. Control band on charts can be opened or tightened as experience directs.

At the conclusion of this article we are impressed with the necessity of understanding exactly what the chart indicates and taking corrective action to prevent a recurrence of the conditions that permitted the operation to go cut of control as shown by the chart.

I know of no better example of manufacture than a foundry with so many variables, that, with the proper knowledge of the control each department has over their part of the operation, the ultimate quality of the product can be predicted before cast-

ings actually are completed.

I have listened with interest to many speakers talking to our local Quality Control Society and one outstanding remark made by a prominent Works Manager for a large industrial plant in our city was to the effect that Quality Control consisted of 85 per cent hard work, 10 per cent good judgment, and 5 per cent mathematics. I am bringing out this point for the reason that many of your people will be confused and annoyed before they start by the mathematics unless this is explained to them in detail. Charts and control limits should be set up by someone experienced in this work and familiar with the operation. The actual charting of the results in the lot selected to be inspected can be done by any careful inspector after he is properly instructed. Care is necessary lest the chart be inaccurate and the decision made from the chart be wrong due to carelessness.

It is also true that in many instances a speaker does not want to make public the particular troubles they are having in their plant, and therefore speaks in generalities rather than concrete instances of fact. In order to best illustrate to people not familiar with the work, it is necessary to take an operation with which they are familiar and go through the process together.

We have found in our plant that it was surprising the interest taken by the men in our organization in a class organized to meet in our own plant and discuss our own problems freely. This class is free and is attended by foremen, inspectors, and operators alike. Attendance is not compulsory, but the response was so great that the one class had to be divided into five groups for better results and discussion. Classes are conducted by our Supervisor of Quality Control and Chief Inspector.

We all know that castings in the true sense of the word are not to the blueprint. Therefore, it is of particular importance, as these co-authors have brought out, of the necessity of knowing the physical and dimensional specifications required by the customer. These are then set upon charts with practical limits

for inspectors to use.

Each machine line has locating or starting points that must be kept uniform within certain tolerances, before castings will go down a production line satisfactorily. If by a periodic spot check castings when inspected fall within the control bands you can be certain they will be satisfactory to the customer.

These locating points and machine lines are as varied as the equipment and methods of process used by the vendors of cast-

ings in the malleable industry.

For this reason I wish to insert here a statement that is well

worth consideration when any new job is started.

The buyer of castings should supply the proper representative of the foundry, marked prints of the casting to be purchased, showing locating points for the machine line on rough surfaces to start the machine work. When gages are used to check finish, core location, or any other pertinent dimension, these gages should be exactly alike. Made from the same tool drawings. A great deal of confusion can be caused by the vendor checking one way and the customer another. Both believing their method is right. This is corrected by the use of Quality Control methods. On new jobs, by close cooperation between vendor and customer these control band limits can be checked by experimentation until the correct answer is found.

Another point to be brought out in connection with this same line of thought. If the buyer of castings knows where he will purchase these castings, the proper representative of the foundry should be contacted and their requirements as to type and design of patterns, core boxes, and core driers when necessary be ascertained before pattern equipment is made. Pattern shops making pattern equipment should always be advised ahead of time, machine locating points on castings so they can give these locations particular attention as to uniformity. This is espe-

cially true when two or more patterns are made. To some this may seem to be a loss of time, but men of experience will agree that this method will save both foundry and customer time and money.

Our authors have made mention of the necessity of producing samples for the customer. These sample castings should be machined complete if possible. Otherwise a complete layout on surface plate is necessary. Inspection layout on surface plate will not tell the purchaser regarding interference in fixtures or some condition of this kind that may be very expensive to correct in the future. The surface plate check is not infallible. Machine castings in the line is to be desired if possible. This more directly effects the buyer but can also affect the foundry if any major change is necessary. It is then pointed out that the foundry after approval of samples have a definite responsibility of checking pins, bushings, patterns, and core boxes in their custody to maintain the quality represented by the samples.

(Sample castings should if possible be produced under regular production conditions that can be duplicated in the foundry

continuously.)

For best results, the buyer of castings also has the responsibility of furnishing to the foundry a periodic report on quality of castings received as they go through the machine line. There are many hidden defects you are all familiar with that can be seen only after machining. This we do at our plant.

This also brings out the necessity not only for cooperation within the foundry organization for a Quality Control Program to be successful, but the more cooperation you can receive from the customer, the better are your chances for improving your

product through Quality Control.

These reports to the foundry show the quantity of good castings machined versus number of rejected castings for each defect or cause for rejection. These reports are sent at regular intervals to the proper foundry representative for their guidance in producing better castings in the future.

I was pleased to note in the part of their thesis sub-titled "Finishing Department Control," that when their sample lot continued to show a defect it finally became necessary to check

this particular lot 100 per cent for this defect.

It is stated by many speakers that "Quality Control" inspection of anything by a sample from the lot is better than 100 per cent inspection. In many instances this is true. The monotony of the job sometimes will be the reason for rejected parts or castings going into the O. K. container. Some speakers have failed to state that if after resampling the lot one or more times and the results show the lot rejected, 100 per cent inspection by either the buyer or vendor is necessary.

The auhors have also mentioned weights as a method of checking castings. I can vouch for the fact that the individual weights on castings and maintaining a histogram type chart on these weights by part number may not only prevent your shipping the customer iron he does not want, free, when castings are purchased on a piece price, but will also warn you beforehand of possible scrap either in your foundry or in customer rejections. Control bands on the weight of castings will depend on the size of the casting and other peculiarities influencing the practical production of the casting in the foundry,

In conclusion, I wish to state that this paper is instructive, and if the steps outlined at the end of the paper are followed by any organization, it will prove its worth. If properly done, it will save more money than the expense of the organization and operation of the department. This has been the experience of several organizations with whom I have come in contact in regular business activities.

Messrs. Price and Hunsaker (Reply to Mr. Nuzum): We believe Mr. Nuzum brought one one very significant point and that is the close cooperation necessary between the customer and the foundry. When the customer gives the foundry regular reports on the number of castings machined, total scrap and causes of scrap, then the foundry has an excellent chance to correct invisible defects and dimensional discrepancies before these items can get out of control. By watching for a trend in the defects listed in these reports, large quantities of scrap castings can be avoided.

Mr. Nuzum's discussion on the subject of 100 per cent inspection of all lots found to be defective by a quality control sample

check is very good.

# MINERAL PERLITE AND ITS USE IN THE FOUNDRY

By

# E. D. Boyle and H. R. Wolfer*

Increasing emphasis on inspection and the development of new inspection tools have focused attention on shortcomings in castings. The more strenuous performance required has caused a number of failures. By far the greatest number of service failures has been in castings operating under pressure. The reasons given by foundrymen are gassy material caused by improper melting or wet sand. Sometimes impurities or improper compositions are blamed, and shrinkage caused by inadequate feeding is determined to be the trouble.

Research on gating and risering of castings indicates definite limitations upon the ability of risers to feed casting adequately unless great care is taken to position and proportion gates and risers.

It has been established that the critical temperature range of a casting is in cooling from the liquidus to the solidus temperatures (Fig. 1). All shrinkage defects in cast materials occur at this point.

Three types of insulation material have been used in this work—gypsum, diatomaceous earth in a sintered form, and mineral perlite. Gypsum presents the problems of moisture pick-up and possible sand contamination. Diatomaceous earth, in a sintered form and bonded with bentonite, worked satisfactorily but does not have the insulation qualities which mineral perlite has on non-ferrous castings.

Perlite is found in several areas in the western states. The crude gray or greenish volcanic rock called perlite is found in all western states; Arizona, New Mexico and Texas are the greatest producers. Geologists have placed the formation of this volcanic series as presumably occurring during the Tertiary, or third age, of the World's creation.

Perlite is an obsidiary from the felsitic rhyolite family. The material need not be rhyolite, but should be a siliceous or acid lava, and contain 2 to 5 per cent water still in solution in the rock. On quick heating to the softening temperature, the material puffs due to the conversion of the contained water

into bubbles of steam. Quick heating is essential to get usable expanded perlite products.

The suitable temperature range for puffing must be determined for each raw material, and will vary from 1600 to 2600 F. If heated to the lower end of the temperature range only, the material will be but slightly puffed, and therefore have low bulk density but high strength. The expansion is greater when the material is heated to higher temperatures and is greatest near the point when it tends to become molten. If the material is heated to the point where it can partly liquify, gas escapes and the granule shrinks. If the perlite is fully melted, all gas will escape and a dense slag is the result.

Proper particle sizing is important in the production of an expanded aggregate. Perlite rock is actually a type of glass which shatters easily, thus producing excessive fines. There has been no known market for these excessive fines up to the present time. Analyses of processed perlite and sintered diatomaceous earth are shown in Table 1.

# **Experimental Perlite Applications**

Risers should be covered with the powdered perlite to approximately 3-in. thickness. In non-ferrous practice, most of this material can be reclaimed and used several times—this makes for an inexpensive in-

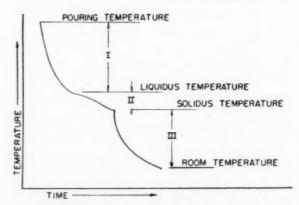


Fig. 1—Typical time vs. temperature cooling curve for a metal.

Master Molder and Quarterman, respectively, Puget Sound Naval Shipyard, Bremerton, Wash.

Note: The opinions expressed in this paper are those of the authors, and do not necessarily reflect the views of the Navy Department.

sulation. Where the ultimate in shrinkage control is desired, a collar of this material is used in the form of a core.

The insulation mixture is as follows: 20 lb grade "0" perlite; 2 lb dextrine; 2 qt asbestos shorts; 4 lb No. 107 resin; 18 per cent moisture (4 qt water).

The perlite, dextrine and asbestos should be mixed dry for 1 min; water added and mixed 1 min. Collars must be rammed firmly. Coreboxes must be coated with kerosene for parting. Collars should be vented for proper drying. Baking for 2 hr at 250 F gives best strength for handling.

Another more satisfactory collar mixture is as follows: 25 qt grade "0" perlite; 2 qt proprietary cement; 1 qt bentonite; 5 qt water. Mix dry ingredients in machine. Add water in a small amount to keep

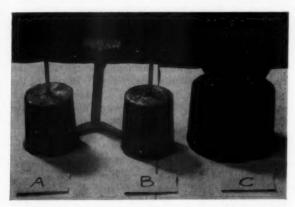


Fig. 2—Billet castings poured of navy valve bronze in the same mold and fed through the same gating system. "A" was uninsulated; "B" was molded with insulating perlite covering "C." Thermocouples were inserted in tops of billets.

material from rising due to the convection currents set up in the action. After 1 min of mixing, add the remaining water and mix for 3 min.

Collars must be rammed firmly. This mixture does not adhere to corebox. Bake for 4 hr at temperature of 350-450 F.

A wooden core box is used to ram the sleeve core for risers up to 6-in. diameter. For risers of 6-in. diameter and up, a light sheet metal form is used on the outside area. The sheet metal form remains on the core and is rammed up in the mold. The authors use metal container cans for this operation.

⁴ Thickness of insulation collars on non-ferrous risers varies. On risers of 4-in. diameter and under a 1-in. thickness is sufficient; 4 to 6-in. diameter riser, 1½-in. thickness; 6 to 8-in. diameter, 2-in. thickness; 8 to 10-in. diameter, 3-in. thickness. This formula was established by experience.

To determine the benefits of insulation, two billets 51/4 in. in diameter and 6 in. long were cast in the same mold, poured through the same gating system. Thermocouple tubes were inserted in the tops of the two billets. The billets were bottom gated and poured at 2140 F of Navy standard valve bronze (Fig. 2).

While pouring the insulated billet reached a peak of 1980 F; the uninsulated billet reached a peak of 1940 F. Over a period of 2 min the temperature of the uninsulated billet dropped to 1850 F, the insulated billet taking 4 min to reach the same temperature. The uninsulated billet then held at 1850 for 3 min, while the insulated billet held for 5 min, making the total advantage at this point for the insulated billet a matter of 4 min. At the 1830 F point the insulated billet had another advantage of 6 min.

Two operating cylinder bonnets of navy valve bronze were then molded exactly alike, except that the riser over the heaviest section formed by the juncture of the cylinder bore, gland bosses and strengthening brackets was insulated in one mold and not insulated in the other mold (Fig. 3). The riser in which the hot junction of the thermocouple was inserted was 33% in. in diameter at the top and 37% in. in diameter at the bottom, and 51/2 in. high. The uninsulated riser was the same in all dimensions.

Both castings were poured from the same ladle of metal. Pouring temperature was 2150 F. The insulated riser was covered over with powdered perlite to a depth of 1 in. Both molds were poured to the same height and, after solidification, the general level of the top of the riser which was insulated was  $\frac{3}{4}$  in. lower than that of the uninsulated riser.

Both risers reached 1900 F, a drop in temperature of 240 F at the riser areas. In 2 min the uninsulated riser dropped to 1830 F, which was considered the solidus temperature range. The insulated riser required 6 min to drop to the 1830 F, which means that the casting benefited by having a molten riser for an extra 4 min.

Aluminum manganese bronze raıı slides were cast in both 12 and 14-ft lengths. Figure 4 shows gating and risering for the two castings. The bottom gating system introduced metal into the mold without turbulence. The collar on center riser is cut away to show the amount of shrinkage which occurred in the riser. Also shown in Fig. 4 is a top view of the risering of the slide. The small risers over the 1½ in. diameter bosses did not need collars as they were covered with the bulk mineral perlite.

Figure 5 is a close-up of the center riser shown in Fig. 4. Metal was poured to the top of the collars. Metal in the risers shrank 1½ in. to the height shown. The height of the boss riser in the background remained the same as when poured, a piping action taking place in these small risers.

TABLE 1—ANALYSES OF PROCESSED PERLITE AND SINTERED DIATOMACEOUS EARTH

	Per Cent				
Component	Processed Perlite		Sintered Diatom- aceous Earth		
SiO ₂	74.22		84.9		
Al ₂ O ₂	12.52		6.1		
Fe ₂ O ₂	0.94		3.8		
CaO	0.84		3.3		
MgO	0.14		1.6		
Na ₂ O	2.68				
Sodium			Trace		
Boron			Trace		
Titanium			Trace		
Loss on Ignition	4.85				



Fig. 3—Cylinder castings poured of navy valve bronze. Center riser of casting at left had insulating perlite collar. Riser on casting at right was not insulated.

Sectioned risers taken from the slide casting are also shown in Fig. 5. The insulating collar of the riser at the left was made of diatomite and covered with powdered diatomite. Slight piping and secondary shrinkage occurred throughout the riser. The center riser collar was made of perlite, and powdered perlite was used for covering, resulting in even shrinkage and soundness for this riser. Several castings were made of this type. The riser to the right shows the ultimate saving in riser metal used. The riser height was cut one third without impairing metal control.

A manganese bronze casting (Fig. 6) was poured with a perlite knock-off core and 2 in. of powdered perlite on the top of the riser. The insulating core covered the thin sections of the casting, helping to keep the metal liquid enough to feed through to the heavy boss section.

A catapult piston head casting of aluminum-manganese brone is shown in Fig 7. Gating was designed to permit filling of the mold without turbulence and dross formation. The 10-in. diameter top riser was surrounded by an insulating collar of perlite and covered with the powdered perlite. This riser had no piping, the metal shrunk evenly.

Figure 8 shows a catapult piston cast in manganese bronze. The top riser, 12 in. in diameter, was surrounded by an insulating collar of perlite and had a powdered perlite covering. Arrow on riser indicates top of casting where rim will be machined off. The casting weighed 650 lb, and the riser 180 lb.

The mineral perlite has been used in the Puget Sound Naval Shipyard Foundry as a main ingredient of insulating collars for promoting feeding in nonferrous risers. The cement-bonded collar core material after being used once in the non-ferrous foundry is then ground up and used as insulation covering for the steel risers. The cement immediately forms a slag and holds the perlite down, forming an excellent insulation cover. The insulation materials form a glasslike slag which follows the shrinkage down. When collar material is not available, a mixture of one part cement to five parts perlite is used.

In pouring stainless steels from an open ladle, the powdered perlite is used as a covering over the molten metal. It forms a molten slag which adheres to

all slag on the ladle.

When steel is poured into a bottom-pour ladle, one shovelful of powdered perlite is placed around the stopper rod. This prevents the slag from holding the rod up due to freezing around the sleeve. A heavy slag often causes difficulty in stopper rod release.

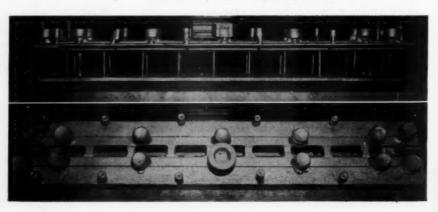
# Perlite Uses in Molding Sands

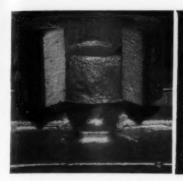
The foundry industry fully realizes the importance of sand control. Competition from many sources made foundrymen surface conscious. The sand control problems presented through research show the necessity for a cushioning material to take care of the expansion of the silica sand at the critical 1000 F temperature range. The following materials have been used as cushioning agents: iron oxide; bark flour; wood flour; rice hulls; corn cobs; and sea coal. Perlite should be added to this list. The authors have found that it excells any of these materials.

The authors developed a facing for bronze plaques, consisting by volume 50 per cent of perlite and 50 per cent Albany "0" molding sand. This sand possessed high green compressive strength when tempered to a moisture content of 10.4 per cent. Table 2 shows typical test data on this sand. The ability of this sand to produce exceptional detail on plaques and other broad, flat surfaces, especially when rammed to a high mold hardness, suggested the possibility of incorporating this toughness in a facing material for machine molding of bronze flanges ranging up to 60 lb in weight.

The naturally bonded Albany "0" sand used in the brass foundry tested as follows: Mix No. 1—sample weight, 170 grams; moisture, 7.0 per cent, green compressive strength, 5.5 psi; permeability, 14; deformation, 0.015 in. This sand mixture produced cutting and scabbing defects in the heavy Navy B-176 modified flanges when poured in green sand molds. Addition of 1 per cent by weight of western bentonite

Fig. 4—Top—Aluminum-manganese bronze rail slide castings, bottom-gated to avoid turbulence. Perlite collar on center riser cut away to show shrinkage. Bottom—Top view showing even shrinkage in risers. Small risers did not have collars. All risers had perlite top coyer.





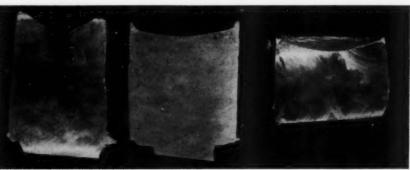


Fig. 5—Above—Close-up of center riser shown in Fig. 4.
Collar height was 35% in. Metal poured to top of collars.
Riser shrinkage was 1½ in. Right—Sectioned risers from guide rail casting (Fig. 4). Riser on left had diatomite insulating collar and cover of powdered diatomite. Center riser had perlite insulating collar and cover of powdered perlite. Riser at right had same insulation and covering as center riser except that it was poured only two thirds as high and still fed the casting.

(mix No. 2) raised the green compressive strength but did not correct the scabbing tendency.

Tempering this sand to its optimum green strength produced excessive deformation which produced straining in the mold and rough surface defects due to poor flowability and stickiness.

Use of 1 per cent by weight of bark flour in addition to the bentonite (mix No. 3) produced castings without scabs or cuts, but x-ray examination showed evidence of defects caused by gas evolution during pouring. Although the bark flour gave greater resistance to heat shock and eliminated high hot strength defects, the low permeability of the sand did not allow for ready escape of the extra gas produced by this additive. Representative test values for these mixtures appear in Table 2.

To produce a sand of high green strength, toughness and good flowability, a sand that could be used by semi-skilled labor on the molding machine unit with consistent results, the following mixture was tried: Brass Facing No. 10—250 lb Albany "0" sand (tempered to 6.0 per cent moisture); 8 lb western bentonite; 12 lb grade "0" mineral perlite.

Ingredients were mulled together 8 min and discharged from the mill through an aerating unit. No additional water was used. Final moisture reading was 6.4 per cent. It was reasoned that this higher reading was due to the change in density caused by the addition of the mineral perlite which weighs approximately 1 lb per gal.

Mixture No. 10 containing perlite can be molded with good results with moisture content ranging from 6.0 per cent to 9.0 per cent, and over a mold hardness range from 30 to 80, depending on the ferrostatic pressures encountered.

To determine resistance to spalling, standard rammed sand samples, 2 in. in diamter by 2 in. high, were inserted in a hot furnace at about 2000 F and held for 3½ min. Samples of sand No. 1, Albany "0" without additions, showed cracking at the end

of 1 min exposure and definite spalling ½ min later. After cooling to room temperature these samples could not be handled without sloughing of the outer shell.

After observing beneficial effects of the mineral perlite additions to the naturally bonded non-ferrous molding sand, it was decided to investigate its effect on a synthetic silica sand for molding steel castings. The following mix was standardized: 1000 lb Ottawa silica sand, A.F.S. No. 57; 50 lb silica flour; 45 lb bentonite; 10 lb fireclay; 9 lb dextrine gums; 12 lb perlite grade "0"; moisture 4.8 per cent.

Standard 2 x 2-in. specimens were rammed and placed in a furnace at 2000 F for 3½ min. Specimens containing perlite showed less change from heat shock than the other facings tested. Especially noticeable was the absence of shell separation on specimens made from the perlite mixture. In fact, some specimens tested in the furnace immediately after ramming (no air drying took place) showed complete absence of outer shell separation and, when broken by a sharp blow, these specimens fractured through planes cutting the central axis without loosening the outer shell (Fig. 9).

Dip tests were made with the perlite mix and the standard facings used in the steel foundry. Samples

TABLE 2-TEST PROPERTIES FOR SAND MIXTURES USED

	Sample Weight		Green Compress re, Strengtl		De- forma-
Moisture No.	grams,		psi		y in./in.
No. 1 Albany "O"					
(no additions)	170	7.0	5.5	14	0.015
No. 2 Albany "O"					
(1% western bentonite) No. 3 Albany "O" plus		7.0	7.5	10	0.016
1% bentonite plus					
1% bark flour	170	7.0	9.5	10	0.018
No. 4 Albany "O," 250 lb.					
bentonite, 8 lb	170	6.0	12.5	12	0.022
Plaque Facing	155	10.4	7.6	7	0.018
No. 10 Albany "O" plus bentonite					
plus perlite	155	6.4	11.0	12	0.018
No. 11 Albany "O"					
plus bark flour	170	6.0	7.5	10	0.012
No. 10 Albany "O" plus bentonite					
plus perlite	155	8.2	12.7	12	0.020
No. 10 Albany "O" plus bentonite					
plus perlite	155	6.2	. 11.5	11	0.022

of the perlite mix showed slightly more wetting of the surface by the metal on prolonged immersion than did the other samples. Fewer cracks were formed on the surface of the perlite specimens after

repeated immersion and withdrawal.

The steel facing containing perlite was first used on the cope half of a mold for a 2000-lb stockless anchor. This casting previously required considerable chipping to remove burned-on and penetrated sand at the junction of the flukes and the body on both cope and drag surfaces. Deep veining also occurred at this spot with the standard facing containing 10 per cent silica flour.

This casting showed the same penetrated condition on the drag face that was present on the other two anchors from the same pattern and poured from the same heat of steel. The cope face of the anchor which had been faced with the perlite mix was quite different. Most of this face peeled in the shakeout. The only sand adhering was in the area where penetration was expected, and this area was clean when the casting came out of the shotblast.

Subsequently, all remaining anchors were cast en-

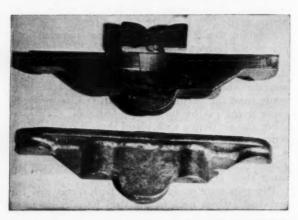


Fig. 6—Manganese bronze lifting pad casting sectioned to show (lower portion) casting contour and (upper portion) the riser and insulating perlite knock-off core in place. The riser was covered with 2 in. of powdered perlite to maintain metal fluidity and promote feeding.



Fig. 7 (left)—Aluminum-manganese bronze casting (weight, 200 lb) had 10-in. diameter top riser (weight, 130 lb) with perlite insulating collar and powdered perlite top cover.

Fig. 8 (right) — Manganese bronze casting (weight, 650 lb) had 12-in. diameter top riser (weight, 180 lb) with perlite insulating collar and powdered perlite top cover. Arrow indicates top of casting where riser will be machined off.

tirely with the perlite facing. Results were consistent from heat to heat on this and other steel castings. Examination of the sand-metal interface showed a shell formation over the entire surface of the mold where it contacted the casting. These castings all showed a smooth coating of blue oxide on the surface. This coating was thin over the lighter sections of metal, becoming heavier over the heavy sections. Variation in thickness of this oxide scale coating on the casting closely paralleled the variation in thickness observed in the shell formation of mold surfaces.

Figure 10 shows two cast steel arresting gear covers, 25 in. in diameter, weight 80 lb. Cover "A" was poured in a mold faced with standard steel sand, and the casting showed buckle and scab defects. Cover "B" was poured in mold faced with steel molding sand containing perlite to counteract expansion de-

fects on flat surface.

A number of heavy steel gate valves were cast with the perlite mix facing. These gate valves have 4-in. thick flanges and require 2400 lb of steel to pour. The sand peeled readily from these valves. The fillet areas around the heavy flanges, always subject to metal penetration and deep veining on previous casts, peeled perfectly when cast with the perlite mix. Figure 11 shows one of these valves with only a light shotblasting to determine whether metal penetration had occurred.

Examination of the shell coating, which is blueblack in color, showed it to be fairly strong but brittle; samples of greatest thickness (3% in.) being

Under the microscope at 50 diameters the sand grains at the sand-metal interface still retained their identity and orientation. They appeared to be coated with a silvery film of iron oxide scale. Random pits were observed in the otherwise smooth and continuous sand surface. The surfaces of these pits were made up of quartz grains of very jagged and distinct outline. No sign of fusing could be detected. Some

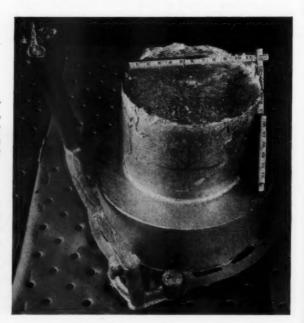




Fig. 9—Standard 2-in. rammed specimens of Albany "0" naturally bonded sand after 3½ min in furnace at 2000 F. Specimen "A" had no additions. Specimen "B" had additions of bentonite and mineral perlite.

pits contain crystalline quartz masses, white to transparent in color, contrasting with the main body of sand grains which appear to be coated with a semitransparent vitreous material.

Samples of this material were sectioned transversely, then mounted and polished for microscopic examination. Under polarized light at 50 diameters magnification the material appeared to consist of ordinary quartz grains in no way changed by the heat of the casting. No evidence of cementing was found except a few random needles of glasslike character. Infiltration of slag could not be substantiated.

Examination by electron diffraction indicated a predominance of alpha quartz. Investigation is being continued to determine whether fayalite has been formed.

Samples tested with a strong magnet showed slight attraction on the sand-metal interface, but when this was freed of all oxide scale no attraction was apparent. In no case was metallic iron found present except in an occasional isolated globule,

Silica (SiO₂) is the main component of molding sands, due to its availability and its high fusion point. The expansion characteristics of silica have been carefully studied. Silica will expand more than 1 per cent in length when heated to 1500 F. The expansion characteristics of silica can be changed only by fusing. This is obviously impractical for foundry sands. The only practical way that expansion characteristics of molding sand can be changed is to alter the composition of the material surrounding the sand grains or to change the grain size distribution of the sand.

Samples of the fused silica molding sand were removed from the steel castings to determine the percentage of metallic iron and iron combined in oxides. The results are shown in Table 3. The analysis

shows that an iron oxide fusion occurs between the casting and molding sand.

Large samples from fillet areas where veining and fissuring to a depth of 5/8 in. had been encountered with previous sand practice showed no defects. Close observation of the sand surface in those areas showed distinct lines of demarcation indicating that the forces tending to cause this type of defect did exist in this area but did not cause rupture sufficient for the metal to enter into the sand. The dark lines may be due to infiltration of oxide in these lines of fault, or to concentration of a cementing material.

Samples peeled from extremely heavy sections of a casting show a greater number of larger pits such as described previously. In all cases these are isolated from the metal of the casting by a heavy layer of iron oxide scale. It is supposed that these pits are formed after skin formation has taken place and are the results of solution of some quartz grains in a

TABLE 3-IRON CONTENT IN FUSED SAND

Sample	Iron (metallic), % by wt.	Iron (oxide), % by wt.
No. 1—Unfused sand	0.00	0.05
No. 2—Unfused sand from		
large fused piece	0.00	0.12
No. 3-Large fused piece	0.27	14.0
No. 4—Small fused pieces	0.09	9.3

chemical reaction with the iron oxide scale. This oxide, and possibly fayalite or other products, may remain longer in a molten or reactive state in those areas of maximum heat concentration, and therefore cause more transformation of the sand face even after solidification of the casting.

The use of perlite as a cushioning agent in nonferrous or Albany sand is shown in Fig. 12. Dimensions of the larger bronze plaque are  $26\frac{1}{2} \times 39 \times \frac{1}{4}$ -in. thickness. This casting was made in gun bronze poured at 2400 F. The metal was melted in a lift coil induction furnace. The green sand facing consisted of equal parts by volume of Albany sand and perlite (1 gal Albany sand "0"; 1 gal grade "0" mineral perlite; 5 oz bentonite; 1 pt water). The laboratory test results on a 160-gram sample were: moisture, 10.4 per cent; compressive strength, 8.6 psi; permeability, 10.0; toughness, 33.5.

Some foundrymen may say that it is impossible to obtain good results with such high moisture, toughness of 33.5, and such low permeability. The writers' opinion is that perlite, being globular and hollow in

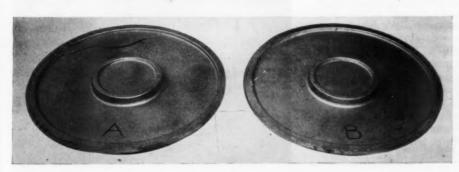


Fig. 10 — Cast steel gear covers (25-in. diameter, 80-lb weight) were poured in molds with "A" standard steel sand facing, and "B" steel sand facing containing mineral perlite to eliminate buckle and scab defects on the flat surface.



Fig. 12—A green sand facing mixture—equal parts of Albany sand and mineral perlite—was used in molding these gun bronze plaque castings.



structure, allows for the easy escape of the created gases. The facing mixture was riddled over the pattern surface for an approximate depth of 2 in., and the heap sand was placed over this with snug ramming. The pattern was withdrawn and the mold closed with no additional finish. The molding time for this plaque was 3 hr. The old method called for special facings, return of pattern to mold, and a skin dry taking 6 to 8 hr, the result then not being as satisfactory as in green sand.

A comparatively large magnesium-aluminum casting is shown in Fig. 13. The metal composition was 10 per cent magnesium and the balance aluminum. The cleaned casting weighed 400 lb. The inside of the housing casting was of a very intricate nature, having a housing cast within and heavily ribbed. In order to control directional solidification it was necessary to maintain riser fluidity for a maximum time. This was obtained by insulation, the top risers had a collar of mineral perlite, and every riser was covered with 3 in. of powdered perlite.

In making an aluminum casting with 10 per cent magnesium it is essential to add inhibitors to the molding sand. Preliminary test specimens were cast in the regular molding sand, and several with the perlite addition used as an inhibitor. The test bars made without the perlite addition had black or dark surfaces and the fractured metal showed a marked discoloration, while those cast with perlite in the sand had a bright sheen.

The following charts show results obtained by using the Navy Standard test bar cast in each sand mixture:

Time after quenching, hr	Tensile Strength, psi	Yield Strength, psi	Elongation, % in 2 in.
19	45,250	29,000	11.72
181/9	45,000	28,500	12.5
48	46,500	29,500	12.5
48	50,000	30,000	15.6
19	39,887	28,600	7.05
19	39,200	25,800	9.38
48	36,800	27.500	6.25
48	40,746	26,400	11.7
	quenching, hr  19 181/2 48 48 19 19 48 48	quenching, hr psi  19 45.250 181/2 45,000 48 46,500 48 50,000 19 39,887 19 39,200 48 36,800	quenching, hr         Strength, psi         Strength, psi           19         45,250         29,000           18½         45,000         28,500           48         46,500         29,500           48         50,000         30,000           19         39,887         28,600           19         39,200         25,800           48         36,800         27,500           48         40,746         26,400

* Cast in perlite sand mixture.

** Cast in standard molding sand mixture.

The writers feel that the cause of failure of the last four tests was due to the lack of mineral perlite in the sand.

The molding sand mixture was as follows: 300 lb fine silica sand, A.F.S. Fineness No. 69; 10 lb bentonite (western); 2 lb dextrine gums; 10 lb perlite No. "0"; 1½ lb pitch.

The molding sand tests: moisture, 3.8 per cent; green compressive strength, 5.2 psi; permeability, 105; deformation, 0.032 in./in.; air dried strength, 228 psi.

Aluminum ingot (220.T.H.) was melted in an oil-fired crucible. Dry nitrogen was introduced through the metal for a 15-min. period.

## Perlite in the Coreroom

Perlite has been incorporated in all core mixtures. The amount of perlite in the mixture was gradually increased until penetration and veining disappeared. The mixture used for heavy steel and non-ferrous castings is as follows: 300 lb old silica sand; 300 lb new silica sand, A.F.S. Fineness No. 48; 7 lb bentonite; 5 lb perlite; 2 gal linseed oil; 5.5 pct moisture.

Silica flour has geen gradually eliminated from the

Fig. 11—Drag half of steel valve cast with perlite mix facing after shakeout and light shotblasting. Cleaned weight, 920 lb.





Fig. 13—Large aluminum, 10 per cent magnesium casting in which riser fluidity was maintained by the use of mineral perlite insulating collars on the top risers, and 3-in. coverings of powdered perlite on all risers. Cleaned casting weight, 400 lb.

foregoing mixture. This mixture has shown better resistance to penetration. Also, hot strength was adequate.

Perlite in the amount of ½ per cent by weight is used in mixtures for small cores for steel and non-

ferrous castings.

Core-blowing mixtures containing ½ to 1 per cent perlite are used. Mixtures of higher green strength than ordinarily used can be blown easily with perlite in the mixture. Silica flour in hot-top and breaker-core shapes has been gradually displaced by perlite in the mixture. Downgates, channel cores and runner cups, with perlite up to 1 per cent by weight, have demonstrated better resistance to erosion when used in the steel foundry. The surface next to the steel develops a glaze of greater depth than formerly. No cutting or eroding has been found even in the most severe applications.

One pound of perlite (ground to pass the 200mesh sieve) per 100 lb of dry materials is used in compounding a mold and core wash. This wash ex-

hibits good hot plastic qualities.

#### Summary

Ways in which perlite can be used in the foundry:
1. Insulation—Collars, powder in riser covers, pads on thin sections enabling the riser to feed through.

2. Use in all molding sands.

Use in core sands (especially beneficial in the core-blowing mixtures).

4. Inhibitor for molding sand.5. Ladle covering (powdered).

Hygiene factors concerning the use of mineral perlite have been studied thoroughly. The industrial hygiene report states that:

 In contact with moisture, perlite is slightly alkaline.

b. Screen analysis showed that a considerable proportion of the material was smaller than 60 mesh, about 19 per cent being less than 200 mesh.

c. Petrographic examination with plane polarized light indicated absence of free silica. This was confirmed by chemical analysis which showed less than 1 per cent free silica.

d. Microscopic examination of the 200-mesh mater-

ial was made. These fine particles have the appearance of shattered glass, having sharp corners.

It is concluded from the foregoing and from other sources of information that perlite is primarily composed of various silicates in a fused, glasslike form.

It is considered, therefore, that perlite is a relatively non-hazardous material.

#### References

1. Bureau of Mines Information Circular 7364.

2. Report on Molding Materials, A.F.S. Transactions, 1951, pp. 101.

### **APPENDIX**

1. Messrs Savage and Taylor's paper, "Fayalite Reaction in Sand Molds," pages 564 to 577, A.F.S. Trans-ACTIONS, vol. 58 (1950), is a very thought provoking paper on fayalite reaction. Quoting Mr. J. Juppenlatz of Lebanon Steel Foundry, he recognizes the importance of the effect of a favalite formation on production castings. There is no doubt about sand mold interface breakdown. It is now shown that liquid fayalite is present after casting, and exists at temperatures above 2200 F. Steel is usually poured at 2800 to 3000 F releasing FeO, which in turn, is very reactive with silica forming fayalite, diffusing under the mold surface. At the same time, metal pressures are greatest, pushing the metal skin of the casting, the fayalite reaction has then formed a tight layer preventing metal penetration. We feel the sintered layer is formed instantaneously, otherwise your sand grains would not retain their identity and orientation. If this reaction did not take place you would have metal penetration, scabbing and buckling of the sand.

2. During the first year of our tests with perlite we were under the impression that we could control the thickness of the fayalite with our pouring temperature, but this can be done only when your facing sands are always constant and you do not have a reducing condition in your mold. Your organic binders and mold washes are a factor in retarding this fayalite condition by setting up a reducing atmosphere.

3. Mr. Juppenlatz states a more practical approach toward the improvement of steel casting qualities seems to be resolved in the use of an inert refractory wash, applied to a silica mold backing that is not reactive to FeO which is present under normal atmo-

spheric casting conditions.

4. We have been using high refractory washes, some containing large amounts of zirconite. The sintered layer is definitely retarded with the higher refractory washes. Using zirconite sand the interface reaction is also retarded. With the use of perlite you can reduce your mold cost by using less organic binders, a good cheaper mold wash, and obtain a better surface with a great reduction in your cleaning room costs due to the protective sintered layer filling the intergranular spaces adjacent to the casting.

# DISCUSSION

Chairman: H. J. WILLIAMS, New Jersey Silica Sand Co., Millville, N. J.

Co-Chairman: STANTON WALKER, National Industrial Sand Assn,, Washington, D. C.

Recorder: D. C. WILLIAMS, Ohio State University, Columbus.

H. E. Lewis (Written Discussion): 1 The authors of this paper are to be commended for their initiative in evaluating the benefits of expanded perlite as a new foundry material. One of the first major technical references on foundry uses of perlite was published by Mr. E. D. Boyle in the July 1950 issue of AMERICAN FOUNDRYMAN and aroused a great deal of interest. Unfortunately at that time, the desired perlite products were being manufactured only in several western cities. Thus much of the interest then exhibited in this new material remained at an academic level in other parts of the country. During the past two years, manufacturing facilities for expanded perlite have been increased tremendously, to the point that satisfactory perlite products are available to foundrymen in the United States, Canada and several foreign countries. This means simply that perlite can now be obtained in quantity for full-scale plant applications in nearly every foundry market.

This second article by Mr. Boyle confirms the work reported earlier and also presents a greatly broadened scope of investigation. It is increasingly apparent that specific forms of expanded perlite offer foundrymen new and useful tools with which to reach their common objectives, namely, the production of higher quality castings at lower net cost. Mr. Boyle's list of case histories is quite complete on different types of foundry applications of perlite. It would be helpful of course for analytical purposes if the improvements he attained were expressed in

terms of dollars served per job or per year.

The specific perlite particle size preferred as a cushioning agent in molding and core sand is identified in this paper as "Type 'O.'" Inasmuch as this nomenclature is not standard within the perlite industry, it seems preferable to use an A.F.S. Fineness Number to more accurately define this size. For example, our company's cushioning agent and riser insulation has an average A.F.S. Fineness No. of 110 to 115. Similarly, a much finer particle size soon will be introduced for core and mold washes. Since perlite foundry products are manufactured to meet density (or weight) limits in addition to the particle size requirements, the A.F.S. Fineness Number does not provide a complete specification—but it helps.

Light weight perlite products are packaged in multi-ply paper bags, usually containing 4 cu ft of material rather than a specific number of pounds. Accordingly, it often is desirable to proportion the use on a volumetric basis. Considering the prevailing bulk density of perlite cushioning agents, it can be determined that 1 per cent perlite by weight of sand is essentially the same as 12 per cent by volume. Two per cent by weight of sand is about 25 per cent by volume; 4 per cent by weight (used in some facings) is 50 per cent by volume.

Delicate measurements of the *increment* of perlite employed as a cushioning agent have not been found necessary in most cases. Users have employed 1,  $1\frac{1}{2}$ , 2 per cent, etc., by weight of the sand, depending on the type of sand, binder and other ingredients in the mixture. These proportions can be simplified on a volumetric basis without serious error as follows:

CONVERSION TABLE FOR PERLITE CUSHIONING AGENTS

% By Weight of Sand	% By Volume of Sand	Cubic Feet Perlite per Ton of Sand	Bags of Perlite per ton of Sand (4 cu ft each)
1/2	6	2	1/2
1	12	4	1
11/2	18	6	11/2
2	25	8	2
3	37	12	3
4	50	16	4

It would be foolhardy to state that perlite is the solution for all foundry sand and riser feeding problems. Obviously the many variables involved in different metals, sands, binders, mixing, molding, washing, etc., preclude the possibility of forecasting or warranting specific results. But the remarkable degree of past successful achievement, as measured both in research and in practical plant usage, leaves little doubt but that the unique properties of perlite can be incorporated in controlled foundry operations to ease some of the foundryman's current problems.

It is hoped that we may look forward to additional articles by the authors translating into monetary savings the successful

perlite foundry uses listed in their summary.

Member: Is there a build-up of perlite in the sand mixture?

Mr. Boyle: When using Albany sand there is a possibility of build-up but a small addition of bentonite will take care of that. When using compounded sand mixtures the fines from perlite are removed at the shake-out.

perlite are removed at the shake-out.

H. F. TAYLOR: "When producing non-ferrous castings we have made studies comparing perlite and gypsum and have concluded that gypsum is superior to perlite. I would like to ask Mr. Boyle to state the economic advantages of perlite.

It is possible that perlite could be used as a cushion in sand mixtures because of its inert nature.

MR. BOYLE: I believe that perlite would be an excellent cushioning material in sand mixtures due to its inert property and in this respect might serve better than wood flour.

When using gypsum sleeves I believe that more castings are

lost than saved.

Sleeves made with perlite can be stored for long durations without fear of moisture absorption resulting in loss of strength. Sleeves made with perlite do not require the delicate handling necessary when using gypsum sleeves.

K. A. Miericke: ⁸ The new permeable type of gypsum will

K. A. MIERICKE: ³ The new permeable type of gypsum will not contaminate the sand mixture. This new material produces a cellular insulating structure with a permeability value of 90.

MR. Lewis: We have heard the respective merits of insulating pads and sleeves which are produced by using either perlite or gypsum. It may be that such pads and sleeves produced from mixtures containing both gypsum and perlite might produce items with qualities superior to those obtained when using each material alone.

¹ Great Lakes Carbon Corp., New York.

Massachusetts Institute of Technology, Cambridge, Mass.
 Baroid Sales Div., National Lead, Chicago.

# HOT-TEAR FORMATIONS IN STEEL CASTINGS

By

U. K. Bhattacharya,* C. M. Adams* and H. F. Taylor**

#### ABSTRACT

Tearing was produced in a flanged casting by placing restraining bars between the flanges. The degree of restraint was controlled by means of a gap in the bar. Apparatus was devised to detect the time and temperature at which cracking initiates in the casting.

In mild steel killed with 0.10 per cent aluminum, the tearing temperatures were between 2475 F and 2510 F. The linear construction measured 0.91 to slightly less than 0.60 per cent respectively. The steel became sufficiently strong below 2475 F to resist tearing by the contraction stresses. Silicon deoxidized steel castings failed at lower temperatures, indicating a greater susceptibility to hot tearing. An addition of 0.05 per cent aluminum was required in the case of high sulphur steel (S =0.05 per cent) to recover hot-tear resistance.

Limited investigations conducted on medium carbon steel (0.49 per cent C) indicated that steel of this composition would not tear below 2400 F, under any of the conditions imposed by experiment.

### Introduction

The phenomenon of hot tearing has troubled foundrymen for years. It is not only peculiar to steel castings, but is also known to occur in cast magnesium and aluminum alloys. It is difficult to prevent systematically its occurrence in a complex casting because so little is known about solidification mechanism in a mold.

Molten steel in cooling to room temperature undergoes the following volume changes—(1) liquid contraction, (2) solidification shrinkage, and (3) solid contraction. When steel is poured into a mold, a thin skin forms as the molten metal comes in contact with the sides of the mold. Its thickness progressively increases as freezing proceeds toward the thermal center of the casting. During this process, the volume changes mentioned above take place simultaneously. The decrease in volume due to the loss of superheat and solidification is compensated by risers whose function is to provide feed metal. The solid contraction is taken care of by making the casting slightly larger than the original dimensions by using the patternmaker's shrinkage rule.

When the solid contraction cannot take place free-

ly, the casting is subjected to stresses. When these stresses exceed the tensile strength of steel, the surface of casting is ruptured; this is due to lack of ductility at high temperature. The tearing generally takes place in the hottest (and weakest) part of the casting and then the tear becomes a source of stress concentration. In the absence of large thermal gradients, the entire casting cools evenly, and passes the critical temperature before the contraction stresses become large enough to exceed the strength of the steel. Below the critical temperature, the casting will deform plastically instead of tearing.

A hot-tear probably progresses from the surface into the interior of the casting. It has a ragged appearance and is usually visible to the naked eye. However, minor cracks may escape detection visually as well as on the radiograph, in which case they can be readily located by the magnetic powder or penetrant method of testing.

In cast steel, hot-tears should be distinguished from internal tears which are in reality shrinkage cavities. These originate from the vicinity of a pipe or shrink region in the interior of the casting. These tears seldom reach the surface of the casting. Radiography is the only reliable method of locating them; unlike hot-tears, internal tears show a ray-like appearance. These are eliminated by improved feeding technique.

Hot-tears occur as a result of stresses arising from (1) casting design; normal contraction of thinner parts being hindered by thicker sections which cool more slowly. A gear casting is a typical example in which tearing may occur at the junctions of the spoke with the rim and hub. (2) Bulk resistance of mold and core prevents contraction of a casting. In a complex-shaped casting, tearing may occur as a result of the combined effect of the above mentioned causes.

The method adopted to prevent a hot-tear depends upon its origin. The preventive measure in the gear casting would consist of suppressing the hot-spot by the judicious use of external and/or internal chills. The mold resistance is decreased by increasing the collapsibility of the mold and core sands.

#### Review of Literature

There is a large amount of literature pertaining to hot tearing of steel. The important works are sum-

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marized below.

Körber and Schitzkowski1 studied tearing in steel castings caused by mold hindrance using acid and basic openhearth steel, with carbon varying from 0.15 to 0.39 per cent. The test bars, 17.72 in. long and of different diameters, were made in hard dry sand molds. Cracks occurred when the recorded contraction was between 0.34 and 1.78 per cent, depending on the size of the bars. The contraction of 1.181-in. diameter bars was then restrained at different stages of cooling by placing steel rods of different lengths between the flanges. The bars broke when the contraction was 1.68 per cent or less. The autographically recorded time-contraction curves of bars poured in the hard molds were superimposed upon those of bars made in soft green sand molds. The curves diverged at about 2372 F, indicating fracture. majority of the castings broke at the junction of the flange and bar, but temperatures were measured at the center of the bar. Therefore, the recorded temperatures probably did not represent true tearing temperatures. This work did not indicate the effect of composition.

In an effort to "continue" the work of Körber and Schitzkowski, Briggs and Gezelius² made studies on the solidification and contraction of cast steel. Tensile strength and ductility were determined at temperatures around 2372 F by pulling cast tensile bars with a machine fitted to the mold. Their limited data led them to state that the tearing temperatures determined by Körber "are nearly correct."

Briggs and Gezelius^{3,4} also measured the free and hindered contraction of cast carbon and alloy steels, and the stresses which were set up in the casting. The experimental casting was 30 in. long, with a diameter of 2½ in. at each end, tapered to a diameter of 1½ in. at the center where it was gated. The contraction was restrained by means of stainless bolts cast into

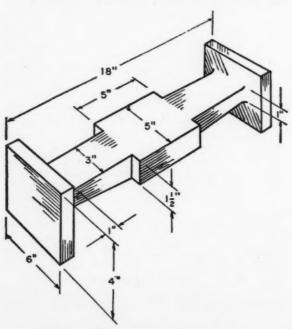


Fig. 1-Sketch of experimental casting.

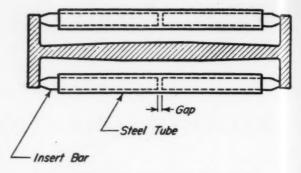


Fig. 2-Arrangement of insert bars in mold.

each end of the bar; one of these was tied to the flask and the other was connected to a calibrated spring. Under a strong spring, the total hindered contraction in casting to room temperature amounted to about 0.50 per cent. The bars did not break since there was no sharp temperature gradient in the casting. Thus, this work could not give a measure of the stress required for tearing. However, the authors plotted contraction or stress-temperature curves and showed that above 2372 F low carbon steel contracted more and therefore developed higher stresses than high carbon steels. Since low carbon steel has greater strength and ductility, this work did not establish which steel would be more susceptible to tearing.

Piwowarsky and others⁵ made determinations of the tensile strength and ductility of cast steel at high temperature (previously cooled to room temperature). They found that steel had no ductility at or above 2282 F.

Hall⁶ showed that cast steel, with the exception of very low carbon steel, had no ductility between the solidus and 2372 F. It was found that high sulphur content lowers the temperature at which steel acquires ductility. In subsequent experiments, the formation of hot-tears in flanged castings was studied. However, tearing temperatures were not determined.

Andrew and Protheroe⁷ reported that hot tearing in steel castings was due to mechanical resistance of sand rather than friction between contracting steel and sand grains. (They attempted unsuccessfully to determine the tearing temperatures by using an electrical circuit through the casting.) However, they concluded that steel castings "break" very near the solidus.

Phillips⁸ studied tearing in a ring casting using hard sand cores. No effect of casting temperatures on tearing was found. The last steel poured from the ladle showed the most susceptibility to tearing, and this was related to a change in the type of inclusions found in the casting. This change in the nature of inclusions was due to a reduction of the aluminum content by the oxidizing action of slag and atmosphere. An increase in the aluminum addition from 0.10 to 0.15 per cent left sufficient aluminum in the molten metal to avoid the formation of adverse type of inclusions.

In their limited investigations, Crafts and co-workers⁹ observed that in cast medium carbon steel, *sili*cate and eutectic type inclusions were more conducive to tearing than alumina and peritectic types. Increasing aluminum additions caused a progressive change in the type of inclusions. The predominantly eutectic type predominated when the steel was killed with 0.025 to 0.05 per cent aluminum. The alumina type appeared when aluminum additions were increased to 0.10 per cent. Steel treated with an addition of 0.10 to 0.20 per cent aluminum, titanium, or zirconium and calcium-bearing alloys showed peritectic type of inclusions.

Middleton and Protheroe¹⁰ recently published the most comprehensive work to date on tearing of steel. They measured the tearing temperatures and breaking loads, and studied the effects of such factors as casting temperatures, compositions, alloying elements, and deoxidation practice on tearing. The test bar was 24 in. long, with a uniform diameter of 1.128 in. They restrained the bars and recorded the loads by means of the technique developed by Briggs and Gezelius. Some of their findings are as follows:

1. Effect of casting temperatures.

Higher casting temperatures increased the susceptibility of steel to tearing and raised the tearing temperatures.

2. Effect of composition.

a. Carbon

d

d

Mild steel (0.21 per cent C) cracked in the temperature range 2532 F to 2593 F, and medium carbon steel (0.31 per cent C) in the range 2489 F to 2570 F, depending upon casting temperatures. No data were available about steel containing 0.49 per cent carbon, because when it cracked liquid metal still remained in the interior of the casting due to the wider freezing range, and partly rewelded the cracks. As a result of this, minor cracks formed which the apparatus did not detect. At a given casting temperature, medium carbon steel was more susceptible to tearing than mild steel and failed under a smaller load.

Subsequent experiments were made on 0.21 per cent carbon steel.

b. Silicon and manganese

0.50 per cent silicon content showed better tear resisting properties than 0.35 or 0.90 per cent silicon.

An increase of manganese from 0.85 to 1.21 per cent decreased susceptibility to tearing. Further increase to 1.71 per cent had a deleterious effect.

c. Sulphur

Increasing sulphur content increased susceptibility to tearing.

3. Effect of deoxidation with aluminum.

The susceptibility of steel to tearing depends upon its sulphur content and the degree of deoxidation. Aluminum additions are generally beneficial; however, small additions of aluminum (0.025 to 0.051 per cent) to silicon deoxidized, high sulphur steel lower resistance to tearing. According to Sims, 11 when sufficient aluminum is added to combine with all of the oxygen, the surface tension of low melting point sulphides decreases. As a result, these sulphides spread out as continuous chains or films along the primary grain boundaries. Excess aluminum seems to increase surface tension of the intergranular liquid, and the sulphides appear as randomly distributed and globu-

lar-shaped inclusions at grain boundaries. This change in the distribution of inclusions increases the ductility of steel at high temperatures.

Middleton and Protheroe observed that the steel bars failed at higher temperatures when contraction was hindered by stronger springs. This indicates that tearing can occur in mild steel at any temperature between 2532 F and the solidus. Their experimental arrangement was such that the bar was being restrained the moment solidification began. This closely corresponds to the behavior of a casting in which a thin section is prevented from contracting by a thick section.

# **Experimental Procedure**

Apparatus—The design of the test casting shown in Fig. 1 was suggested by the paper of Bishop and Taylor. ¹² Certain modifications were made for convenience, and two blind risers were used to provide perfect feeding. Rigid separating bars, separated by a controllable gap, were molded between the flanges of the casting to induce hot tearing. Bars of different lengths were used to control the amount of free contraction which could take place before the flanges of the castings impinged upon the restraining bars. In other words, restraint was delayed for different time intervals after casting, and the amount of total solid contraction was directly controlled. The rectangular cross-section of the part which failed measured 1½ by 5 in.

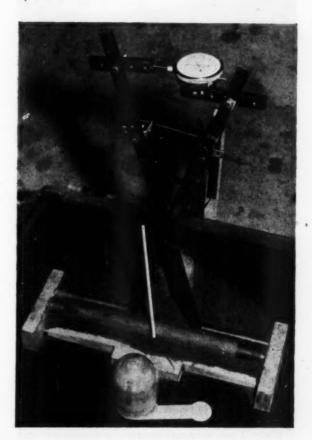


Fig. 3-Position of apparatus in the mold.

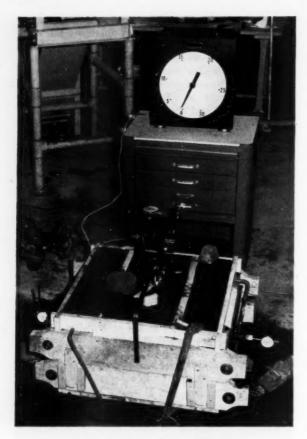


Fig. 4-Illustration of assembled mold.

The apparatus shown in Fig. 3 was devised to detect the initiation of casting failure. It consists of two steel bars crossed to form a scissors; the bottom of the scissors is linked mechanically to the flanges of the test casting, and the top is fitted with an Ames dial. Any change in the distance between flanges (positive or negative) is directly reproduced on the Ames dial. The lever arm magnification ratio is 1.5 to 1. (The spring shown in Fig. 3 does not influence casting restraint; its purpose is to insure mechanical contact between the scissors and the casting.)

The temperature was measured ½ in, beneath the surface on the section which failed. A platinum/platinum-rhodium thermocouple with a bare, hot junction was used.

# Melting Procedure

The heats were made in a high frequency induction furnace, using 75-lb charges. The castings were made from steel of the following analysis: 0.20-0.24 per cent C; 0.45-0.55 per cent Mn; 0.30-0.40 per cent Si; 0.02-0.025 per cent S; 0.02 per cent P max. No slag was used. The bath was deoxidized with ferromanganese and ferrosilicon at about 2830 F. When aluminum was used, in addition to silicon, it was added immediately before tapping. The tapping temperature was about 2910 F.

When it was necessary to increase the carbon and sulphur content of the charge, granulated carbon and iron sulphide powder, respectively, were included with the charge.

Bath temperature measurements were made with an optical pyrometer.

#### Observations

After the casting was poured, the indicator point on the Ames dial, initially at zero, showed a steady negative deflection, indicating that the casting was contracting. When the flanges encountered the restraining bar, one of two things happened:

1. The indicator stopped, remained stationary for a short time, and then abruptly moved in the oppo-

site direction, indicating an expansion.

It was experimentally verified that the sudden expansion marked the beginning of rupture. A test casting was made with an arrangement to release the restraining bars the instant an expansion was observed. Immediately the mold was broken and minute cracks were found on the surface of the casting. This observation was also made using the test casting designed by Middleton and Protheroe. The fact that measured expansion was always accompanied by cracking indicates that at the temperature of hot tearing, steel has no ductility.

2. The deflection rate changed abruptly, became very slow, but did not entirely cease. The casting was strong enough to continue contracting by bending the bar elastically, and failure never took place.

All test castings poured fell into one of the two categories outlined above. Those castings which failed did so at the geometric center of the system (the hot spot) as shown in Fig. 5. Several series of castings were poured to study the influence of composition and deoxidation practice. The size of the "gap" in the restraining bar was systematically varied. If the gap in the restraining bar were large, the casting could contract enough to remain sound. If the gap were small, tearing was produced. By adjusting the gap, that degree of restraint was obtained which would just barely crack the casting. The cracking temperature of this critical casting approaches the temperature below which hot tearing will not occur.

### Results

The results of tests made upon mild steel castings

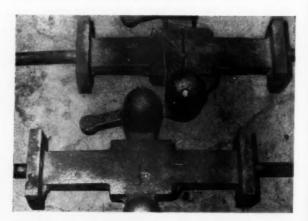


Fig. 5-Illustration of cracked castings.

containing less than 0.04 per cent sulphur, are set forth in Table 1. The castings did not tear unless the total solid contraction was suppressed to less than 1.0 per cent. The tearing temperature increased with reduced contractions, or in other words with more restraint.

Table 1—Hot Tearing in Mild Steel (S not more than 0.04%)

Aluminum Additions,	0	Width of Gap Between Restraining Bars, in.	Per Cent Contraction to Room Temp.,	Tearing Temp.,
0.10	2867	1/16		2510
0.10		0.07	0.60	2500
0.10	2849	3/32	0.91	2475
0.10	2840	4/32	1.15	no tear*
0.10	2867	5/32	1.34	no tear
0.10		8/32	1.90	no tear
* Slight	expansion	was noted at 2	430 F.	

Table 2 shows the influence of deoxidation practice on tearing resistance of low sulphur mild steel. Silicon deoxidized steel was clearly shown to be the most susceptible to tearing. Addition of aluminum increased the hot strength enough to prevent failure.

TABLE 2—RESULTS OF DEOXIDATION PRACTICE IN MILD STEEL

Aluminum Additions,	Casting Temp.,	Width of Gap Between Restraining Bars, in.	Tearing Temp., F	Remarks
			Below	
0	2759	1/8	2450	Slight tear
0	2822	1/8	2400	Gassy metal.
		76		1% C; 0.38% Mn; .25% Si; 0.04% S
0.02	2822	1/8	no tear	70 01, 010170 0
0.05	2786	1/8	no tear	

The data recorded in Table 3 indicate that when the sulphur content exceeds 0.05 per cent, a minimum addition of 0.05 per cent aluminum is necessary to prevent the casting from tearing.

TABLE 3—RESULTS OF DEOXIDATION PRACTICE IN MILD STEEL

Aluminum Additions,	Casting Temp.,	Width of Gap Between Restraining	Tearin	٠,		
%	F	Bars, in.	F	Remarks		
0	2867	4/32	2450	Linear contraction 1.0 per cent.		
0	2795	5/32	no tea	r		
0.02	2786	4/32		0.19% C; 0.44% Mn; 0.25% Si; 0.053% S		
0.02	2840	5/32	no tea	r		
0.05	2813	4/32	no tea	r		

A limited investigation of medium carbon steel indicated that tearing will not take place below 2400 F. The results are shown in Table 4.

TABLE 4—HOT TEARING IN MEDIUM CARBON STEEL

Aluminum Additions,	Casting Temp., F	Width of Gap Between Restraining Bars, in.	Tearing Temp., F	Remarks
0.10	2777	3/32	2400	0.49% C S<0.04
0.10	2759	1/8	no tear	

Middleton and Protheroe, using the test bar described earlier, have shown that castings containing 0.21 per cent carbon and 0.035 per cent sulphur, and killed with 0.075 per cent aluminum, tore at a temperature of 2532 F. This is a much higher temperature than found at M.I.T. To determine whether or not the British test bar will tear at lower temperatures under the condition of the present investigation, several heats were cast; pouring temperature determined by optical pyrometer was 2858 F. The results are shown in Table 5.

Table 5—Hot Tearing Experiments with Test Bar of Middleton and Protheroe

Remarks	No. of Sec. Elapsed Before the Casting Started to Tear	straining	Aluminum Additions, %	Approx. Casting Temp.
Temp. was neasured with an optical pyrometer.		0.03	0.075	2858 F
.,	100	0.04	20	**
	105	0.05	99	,,
	110	0.05	99	**
	no tear	0.06	**	**
Silicon deoxi- dized steel.	140	0.05	0	**

Unfortunately, no satisfactory correlation of cooling curves could be obtained, although temperatures were measured by putting bare platinum-rhodium thermocouples at a depth of 1/4 in. in the castings. Middleton and Protheroe do not report details of temperature measurement. The above results will be compared with those of the British workers on the basis of the time elapsed before initiation of cracking.

They have found that tearing occurred 65 to 80 sec after the bars were cast, and when poured at 2868 F, the bars cracked at 2548 F. Table 5 shows that the M.I.T. test castings cracked at 90 to 110 sec after the casting was poured. These data leave little doubt that tearing took place at temperatures lower than 2532 F and probably within the range established in the present investigation.

From the table, it is shown that when the steel was deoxidized only with silicon, tearing took place as late as 140 sec after the mold was cast. Therefore, this type of steel will crack at a temperature well below the tearing range reported by Middleton and Protheroe.

Experience, and current experimentation, lead the present authors to believe any measurement of "hot tearing temperature," as a physical property of metal, is meaningless and impossible. Clearly, any casting will tear at any temperature if the stress becomes

great enough. The "temperature below which a casting will not tear under normal conditions" seems a more appropriate and meaningful measurement.

# Conclusions

There is a critical temperature below which cast steel attains sufficient tensile strength and ductility to withstand contraction stresses and thereby resists tearing. This investigation has shown the critical temperature lies below 2475 F for the case of aluminum-killed, low sulphur mild steel and below 2400 F in medium carbon steel; above these temperatures, tearing occurs.

Silicon deoxidation lowers these critical temperatures and also increases the susceptibility of steel to tearing. In low sulphur steel, addition of 0.02 per cent aluminum to silicon deoxidized steel nullifies the deleterious effect of silicon. When the sulphur exceeds 0.05 per cent, the addition of aluminum should be increased to 0.05 per cent.

# Acknowledgment

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Bibliography

F. F. Körber and G. Schitzkowski, "Determination of the Contraction of Cast Steel," Stahl und Eisen, p. 128, 1928.

2. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction and their Relation to the Formation of Hot-Tears in Steel Castings," Transactions, A.F.A., vol. 41, p. 385

3. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings II-Free and Hindered Contraction of Cast Carbon Steel," Transactions, A.F.A., vol. 42, p. 449 (1934).

4. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings-Free and Hindered Contraction of Alloy Cast Steels," Transactions, A.F.A., vol. 44, p. 1

5. E. Piwowarski, B. Bozic, and E. Söhnchen, "Tensile Strength and Elongation of Cast Steel between 650 and 1450°C, Archiv. für das Eisenhüttenwesen, vol. 7, p. 127, 1933-1934.

6. H. Hall, "The Strength and Ductility of Cast Steel During Cooling From the Liquid State in Sand Molds," Part I, J. Iron & Steel Inst., special Report No. 15, p. 65, 1936; Part II-J. Iron & Steel Inst., special Report No. 23, p. 73, 1938.

7. J. H. Andrew and H. T. Protheroe, "Investigation of the Influence of Mold Friction on Tearing in Castings," J. Iron & Steel Inst., p. 101, 1942.

8. W. Phillips, "Prevents Hot-Tears in Steel Castings,"

Foundry, vol. 68, no. 7, p. 27, July 1940. 9. W. Crafts, J. Egan and W. Forgeng, "Formation of In-

clusions in Steel Castings," Trans., A.I.M.E., vol. 140 p. 233.

10, F. M. Middleton and H. T. Protheroe, "Hot Tearing of Steel," J. Iron and Steel Inst., vol. 168, part 4, p. 384, August 1951.

11. C. E. Sims, "Things We Know and Don't Know About Cast Steel," Foundry, vol. 79, no. 9, p. 92, Sept. 1951.

12. H. F. Bishop and H. F. Taylor, "Internal and External Hot-Tears," Foundry, vol. 74, p. 82, July 1946.
13. K. Singer, "Hot-Tear in Steel Castings," Giesserei, vol. 15,

p. 1225, 1928.

14. K. Singer and H. Bennek, "A Description of Hot Cracking in Steel Castings," Stahl and Eisen, p. 61, 1931.

15. First Report of Steel Castings Research Committee, Iron & Steel Inst., Special Report No. 3, p. 10, 1933.

16. C. H. Cain, "Notes on Some Common Defects in Steel Castings," Proceedings, Inst. British Foundrymen, vol. 26, p. 501, 1932-33.

17. J. Vero, "Hot-Shortness of Aluminum Alloys," Metal Industry, p. 431, April 10, 1936; Metal Industry, p. 491, April 24,

18. C. W. Briggs and R. A. Gezelius, "Studies on Solidification and Contraction in Steel Castings-Contraction Stresses," Transactions, A.F.A., vol. 45, p. 61, 1937.

19. E. Liddiard and W. A. Baker, "Cause and Control of Magnesium Alloy Microporosity," Transactions, A.F.A., vol. 53, p. 54 (1945).

20. D. C. G. Lees, "Hot Tearing Tendencies of Aluminum Casting Alloys," J. Inst. Metals, vol. 13, p. 343, 1946.

21. N. P. Gelperin, "Steel Susceptibility to Hot-Tear Formation," Transactions, A.F.A., vol. 54, p. 724 (1946).

22. F. M. Middleton, "The Hot Tearing of Steel-A Review," Iron and Steel, vol. 22, p. 407, 1949.

23. C. W. Briggs, The Metallurgy of Steel Castings, McGraw-Hill Book Co., 1946.

# FUNDAMENTALS OF FOUNDRY REFRACTORIES

By

# J. D. Custer* and J. Spotts McDowell*

The noteworthy progress of recent years in the metallurgy, in operating techniques, and all phases of production in foundry practice has exerted an important influence on the selection and development of foundry refractories. Brief mention of several examples of advances in foundry operations which have been dependent on equal progress in the commercial manufacture of refractories, will serve to illustrate the point.

Basic cupola operation, which is still in its infancy, necessarily involves considerable experimentation with refractories. Until a few years ago, basic refractories were not used in the conventional cupola, and the best technique for using them in the basic cupola remains to be fully developed. In the malleable iron foundry, operating conditions in many furnaces are now so severe that it has become necessary to use super-duty fireclay and high-alumina refractories. In the operation of electric steel-melting furnaces, much higher temperatures are now employed for the production of the larger tonnages melted in a furnace of given size as well as for the production of new alloys. In the roofs of these furnaces, high- alumina refractories of various classes, super-duty fireclay brick, and even basic refractories have been used economically. As to basic refractories, no longer can magnesite and chrome brick alone fulfill all the requirements of industry. More than a dozen different kinds of basic refractories are now manufactured, each of which is especially well-suited for a definite combination of operating conditions. A classification of refractories is shown in Table 1.

# Intermediate and Low-Duty Fireclay Brick

In certain portions of many industrial furnaces, the refractory structure is subjected to temperatures which are relatively moderate or low. For such conditions, fireclay brick of the intermediate or the low duty class may give excellent service. In the low temperature range, these brick are sometimes stronger and more resistant to abrasion than high-duty fireclay brick.

Low-duty fireclay brick have a minimum P.C.E. of cone 19, corresponding to a softening temperature of 2768 F while intermediate duty fireclay brick have a P.C.E. of cone 29, corresponding to a softening temperature of 2984 F.

Intermediate and low-duty brick in general are dense, mechanically strong, tough and resistant to abrasion and to slag action at moderate temperatures. For service in which there are rapidly fluctuating temperatures, power pressed brick are to be preferred to those made by the extrusion process. Intermediate and low-duty brick show little permanent change in volume when heated within the temperature range for which they are suited.

# High-Duty Fireclay Brick

The high-duty class of fireclay brick have a P.C.E. of cone 31 to cone 32½, corresponding to a softening temperature of 3056 to 3131 F.

In the manufacture of machine-made brick of the extruded type, the ground batch is moistened, pugged

TABLE 1—CLASSIFICATION OF REFRACTORIES

Basic Brick
*Magnesite Brick — Fired
<90 pct MgO (periclase brick)
*Chrome Brick — Fired
*Magnesite-Chrome Brick† chemically bonded
•
*Chrome-Magnesite Brick† high fired
chemically bonded
*Metal Encased Brick
chrome-magnesite
*Forsterite Brick — Fired Spinel-Bonded Magnesite-Chrome Brick — Fired

# Silica Brick

super-duty conventional

Insulating Refractories

*The ceramic bond consists of magnesium and calcium silicates. The spinel-bonded refractory is a relatively recent development.

†The terms magnesite-chrome and chrome-magnesite refer to blends of magnesite and chrome ore. The constituent predominating is named first.

^{*}Harbison-Walker Refractories Co., Pittsburgh.

and extruded through a die in the form of a dense column. This column is cut by wires into brick of the required size. The brick are then repressed, dried and fired. They have great physical strength, high density and a porosity within the range of about 10 to 18 per cent.

In the upper section of cupola linings, where the refractories are subjected to mechanical abrasion and to the impact of the charge, the high strength of the extruded blocks provides good resistance to wear, and

in some cases these give best service.

In the melting zone, abrasion may also be a factor, but usually resistance to corrosion and erosion by the slag and metal is of greater importance. High-duty fireclay brick of the extruded or power pressed types,

are usually adequate for this service.

In the manufacture of brick by the power press process, the moistened batch is fed to a mechanically operated press in which the brick are formed under high pressure. Power pressed cupola blocks have a dense uniform structure, as shown by their porosity of approximately 18 to 22 per cent. They also have excellent resistance to thermal spalling. When necessary to operate the cupola on short heats, the refractory lining is subjected to severe thermal spalling conditions. Under these conditions longer service may be obtained from power pressed cupola blocks, than from extruded blocks.

Only a good quality of ground fireclay should be used for laying the cupola blocks or for repairing the lining. No additions of Portland cement, lime or water-glass should be made to the fireclay, since such additions greatly reduce its refractoriness.

# Super-Duty Fireclay Refractories

The properties of super-duty fireclay refractories which are of prime importance are their high refractoriness, low porosity, and excellent volume stability in the high temperature range. By way of illustration, high-duty fireclay brick with a P.C.E. of cone 31 to 321/2 have a softening temperature of 3056 to 3131 F. Super-duty fireclay brick have a minimum P.C.E. of cone 33, equivalent to a softening temperature of 3173 F. In service where the refractory is used at temperatures near its ultimate working temperature, such an advantage in refractoriness is exceedingly advantageous. While the spalling resistance of superduty fireclay brick of the regular classes is superior to that of high duty brick, in some modifications of the former, other properties are enhanced to great advantage at the expense of spalling tendency. For certain particular applications rapid temperature changes which cause spalling are not involved.

Super-duty fireclay refractories are available in several burns, each of which is especially well-suited for certain conditions. The porosity of some brands of super-duty fireclay brick of conventional burn is as low as 13 to 16 per cent, as compared to 14 to 22 per cent for high-duty fireclay brick. The excellent volume stability of super-duty fireclay brick at high temperatures becomes important as operating temperatures approach the point at which appreciable shrinkage might take place in high-duty fireclay brick. For this

reason super-duty fireclay brick are used for the bottoms and lower sidewalls of air furnaces. Their use has proven economical also in electric furnace roofs where severe thermal spalling conditions may prevail. Because of their high density and low porosity, superduty fireclay brick are highly resistant to slag penetration.

Super-duty fireclay brick of extra hard burn have a porosity within the range of 12 to 15 per cent. The extra hard burn results in even further improvement in the volume stability and load bearing properties at high temperatures. In addition, these refractories have low porosity, permeability and excellent resistance to penetration by molten metals or slags. These properties account for the excellent service that has been obtained in air furnaces with tap out blocks of super-duty brick of the extra hard burn. In ladles where desulphurizing is accomplished, super-duty brick provide the maximum resistance to the fluxing action of the soda ash. They also serve to advantage for the slag knife in the front slagging trough of cupolas.

## High-Alumina Refractories

High-alumina refractories are manufactured mainly from diaspore clays. They are fired at high temperatures and contain the highly refractory minerals mullite and corundum, in proportions which depend upon the alumina content.

The excellent refractoriness of 60 per cent alumina brick is shown by its P.C.E. of cone 36 to 37, equivalent to a softening temperature of 3250 F. The 60 per cent alumina refractories have good volume stability and when subjected to the 1600 C (2910 F) reheat test, in which that temperature is held for 5 hr, they show practically no linear shrinkage. Another important property of this type of refractory is its excellent resistance to the spalling effect of rapid temperature changes.

Refractories in the burner ends of air furnaces are subjected to extremely severe service conditions. While the bottoms of these furnaces are usually built of super-duty fireclay brick, some plants use 60 per cent alumina brick at the burner end and super-duty fireclay brick for the remainder of the bottom. This practice results in the proper balance in the service life and has proven economical,

The 90 and 99 per cent classes of high-alumina refractories provide a greater margin of refractoriness, volume stability and load-bearing properties at high temperatures, than is obtainable from any other alum-

ina-silica material.

In some high-temperature electric furnaces the operating conditions are such that silica roofs may give relatively short service. The use of 60 or 90 per cent alumina brick in the center section of the roofs, with silica brick in the annular rings, may balance out the service life and prove economical. For the slag knife in troughs of cupolas, the 90 and 99 per cent alumina brick have given outstanding service. They have also been used successfully for the bottoms of aluminum melting furnaces, particularly where special alloys are being melted. These refractories are

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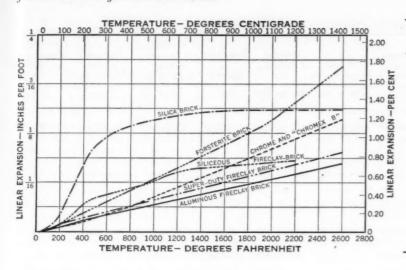


Fig. 1—Approximate reversible thermal

not readily wetted by molten aluminum and provide better resistance to thermal spalling than do basic refractories.

# Silica Refractories

In past years silica brick of conventional quality were used almost exclusively for open-hearth and electric furnace roofs. The utility of silica brick in these applications depends primarily upon their exceptional strength at high temperatures. For example, a silica brick subjected to a load of 25 psi shows negligible subsidence even at 3000 F. By way of comparison, a deformation as high as 6 per cent at 2462 F, with the same load, is permitted by the ASTM requirements for high-duty fireclay brick.

One of the characteristics of silica brick is relative sensitiveness to thermal shock within the low temperature range. The reason for this is made evident by a study of the thermal expansion curve shown in Fig. 1. The high rate of expansion up to 1000 F clearly indicates this characteristic. It is necessary to heat silica roofs very slowly through the critical range, and to make proper provision for thermal expansion by means of expansion joints. During construction, these are filled with cardboard or wooden strips which will burn out and permit the brick to expand freely.

Table 2—Composition and Properties of Silica Brick*

	Conventional	Super-Duty
Chemical Composition, Pct		
Silica, SiO ₈	95.63	96.33
Alumina, Al ₂ O ₃	0.75	0.28
Lime, CaO	2.60	2.74
Iron Oxide, Fe ₂ O ₃	0.75	0.56
Soda, Na ₈ O	0.04	0.04
Potash, K ₂ O	0.15	0.04
Titania, TiO ₂	0.08	0.03
Physical Properties		
Porosity, Pct	25	25
Refractoriness Under Load	1	
at 25 psi, °F	3000	3070
Linear Thermal Expansion	n	
at 2600 °F, Pct	1.30	1.30
* Typical average values	k.	

While super-duty silica brick is a relatively new refractory, it has contributed greatly to longer life of furnace linings. The important difference between conventional and super-duty silica brick are illustrated in Table 2. The lower content of alumina, titania and alkalies accounts for the higher refractoriness of the super-duty silica brick. They are made to meet the specification that the combined content of alumina, titania and alkalies may not exceed 0.5 per cent. The effects of these impurities on the melting point of silica are illustrated graphically in Fig. 2 in which it is shown that 1 per cent titania (TiO₂) lowers the melting point about 10 F, 1 per cent alumina (Al₂O₃) about 25 F, 1 per cent potash (K₂O) about 45 F, and 1 per cent soda (Na₂O) about 65 F.

Another significant property of super-duty silica brick is its increased ability to withstand pressure at

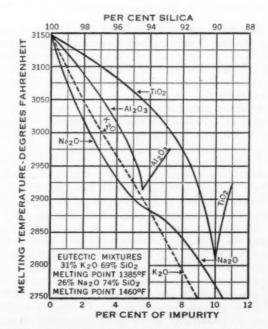


Fig. 2—Effect of certain impurities on melting of silica.

TABLE 3—PROPERTIES OF BASIC REFRACTORIES

Туре	Bond	Principal Minerals		pare prosi Pct	ty,	V Lb p	,	ght, Cu F	of I		lus ture,	Resistance to Spalling	Stability of Volume at High Temperatures	Strength at High Temperatures†	Thermal Expansion at 2600°F Pct
Chrome	Fired	Chrome- Spinel	18	to	21	190	to	200	1200	to	2000	Fair	Fair	Fair	1.2
Chrome- Magnesite	Chemical	Chrome- Spinel Periclase	18	to 2	22*	192	to	202	750	to	1200	Excellent	Excellent	Excellent	2.1
Chrome- Magnesite	High- Fired	Chrome- Spinel Periclase Forsterite	20	to	24	184	to	194	1100	to	1500	Good	Excellent	Excellent	1.2
Magnesite 83-88 MgO	Fired	Periclase	18	to	22	165	to	175	2000	to	3000	Fair	Fair	Fair	2.1
Periclase 92-93 MgO Magnesite- Chrome	Fired Chemical	Periclase Periclase Chrome-		to 2				177 187			2600 1400	Good Excellent	Good	Excellent	2.1 2.1
Spinel- Bonded Magnesite- Chrome	Fired	Spinel Periclase Chrome- Spinel Magnesia- Alumina- Spinel	21	to 3	24	170	to	180	800	to	1100	Excellent	Excellent	Excellent	1.5
The strens	th at high ter	Forsterite heating the brid mperatures is me F. "Good." 260	ck to	by 1	re no	n-perm	anei	of fa	ding in	ngree n t	he load	Good from pores.	Good	Excellent ponds to a tempera	1.8

high temperatures. While conventional silica brick of highest quality withstand a load of 25 psi up to about 3000 F before failure, super-duty silica brick withstand the same load up to about 3070 F.

In the roofs of basic open-hearth furnaces, superduty silica brick are used most economically for roof shoulders, with the conventional silica brick in the central portion of the roof. In many instances, superduty silica brick are used for the entire roof.

### **Basic Refractories**

For many years the only basic refractory brick produced were fired magnesite and fired chrome brick. However, as the result of continued research, more than a dozen different classes of basic refractories are now manufactured, as shown in Table 3. While there is considerable overlapping in their many uses, the distinctive properties of each render it particularly suited for certain conditions and purposes. Naturally in applications where two or more kinds perform equally well, cost governs the selection.

Hard-burned magnesite brick, containing approximately 88 per cent magnesia, have been used extensively in basic open-hearth steel furnaces. This type of basic refractory is used successfully for the melting zone linings of cupolas operated on basic practice.

Magnesite brick which contain 92 per cent magnesia are available. This refractory, accurately termed periclase, is higher in magnesia content and lower in other oxides such as silica, lime, ferric oxide, and alumina, than the conventional magnesite brick. The lower amount of impurities and the higher magnesia content provide better resistance to thermal spalling and a considerable improvement in volume stability and mechanical strength at high temperatures. The

periclase type of refractory may prove economical for use in basic electric melting furnaces, in melting furnaces for non-ferrous metals and in other applications where the operating conditions are severe.

The use of high-fired chrome-magnesite refractories in copper, nickel, lead, and aluminum furnaces has increased. The volume stability of these brick at high temperatures, and some of their other desirable physical properties are attributable to the unusually high temperatures at which they are fired. The strength of this refractory at high temperatures is such that in laboratory tests it withstands a load of 50 psi up to a temperature of about 3000 F. Moreover, its resistance to spalling is greater than that of other fired basic refractories. In strongly reducing atmospheres at high temperatures, the oxides of iron do not seriously affect the chrome-magnesite composition.

In some lead reverberatory furnaces, the hardburned chrome-magnesite refractory is used in lower and upper sidewalls and roofs and the longer service obtained well justifies the higher costs as compared with alumina-silica refractories. This refractory has high resistance to wetting by molten alumina. Advantage of this fact is taken in the use of the highfired chrome-magnesite refractory in the bottoms and lower sidewalls of aluminum melting furnaces. Another important factor in this application is avoidance of silicon pickup, which is especially important in the case of some alloys.

# Ganister-Fireclay Mixes

Considerable progress has been made within the past few years in the use of ganister-fireclay mixes for maintaining the melting zone linings of cupolas. The air-placement method of application makes it possible to use the material with the minimum amount of moisture and to obtain a lining of maximum density. This property is essential to satisfactory service. Ganister-fireclay mixes of correct chemical composition and proper screen sizing are available commercially in ready-mixed form.

A special ganister-fireclay mix recently developed contains approximately 88 per cent silica and 7 per cent alumina, with the remainder consisting of small percentages of titania, iron oxide, lime, magnesia and alkalies. The silica ganister used has a maximum alumina content of 0.5 per cent. The fact that the fireclay used in ganister-fireclay mixes must be of good quality has been frequently overlooked in the foundry. The clay should not only be of good plasticity but also have a minimum P.C.E. of cone 30 to 31. In order that the ganister-fireclay may have the desired refractoriness, the composition of the clay must be such that the alumina content of the mixture is more than 5.5 per cent. The eutectic between silica and alumina contains 5.5 per cent alumina and the melting point of alumina-silica mixtures rises as the alumina content increases above 5.5 per cent.

To obtain a lining of the required density, the ground silica ganister should contain the correct proportions of coarse, intermediate and fine sizes. Proper grain sizing will reduce the amount of rebound to the minimum.

It is exceedingly important that each operator of the air-placement gun accurately determine the proper moisture and pressure by actual trial in his own plant. Only by so doing can he obtain the maximum density in the lining and best results.

#### Hydraulic Setting Castable Refractories

While various grades of hydraulic setting castable refractories are available, this discussion will be confined to castable refractories of high strength, and to a 3000 F castable refractory as shown in Table 4. The regular castable refractory is shown in the table for purposes of comparison.

TABLE 4—PROPERTIES OF CASTABLE MATERIALS

	P.C.E.	Equivalent Temperature, F	Cold Crushing Strength, psi
Regular Castable Refractory	15-19	2690	1500
High Strength Castable Refractory 3000 F	15-16	2640	5000
Castable Refractory	35	3245	1500

Castable refractories are shipped in the dry form and require only mixing with water to prepare them for use. On drying, a hydraulic bond is developed that provides good strength throughout the entire body.

The high-strength castable refractory develops a cold crushing strength after drying of approximately 5000 psi; after heating to 1500 F, 2100 psi, and after heating to 2250 F, 3750 psi. In consideration of this high strength some service trials have recently been

made in the upper section of the cupola. The castable refractory was installed by the air-placement method in the charging door area. In service, the high-strength castable refractory satisfactorily withstood the impact of the charge and the mechanical abrasion which occurs in this section of the cupola.

A newer product is the 3000 F castable refractory, which posesses not only high refractoriness but also excellent volume stability at high temperatures. It can be poured into place, trowelled, rammed, or applied by the air-placement method. Excellent results have been obtained by using it for patching the rim and lips of iron ladles. Service trials now in progress employ the 3000 F castable refractory in the center sections of electric furnace roofs. While the trials have not all been completed, it appears that this type of material holds considerable promise for the severe service conditions involved.

#### Plastic Fire Brick

Plastic fire brick is manufactured from carefully selected calcined and bond clays which are ground, blended and mixed with the correct amount of water to give the best ramming consistency. Plastic fire brick of both high-duty classes and super-duty classes have been used for gray iron ladle linings, cupola troughs, and many other applications.

Plastic fire brick with added graphite is also available. Such material also contains special components that tend to seal the outside surfaces and thus retard oxidation of the graphite. The mixture is very refractory, and in the 5-hr reheat test at 2730 F shows no linear shrinkage. Mechanical penetration by molten metal is reduced or eliminated, because the graphite is not wetted by the metal. The plastic fire brick containing graphite has been used successfully for gray iron ladle linings, tap out blocks, and also for the cupola spout. However, it is not recommended for use in the linings of ladles for low carbon steel.

#### DISCUSSION

Chairman: R. H. STONE, Vesuvius Crucible Co., Swissvale,

Co-Chairman: W. R. JAESCHKE, The Whiting Corp., Harvey, III.

Recorders: R. H. STONE and W. R. JAESCHKE.

D. E. MATTHIEU:1 In our pipe foundry we tap metal from the cupolas at 2600 F and use a refractory material with a P.C.E. 27 instead of higher grade materials with P.C.E. 31 or 321/2 as recommended by the author. We find this material adequate for our service conditions and more economical than the usual air-placement ganisters or cupola brick. What is the author's

MR. CUSTER: The difference between P.C.E. 27 and 31 or 321/2 is not great and in your case the material is satisfactory to resist severe erosion. For conditions at higher temperatures that material might be unsatisfactory.

R. W. KRAFT, JR.:2 Why are magnesite brick not made in cupola block sizes?

MR. CUSTER: Magnesite brick have a high rate of expansion and are susceptible to spalling. Cupola block shapes have been tried and were found to spall much more than the smaller shapes like arch and key brick.

R. H. ZOLLER:3 Speaking of the life of high magnesite blocks, the larger shapes give better life if temperatures are constant.

¹ Chief Metallurgist, Alabama Pipe Co., Anniston, Ala. ² Metallurgist, American Brake Shoe Co., Mahwah, N. J. ³ President, The Zoller Casting Co., Bettesville, Ohio

This is particularly true of lower magnesite content blocks. If larger blocks are brought up to temperature more slowly their life will be just as good as that of the smaller sizes. Higher magnesite blocks are not as susceptible to spalling and cracking if made with maximum density and fired to high temperature. The larger blocks cut down on the joint length which is particularly desirable in resisting slag attack.

MR. CUSTER: Uniform temperatures certainly would be helpful. Also, if the high MgO-content brick were burned at high temperatures, the density would be greater with resulting better service.

MR. ZOLLER: I prefer the cupola block shapes because of less

joint area and resulting better service.

R. A. CHAVY:4 What is the influence of graphite addition to magnesite or other refractories on corrosion resistance? Refractory must be correct chemically to resist slags. High porosity is not necessarily bad as it helps to seal pores. For example: Porosity of 20 to 22 per cent of high magnesia refractories would be considered high for some refractories but actually resistance to molten slags is good. Graphite is a very refractory material and added to clay increases refractoriness. Molten metal does not attack clay mixtures with 20 to 25 per cent graphite.

MR. CUSTER: That is chemically correct especially when the composition of the slag is not too different from that of the refractory. When using 92-pct MgO brick of 20 to 22 per cent porosity the basic slag seals over the brick and good service is obtained. The addition of graphite to fireclay materials increases their refractoriness. Molten metal does not wet or pene-

trate graphite.

J. H. RICKEY, JR.:5 We have found that 5 per cent graphite was beneficial and it was suggested that benefit in that case might arise not from refractoriness imparted but due to low fusion ash of some graphites used in the U. S. which would serve to glaze the surface and thus protect the refractory from slag erosion.

CHAIRMAN STONE: What percentage of graphite is required? MR. CUSTER: The graphite content should be 15 to 20 per

H. W. LOWNIE, JR.:6 What minimum density is required for an air-placement type of cupola lining? Also, what is your

definition of a castable refractory?

MR. CUSTER: The minimum density for air-placement linings has not been defined. Castable refractories are those that when mixed with water can be poured and acquire a hydraulic set. Air-placement materials are blown into place and are not considered as castable refractories. The high-strength castable refractory is a mixture of calcined diaspore, fireclay and a heatresistant hydraulic cement.

CHAIRMAN STONE: Are natural sandstones still used for cupola linings?

MR. CUSTER: Sandstone linings are still used for cupola linings but the cost is high due to labor involved.

MR. MATTHIEU: In the Birmingham district air or gun-placed linings cost twice as much as sandstone block. This may be due to lower wages and class of labor employed in Birmingham on this job. The reverse cost experience was reported by others.

MR. RICKEY: In regard to firestone for acid cupola linings, remember that it is leafy in structure and must be placed carefully to avoid spalling. Layers should be perpendicular to the cupola wall when blocks are laid. We have found as little as 5 per cent addition of graphite helpful to fireclay refractories in counteracting the wetting tendencies of slags.

CHAIRMAN STONE: What kind of graphite was used in the 5 per cent addition? There is a difference between natural graph-

ite and manufactured carbon.

MR. RICKEY: The 5 per cent addition was of Mexican graph-

MR. KRAFT: The author shows that at an intermediate temperature the castable refractory has minimum crushing strength as compared with both the cast state and high-fired state. What is the reason?

MR. CUSTER: This is because the chemical bond imparts a high strength on air setting which is weakened by the firing until the high temperatures develop the ceramic bond by vitri-

fication of the clays.

H. A. HORNER:7 We experienced difficulties with the ganisters used in air-placement linings. Some do not stick in place well and we found variations of 37 per cent, 17 per cent, and 20 per cent clay contents in three different shipments of this material. The 37 per cent clay content material gave good results and the others not so good. What can be done about this?

MR. CUSTER: With such variations in clay content the water content must be adjusted. Good results can be obtained with clay as low as 10 per cent if the right amount of water is added. The recommended clay content is 15 to 20 per cent.

MEMBER: We found that 12 to 15 per cent clay content is

best and that too much moisture is very harmful.

MR. HORNER: What is the optimum percentage of moisture? MR. CUSTER: The correct amount is determined by trial. There is found to be less droppage with the gun mix when the batch is on the dry side than with higher amounts of water.

Co-CHAIRMAN JAESCHKE: There may be some misunderstanding as to what is meant by good staying qualities of airplacement ganisters. A wet plastic mixture may stick better as applied, with a small amount lost by rebound, but in at least one case observed this more plastic mixture did not give as good service in melting as did a leaner, drier, mixture that did not stick as readily and dropped more material during application.

MEMBER: It is my opinion that too little research work has been done on air-placement lining materials, which we call blow mixes. We find that 70 per cent of the dropped or rebounded material is the round pebble ganister. More research should be done with angular grain type material.

Engineer, E.S.F., Asnieres, France
 Sales Manager, Ironton Fire Brick Co., Ironton, Ohio
 Assistant Supervisor, Battelle Memorial Institute, Columbus, Ohio
 Metallurgist, Frick Co., Inc., Waynesboro, Pa.

# MECHANICAL PROPERTIES OF SPHERULITIC GRAPHITE CAST IRON

By

C. F. Reynolds and H. F. Taylor*

#### I. Introduction

Cast iron is an alloy of iron that usually contains from 2.5 to 4.2 per cent carbon, 1.0 to 4.5 per cent silicon, and less than 1.0 per cent manganese, plus small amounts of impurities such as phosphorus and sulphur. In this type of alloy, most of the carbon separates from the matrix either as flakes of graphite or as spherulites. When the graphite is in flake form as it is in gray cast iron, the metal has little if any ductility; gray cast iron has always been considered a brittle material. Malleable iron was developed to provide a material in which the graphite would be in a more desirable spherulitic form, with the casting qualities of iron but less brittle.

Metallurgists have long dreamed of casting iron in which the graphite precipitates as spheres and in which the ductility would approach that of steel. At the 1948 Convention of the American Foundrymen's Society, representatives of the B.C.I.R.A. in England¹ and of the International Nickel Company of America² introduced rival methods for producing spherulitic graphite cast irons.

The British used cerium and the Americans used magnesium to spheroidize the graphite; a typical flake iron was converted to spherulitic form by adding these materials to a ladle of iron. Barium, calcium, lithium, and strontium have also been tried as "spherulizing" agents, but of these only cerium and magnesium are successful commercially. Unfortunately cerium and cerium-base alloys are relatively expensive; also they can only be used for spherulizing hypereutectic irons. Magnesium and magnesium-base alloys spherulize both hypoeutectic and hypereutectic irons and are considerably cheaper than either cerium or any of its alloys.

Spherulitic graphite irons are popularly called "nodular" or "ductile" irons to distinguish them from other forms of cast iron. Known properties for this material in the as-cast condition have been as high as 80,000 psi U.T.S., 60,000 psi Y.S., with as much as

20 per cent elongation and reduction of area. These properties approach requirements for some Ordnance castings and the material was of interest to the armed services. Accordingly, a contract was made by Watertown Arsenal for research to be done at M.I.T. during the fiscal year August 1950 to August 1951, of which the following is a report.

#### II. Purpose

The purpose of this research was to determine the properties of the new ferrous alloy, spherulitic cast iron, particularly the effects of heat treatment and composition, to indicate to what extent this material might prove of practical value for Ordnance materiel.

#### III. Plan of Research

To the end discussed above, studies were made to determine—

(1) the effect of variations in melting practice on the mechanical properties of spherulitic iron,

(2) the efficacy and practicability of known methods for addition of magnesium alloy.

(3) types of magnesium alloys which could be utilized as an inoculant to produce spherulitic cast iron with optimum tensile properties,

(4) the effect of various percentages of carbon, silicon, manganese, sulphur, and phosphorus on the tensile properties of spherulitic cast iron,

(5) effect of heat treatment on the properties of the cast structure,

(6) the effect of section size (varying from  $\frac{1}{2}$  to  $\frac{4}{2}$  in.) on the yield and tensile strength, elongation, and reduction of area.

(7) a correlation between tensile properties and microstructure in an attempt to determine optimum mechanical properties obtainable.

#### IV. Experimental Procedure

A 300-lb, tilting type, alundum-lined induction furnace was used for melting most heats of spherulitic iron. Temperatures were taken with a Leeds and Northrup optical pyrometer. The first melts, Table 1, were all superheated to 2650 F and poured into a preheated, clay-graphite reaction crucible which con-

Metai rrocessing Division, Department of Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

TABLE 1-AVERAGE COMPOSITION OF FIRST MELTS

Carbon, %	3.50
Silicon, %	3.00
Manganese, %	0.24
Phosphorus, %	.025
Sulphur, %	.024
Magnesium, %	0.07

tained the graphite-spherulizing inoculant. The only inoculant available for the early experiments was an iron-silicon-magnesium alloy containing 43 per cent iron, 43 per cent silicon and 11 per cent magnesium. Two and two thirds pounds of this inoculant were added to each 100 lb of low sulphur iron. The reaction was moderately violent and continued until the crucible was almost filled. A dry, crusty slag floated on the surface of the iron. When the crucible was about three-fourths full, a post-inculant (ferrosilicon containing 75 per cent silicon) was added to the metal stream. The amount of ferrosilicon added was 1/3 lb per 100 lb of iron treated. Standard 1-in. keel-blocks were cast; the temperature of the metal at the time of pouring was about 2500 F.

The above method for making spherulitic cast iron proved unsatisfactory. Low spherulizing efficiencies were realized as indicated by the presence of appreciable amounts of unreacted inoculant in the slag and by a partially spherulized graphite structure. A typical cast iron structure, obtained by the above treatment, is

shown in Fig. 1.

In another melting procedure, the bath was superheated to 2850 F in the induction furnace and held 20 min; then it was cooled to 2750 F before tapping. The iron-silicon-magnesium inoculant was placed on the bottom of the preheated reaction crucible before tapping; the inoculant was added on the basis of 2% lb per 100 lb of low sulphur iron treated. Nickelmagnesium, containing about 15 per cent magnesium, was also tried. In this case, 1 lb of inoculant was used per 100 lb of low sulphur iron treated. When the reaction between the iron and inoculant was complete, the slag was skimmed from the surface of the metal and the post-inoculant stirred into the liquid iron. Four-fifths of a pound of ferrosilicon, containing 85 per cent silicon, was added as a postinoculant to each 100 lb of iron treated. Standard 1in, keel blocks were poured at about 2550 F. Microscopic investigations indicated the graphite was completely spherulized by following this melting procedure; however, appreciable amounts of free carbide were present, Fig. 2.

Based on experience and discussions with the International Nickel Co., a method of treatment was developed which proved completely satisfactory and was used throughout the investigation. In brief, the procedure is (1) to place 2% lb of Fe-Si-Mg or 1 lb of Ni-Mg per 100 lb of low-sulphur metal on the bottom of a receiving ladle and pouring the cast iron upon it. (2) The metal in this ladle is then skimmed of slag and poured into a second ladle. (3) The post-inoculant (4/5 lb ferrosilicon containing 85 per cent silicon and 1 oz calcium-silicide, per 100 lb metal) is added to the stream while the second ladle is being filled. Figure 3 shows a completely spheroidal structure produced by this method.

To determine the effect of time at the temperature of superheat upon spherulitic cast iron, a heat was made during which the bath was held at 2850 F for zero, 10, 20, 30, and 40 min before cooling to treating temperature. Fractions of metal at each time interval were tapped at 2750 F and treated. Tensile test data, Fig. 4, serves as a measure of graphite spheruli-

zation

There are four main proprietary magnesium-base alloys available commercially as inoculants for producing spherulitic cast iron. These are iron-silicon-magnesium, n i c k e l-magnesium, iron-nickel-silicon-magnesium and copper-magnesium; these contain from 10 to 20 per cent magnesium. Of these, only the first two were used in this research, as the latter ones were of no particular advantage and introduced an extra element unnecessarily.

A comparison was made of the relative effects of the iron-silicon-magnesium and nickel-magnesium alloys upon the tensile properties of nodular iron. The data, Table 2, are based on recovering the same amount of residual magnesium in the metal after casting. The efficiency of the nickel-magnesium alloy is greater than it is for iron-silicon-magnesium alloy and the amount of contained magnesium is higher; so one third as much nickel-magnesium was needed.

Considerable information3,4,5 has been published

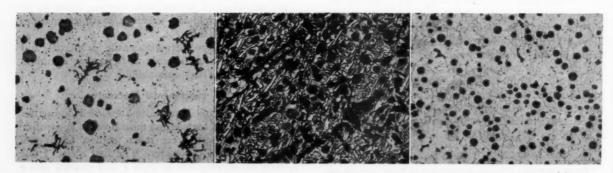


Fig. 1—Graphite structure due to improper melting.

Fig. 2—Free carbides due to improper post inoculant.

Fig. 3—Graphite structure produced by proper melting, inoculation, and post inoculation.

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TABLE 2—COMPARISON OF THE RELATIVE EFFECT OF FE-SI-MG AND NI-MG ALLOYS ON MECHANICAL PROPER-TIES OF SPHERULITIC IRON

Composition		
Carbon, %	3.38	3.49
Silicon, %	3.19	3.17
Manganese, %	0.24	0.28
Phosphorus, %	0.04	0.03
Sulphur, %	0.016	0.017
Magnesium, %	0.067	0.068
Nickel, %	0	0.85
As-Cast		
Tensile Strength, psi	73,200	85,000
Yield Strength, psi	56,600	61,500
BHN	170	182
Elongation, %	21	14
Reduction of Area, %	27.5	12
Heat Treated		
Tensile Strength, psi	73,200	73,500
Yield Strength, psi	56,000	57,000
BHN	158	169
Elongation, %	21	19
Reduction of Area, %	24	21
Average Recovery *	25	50
• % Mg _{added} = % Mg residual	+ % Sulphur.	

concerning the effect of various alloying elements upon the microstructure and mechanical properties of spherulitic cast iron. With the exception of most recent data regarding the effects of phosphorus⁶ and silicon,⁷ no systematic approach to the effect of chemistry has yet been made; consequently, it has been impossible to evaluate existing data quantitatively. In this work, a systematic study of the effects of carbon, silicon, manganese, phosphorus, and sulphur

Table 4—Mechanical Properties of Armco Iron and Carbon

Composition	
Carbon, %	3.27
Silicon, %	2.83
Manganese, %	0.003
Phosphorus, %	0.016
Sulphur, %	0.019
Magnesium, %	0.056
As-Cast	
Tensile Strength, psi	65,000
0 1	65,000
Yield Strength, psi	46,500
	47,750
BHN	149
	154
Elongation, %	22.7
	23.9
Reduction of Area, %	28.8
	28.2
Heat Treated	
Tensile Strength, psi	63,750
0 / 1	63,500
Yield Strength, psi	43,750
	43,500
BHN	149
	148
Elongation, %	25.8
~ /*	25.8
Reduction of Area, %	220
	26.8

were realized by varying the alloying element under consideration and holding all others constant. The carbon ranged from 2.26 to 4.00 per cent, silicon from 1.39 to 4.38 per cent, manganese from 0.033 to 2.13 per cent, phosphorus from 0.026 to 0.800 per cent, and sulphur from 0.017 to 0.09 per cent. The specific compositions studied are given in Table 3. The photomicrographs are included in Figs. 14, 15, 17, 18, 20, 21, 23, 24, 26, and 30 and tensile test data, plotted as a function of the variable alloying element, are shown in Figs. 16, 19, 22, 25, 27 and 29.

One heat was made using Armco iron melting stock and pure carbon to make a cast iron as pure as possible. The reason was to determine if small amounts of incidental impurities found in commercial iron had an unusually deleterious effect on spherulitic iron. The results are given in Table 4.

In many cast metals, section size has marked effect

Table 3—Compositions of Spherulitic Irons Investigated

No.	C %	Si %	Mn %	P %	Final S%	Residua Mg %	Ni %	
1	2.26	3.27	0.42	0.036	0.016	0.057		
2	2.45	3.13	0.42	0.036	0.016	0.054	_	
3	3.00	3.17	0.42	0.036	0.016	0.057		
4	3.55	3.05	0.42	0.036	0.016	0.045		
5	4.00	3.04	0.25	0.034	0.011	0.069	_	
_	Average	3.13	0.39	0.036	0.015	0.056	_	
6	3.55	1.39	0.27	0.036	0.016	0.057	0.72	
7	3.44	1.76	0.27	0.036	0.016	0.061	0.72	
8	3.44	2.15	0.27	0.036	0.016	0.059	0.70	
9	3.41	2.78	0.27	0.036	0.016	0.065	0.72	
10	3.49	3.17	0.28	0.032	0.017	0.068	0.85	
11	3.14	3.73	0.28	0.032	0.017	0.075	0.81	
12			0.28	0.032	0.017	0.068	0.89	
_	3.38	Average	0.27	0.34	0.016	0.065	0.85	
13	3.38	3.19	0.24	0.040	0.016	0.067	-	
14	3.25	3.19	0.46	0.040	0.016	0.068	_	
15	3.30	3.19	0.67	0.040	0.016	0.062	_	
16	3.38	3.19	0.92	0.040	0.016	0.058	_	
17	3.33	3.09	0.92	0.046	0.018	0.048	_	
18	3.19	3.15	1.67	0.046	0.018	0.054	_	
19	3.30	3.14	2.13	0.046	0.018	0.051	-	
_	3.30	3.16	Average	0.043	0.017	0.058	_	
20	3.40	3.40	0.24	0.026	0.015	0.063	_	
21	3.35	3.49	0.32	0.040	0.014	0.063	_	
22	3.35	3.49	0.32	0.080	0.014	0.064	_	
23	3.35	3.49	0.32	0.158	0.014	0.064	_	
24	3.35	3.49	0.32	0.340	0.014	0.064	_	
25	3.19	3.08	0.42	0.800	0.013	0.054	-	
_	3.33	3.41	0.30	Average	0.015	0.062		
26	3.16	3.35	0.25	0.030	0.017	0.085	_	
27	3.16	5.36	0.25	0.030	0.017	0.067	-	
28	3.16	3.30	0.25	0.030	0.018	0.058	_	
29	3.16	3.10	0.25	0.030	0.028	0.044	_	
	3.16	3.28	0.25	0.030	0.020	Average	-	
30	3.25	3.17	0.25	0.030	0.025	0.054	-	
31	3.49	3.17	0.28	0.032	0.017	0.068	0.85	
32	3.25	2.96	0.26	0.034	0.018	0.049	1.39	
_	3.33	3.10	0.26	0.032	0.020	0.057	Average	
33	3.27	2.83	0.033	0.016	0.019	0.056	_	

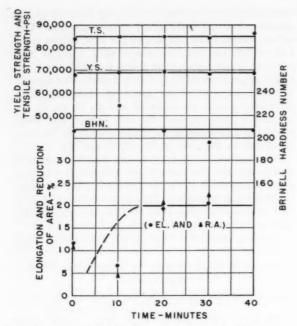


Fig. 4—Effect of holding time in furnace at 2850° F on mechanical properties.

upon microstructure and mechanical properties. Keel blocks were made, varying in cross-section from ½ to  $4\frac{1}{2}$  in. The castings were poured at about 2550 F. Tensile test data on specimens taken from the various sections, shown in Figs. 34 to 36, are plotted as a function of section size in Fig. 33. Photomicrographs are included in Figs. 31 and 32.

To determine how nearly the properties of spherulitic iron might approximate low-carbon steel, in particular its good strength and high ductility, specimens were heat treated to produce a structure in which the spherulites were distributed through a completely ferritic matrix. One inch cubes were packed in annealing boxes with cast iron chips and heated to 1650 F. After holding at temperature for 1 hr, the specimens were furnace cooled to 1275 F and held at this temperature for different lengths of time. The specimens were removed from the furnace at

TABLE 5—EFFECT OF SECTION SIZE ON MODULUS OF ELASTICITY

Section Size, in.	Modulus of Elasticity psi × 10 ⁻⁶
1/9	24.4
1/2 1/2	21.2
1 ~	30.4
1	30.0
2	31.0
2	27.5
3	26.0
3	28.0
41/6	33.0
2 2 3 3 4½ 4½	27.0
Average	27.9

successive intervals and cooled in air to room temperature. Brinell hardness was measured to determine the rate of pearlite decomposition. These data, plotted as a function of holding time at 1275 F, are included in Figs. 39 to 41. These particular heat treating temperatures are those used in industry. However, based on current results and the constitution diagram of iron-carbon-silicon the temperatures were too low for optimum ferritizing treatment. The tensile bars were heated to 1768 F for 1 hr, furnace cooled to 1350 F, and held 5 hr. Tensile test data for the heat treated specimens are included in Figs. 16, 19, 22, 25, 27 and 29.

Evaluation was made of modulus of elasticity using specimens cut from the various sections (from 1/2 to 41/2 in. thick). The values are reported in Table 5.

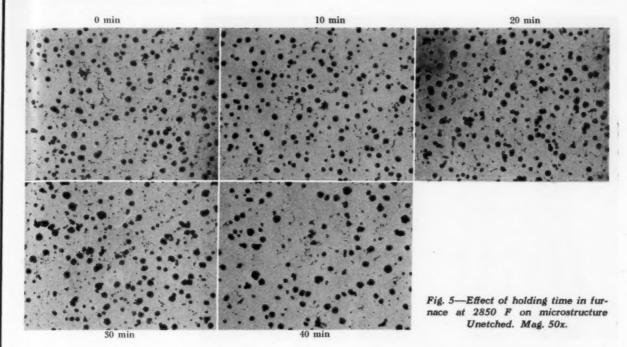
#### V. Discussion of Results

For simplicity, results of the various experiments will be discussed under separate sub-headings.

A. Microstructure as a check on practice—The spherulitic form of graphite is so well defined it is easy to predict inoculating efficiency from microstructures. Figures 1, 2, and 3 show respectively an unsatisfactory structure containing so-called "crab-graphite," an unsatisfactory structure containing carbide, and a fully satisfactory matrix of ferrite with well formed and well dispersed spheroids of carbon. The unsatisfactory structures are due to improper metal temperature and incomplete post inoculation; the good structure was dependably produced by close control of metal temperature, by altering the composition of the post inoculant (to contain both ferro-silicon and calcium-silicide) and by reladling the metal to insure proper mixing.

B. Effect of holding time before inoculation—2850 F was selected as suitable metal temperature for determining the effect of holding time upon the efficiency of inoculation. It was thought undissolved particles in the melt might act as foreign nuclei and alter the spherulizing behavior. Figure 4 contains the data showing effect of holding time in furnace on mechanical properties. Ductility (% El. and R.A.) increased to a maximum of 20 per cent after 20 min holding time. At higher temperatures the critical holding time would surely decrease. After 20 min, the ductility reached a constant value which was essentially unaltered up to 40 min holding time. The photomicrographs of Fig. 5 show evidence of an undesirable phase in the samples held for 0 and 10 min; these are the samples giving low ductility.

C. Effect of holding time after inoculation—It is later shown that some residual magnesium must be left in the iron after inoculation to produce a satisfactory spherulitic structure; this amount is from 0.045 to 0.09 per cent. The amount of residual magnesium decays with holding time in the ladle. To determine the critical time for conditions used in the laboratory, ladles of treated metal were held for 2, 4, 8, 16 and 32 min. Analyses were obtained for residual mag-



nesium and the results are given in Fig. 6. This shows that holding times up to 20 min, at least, are not deleterious. For larger volumes of metal, for ladles with different shapes, the critical time could be expected to vary. It is naturally best to pour the metal as soon as possible after treatment, but it is published³ that holding time up to 1 hr in large ladles still gave satisfactory results.

D. Fluidity measurements—Using the spiral fluidity mold developed at N.R.L.,8 the fluidity of spherulitic iron was compared with gray iron of the same base composition. Temperatures were measured with a calibrated Pt-Pt 10% Rh thermocouple and spirals were poured successively as the metal cooled in the ladle. The metal temperature was equalized in the ladle by stirring before pouring each spiral.

The fluidity of spherulitic iron is the same as gray iron of the same base composition, as shown by the juxtaposition of the two curves in Fig. 7.

E. Shrinkage characteristics-No quantitative mea-

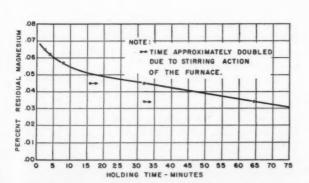


Fig. 6—Effect of holding time in ladle on residual magnesium.

surements were made on the shrinkage characteristics of spherulitic iron; the qualitative visual appraisal seemed adequate. In early experiments, tests were made to determine the best type casting for obtaining material for tensile testing (the keel block proved best). Figure 8 shows how the comparative castings were made. Figures 9, 10, and 11 indicate spherulitic iron shrinks considerably during freezing. Gross shrinkage, Fig. 9, sponge shrinkage, Fig. 10, and surface dishing, Fig. 11, indicate shrinkage similar to that of cast steel. Based on observations of many castings, it can be observed generally that shrinkage is slightly less than for steel and the shrinkage cavity is flatter and less "carrot shaped." Also, there is more of a tendency for sponge shrinkage. The flat nature of the shrinkage cavity indicates the shrinkage interval is considerably shorter for nodular iron than for cast steel; in cast steel shrinkage is a continuous process from beginning to end of solidification, giving rise to the "carrot shape" of the shrinkage cavity.

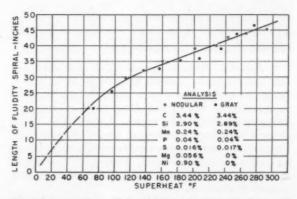


Fig. 7—Comparison of the fluidity of spherulitic iron and gray iron.

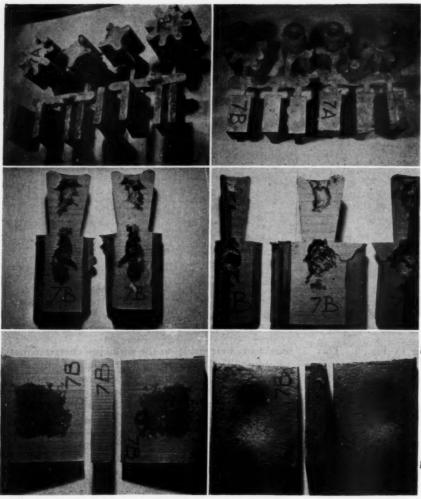


Fig. 8—Various test coupons cast in common mold from a single sprue.

Fig. 9—Cloverleaf castings having gross shrinkage.

Fig. 10 (Left)—½-in, keel block having sponge type shrinkage.

Fig. 11 (Right)—1/2-in. keel block having dish type shrinkage.

Addition of alloys increased shrinkage. The most spectacular effect was shown by phosphorus; Figure 12 shows castings containing 0.04 and 0.800 per cent phosphorus. Not only is gross shrinkage much greater for the higher phosphorus content, but the casting increased in size during or after solidification. The molding sand was actually forced from the flask, as shown in Fig. 13. Dimensions are given on the castings to indicate final size; the same pattern was, of course, used to prepare each mold originally, and except for phosphorus content, the analyses of the castings are the same.

F. Pinholes—Pinholes in spherulitic iron are a serious and recurrent problem in industry, as in manufacture of castings from most other metals. In early experiments of this investigation, much trouble was encountered from this source. Later, after many changes including better overall control of processing, composition, and pouring temperature, pinholes were no longer a serious problem. From work to date, it is not possible to develop a theory for the cause and control of pinholes.

G. Effect of alloying elements—The base iron used

for evaluating the effect of composition contained 3.30 per cent carbon, 3.20 per cent silicon, 0.24 per cent manganese, 0.04 per cent phosphorus, and 0.017 per cent sulphur; each element was varied independently from the above base values, holding all else constant. The particular values for carbon and silicon were chosen from previous data to give an iron having relatively good properties without addition of alloys.

1. Effect of carbon—The photomicrographs of Figs. 14 and 15 and the curves of Fig. 16 show the changes in microstructure and mechanical properties as carbon is increased from 2.26 to 4.00 per cent; photomicrographs and curves are included for heat treated and as-cast properties.

Maximum tensile strength occurred at about 3.0 per cent carbon; tensile strength increased from 80,000 psi at 2.26 per cent carbon to 100,000 psi at 3.0 per cent carbon falling off to about 70,000 psi as carbon increased to 4.0 per cent. As-cast ductility, elongation and reduction of area, increased steadily from 2.0 per cent to about 25 per cent over the same range. Hardness fell steadily from 250 BHN to 160 BHN, while yield strength dropped from 75,000 psi to 50,000 psi.

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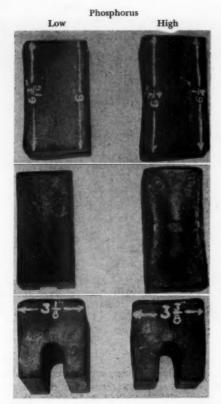


Fig. 12-Effect of phosphorus on shrinkage and growth.

Heat treatment lowered tensile and yield strength and hardness, and levelled out the effect of carbon; ductility was increased for all values of carbon except 4.0 per cent, where the structure was already com-

3.55%

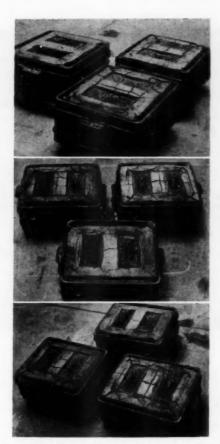
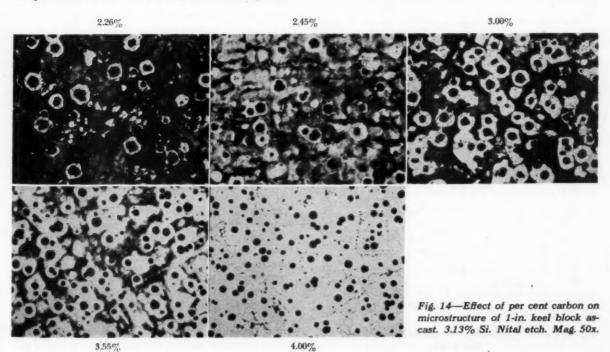
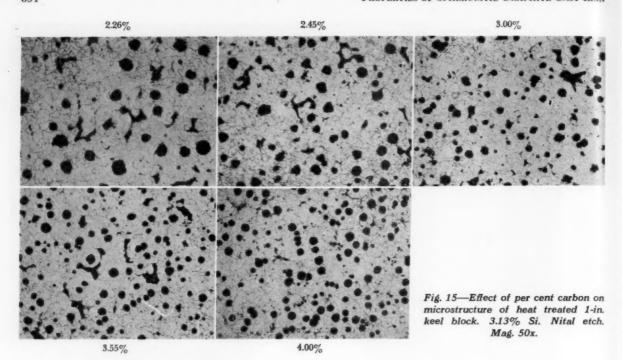


Fig. 13-Effect of low and high phosphorus castings on sand molds.





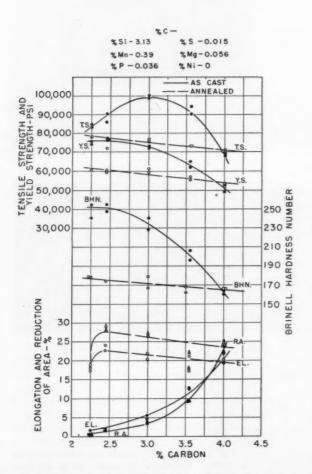


Fig. 16—Effect of carbon on mechanical properties of 1-in. keel block.

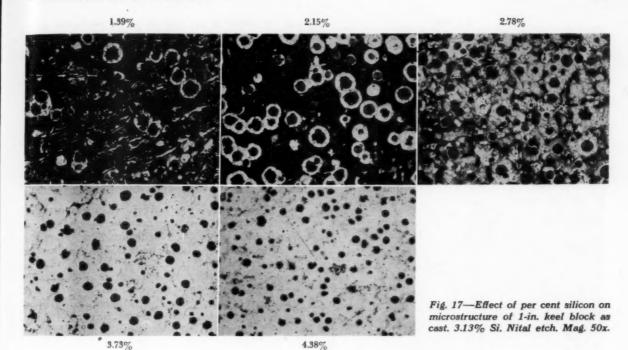
pletely ferritic and no decomposition of pearlite was needed. The maximum change in ductility due to heat treatment occurred for the 2.45 per cent carbon iron where it increased from about 2 per cent to 28 per cent reduction of area.

The effect of heat treatment, seen by comparing microstructures, is largely one of decomposing pearlite. The spherulitic structure is not satisfactory in the lower carbon compositions. There will, of course, be some stress relief due to heat treatment, but for large, slowly cooled, castings it is not significant.

2. Effect of silicon—The effect of silicon, varying from 1.39 per cent to 4.38 per cent, is shown in the as-cast and heat treated microstructures of Figs. 17 and 18, and in the curves of Fig. 19. At low silicon levels, the microstructure is chiefly pearlitic with massive carbides. The spherulites are large and relatively few in number. That silicon is a strong ferrite former is shown in the microstructures; at 3.0 per cent silicon the matrix is largely ferritic.

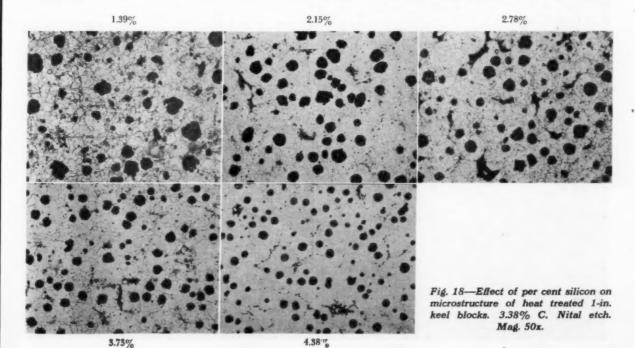
As-cast tensile strength decreases from 100,000 psi at 1.39 per cent silicon to 80,000 psi at 3.17 per cent and increases to 98,000 psi at 4.5 per cent; yield strength remains fairly constant at 60,000 psi to 3.0 per cent silicon and increases to 75,000 psi at 4.5 per cent. Brinell hardness drops from 230 to a low of 180 and increases to 215, while ductility increases from 5.0 per cent to about 18.0 per cent and then decreases to 8.0 per cent. The critical silicon content for maximum and minimum values of mechanical properties is about 3.0 per cent.

Obviously the as-cast mechanical properties follow the relative proportions of carbide, pearlite, and ferrite, and these will be influenced by cooling rate. Heat treatment reduces all structures to a ferrite mat-



rix. Tensile strength increases steadily from 60,000 psi to 95,000 psi at 4.38 per cent silicon, yield strength from 45,000 to 75,000 psi; hardness increases from 120 BHN to 220 BHN over the same range, and ductility drops steadily from a maximum of about 25 per cent to 15 per cent. The increase in heat treated tensile and yield strength and hardness and reduction in ductility is due directly to the solid solution effect of silicon in ferrite. Silicon is an excellent ferrite strengthener.

3. Effect of manganese—Manganese is a carbide former in spherulitic iron and contributes to pearlite formation; dissolved in ferrite, it behaves as a solution hardening agent. Tensile and yield strengths and hardness increases, Fig. 22, are accompanied by a marked decrease in ductility. Addition of 1.0 per cent manganese increased as-cast tensile and yield strength values some 3,000 psi and dropped ductility from about 30 to 15 per cent. Manganese, Figs. 20 and 21, does not appear to effect the size or number



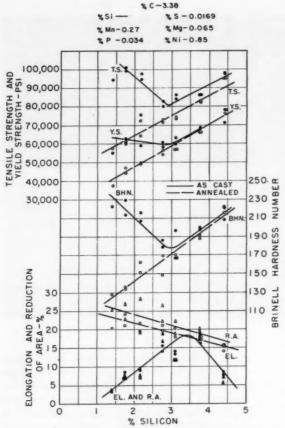


Fig. 19—Effect of per cent silicon on mechanical properties of 1-in. keel blocks.

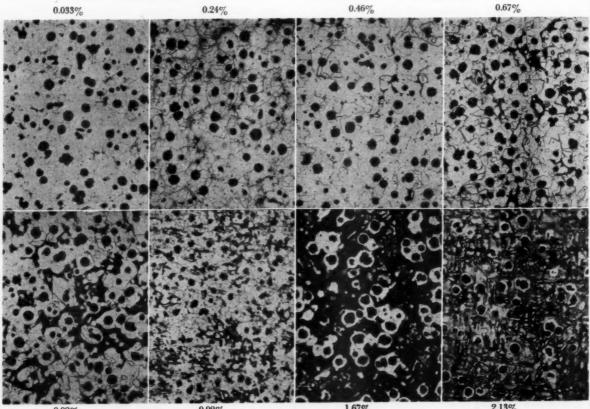
of spherulites.

Values for heat treated specimens were not materially different than as-cast values except where carbides were prevalent at about 1.0 per cent. At 2.0 per cent manganese, ductility in heat treated specimens was only about 10.0 per cent for tensile and yield strengths of 70,000 and 55,000 psi respectively. Five hours at 1350 F was not sufficient to completely decompose the pearlite of the high manganese irons; carbides were also detrimental to graphite shape.

4. Effect of phosphorus—As phosphorus is increased from essentially 0 to 0.8 per cent, the amount of pearlite increases in the as-cast structure, Figs. 23 and 24. From the curves of Fig. 25, there is a marked increase in tensile strength from 65,000 to 82,000 psi as phosphorus increases from 0 to 0.08 per cent; yield strength increases from 45,000 to 70,000 psi in going from 0 to 0.34 per cent phosphorus, at which point it is interesting to note the yield to tensile ratio is unusually high, about 92 per cent. Ductility drops sharply from about 25 per cent to a low of 3.0 per cent at 0.2 per cent phosphorus.

Heat treatment levels out the effect of phosphorus

Fig. 20—Effect of per cent manganese on microstructure of 1-in. keel blocks as-cast. 3.30% C, 3.16% Si. Nital etch. Mag. 50x.



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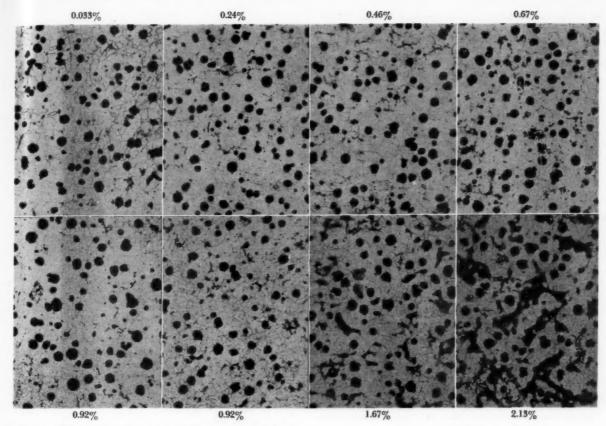


Fig. 21—Effect of per cent manganese on microstructure of heat treated 1-in. keel blocks. 3.30% C, 3.16% Si. Nital etch. Mag. 50x.

on yield and tensile strength, and hardness, and markedly lessens the slope of the curve showing decreased ductility; the decrease is steady from about 25 per cent to essentially 0 per cent at 0.60 per cent phosphorus.

Attention is called to the photographs of Fig. 12, which shows the unique shrinkage behavior of high phosphorus, 0.80 per cent, spherulitic iron. Not only did the casting "grow" in size during or after solidification, but the surface of the casting is cracked and veined in a network, or checkerboard, fashion. The photomicrographs of Figs. 23 and 24 show considerable steadite, an extremely low melting point ternary eutectic of alpha ferrite, iron phosphide, and cementite, surrounding the previously solidified iron.

Phosphorus appears to reduce spherulite size and increase the number of spherulites; as compared with an 0.04 per cent phosphorus iron, Fig. 24, 0.80 per cent phosphorus, about doubles the number of spherulites and halves their size.

5. Effect of residual magnesium—Magnesium is a strong deoxidizer and desulphurizer; by some mechanism, not completely understood, magnesium also acts to form or stabilize carbides. The role which seems of paramount significance in spherulizing graphite is the ability of magnesium to desulphurize iron; at least one rigid requirement for spherulitic

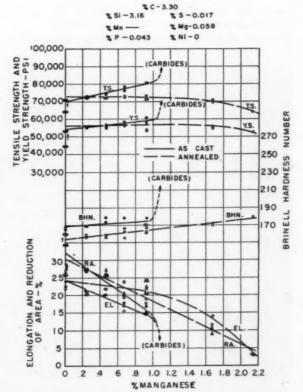
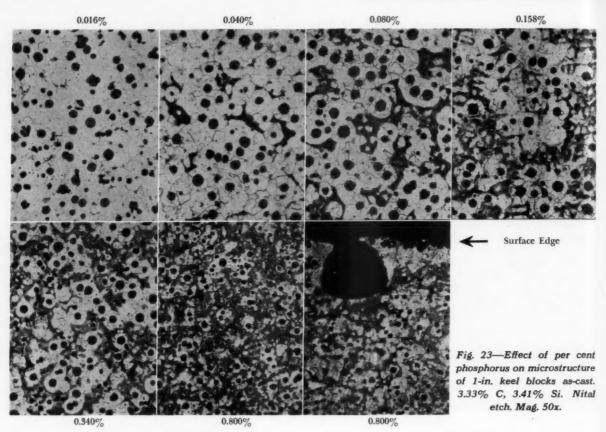


Fig. 22—Effect of manganese on mechanical properties of 1-in. keel blocks.



irons is that the sulphur remaining in iron after treatment must not be higher than about 0.02 per cent. The form in which magnesium is added seems unim-

portant from the standpoint of effect upon the form of graphite. Pure magnesium will spherulize graphite in iron, but the reaction is prohibitively violent and

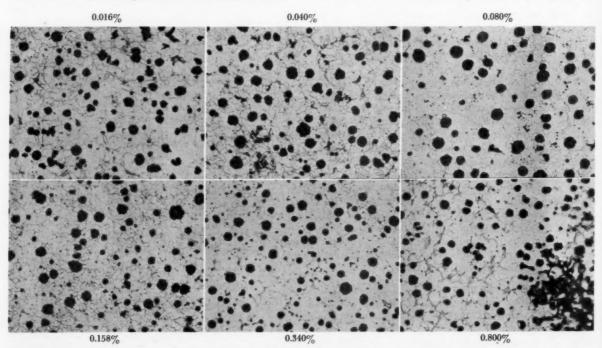
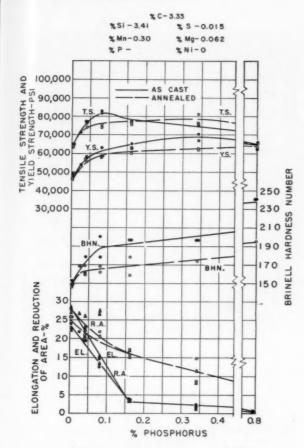


Fig. 24—Effect of per cent phosphorus on microstructure of heat treated 1-in. keel blocks. 3.33% C, 3.41% Si. Nital etch. Mag. 50x.



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Fig. 25—Effect of phosphorus on mechanical properties of 1-in. keel blocks.

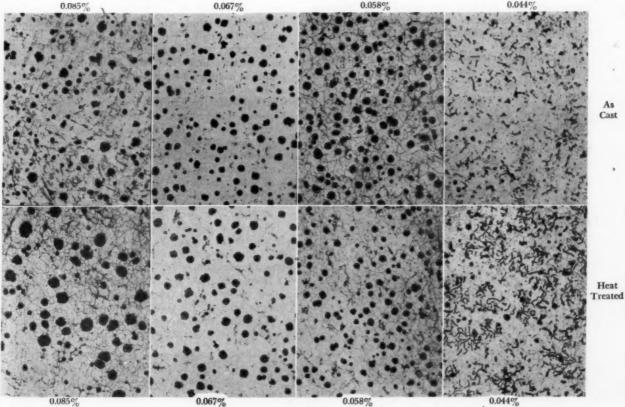
copper, nickel, or ferro-silicon is alloyed with magnesium to reduce violence of reaction. The sulphur and magnesium combine to form a sulphide which rises and can be skimmed from the surface of the molten metal. It is essential that some residual magnesium remain in the iron after treatment; the form in which it exists is not currently known.

Figure 27 shows the effect of residual magnesium on mechanical properties. The composition of base iron selected is one known to be typical and satisfactory for treatment. Below about 0.04 per cent residual magnesium the iron remains gray and above about 0.085 per cent becomes mottled. This limits the permissible safe range to these values, i.e. between 0.04 and 0.085 per cent residual magnesium. The strength, hardness, and ductility all rise sharply in the region of 0.04 per cent residual magnesium and strength and hardness may rise or fall sharply again at 0.085 per cent; at the higher limit ductility drops sharply.

Heat treatment does not change mechanical properties of the as-cast iron; this is because the size and composition of castings were purposely selected to produce a completely ferritic matrix.

Figure 28 shows a linear relationship for weight

Fig. 26—Effect of per cent residual magnesium on microstructure of as-cast and heat treated 1-in. keel blocks. 3.16% C, 3.28% Si. Nital etch. Mag. 50x.



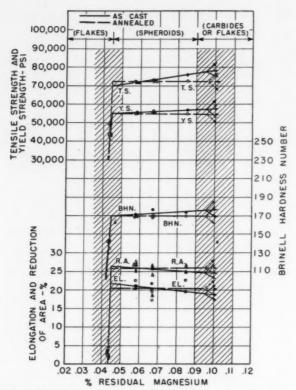


Fig. 27—Effect of residual magnesium on mechanical properties of 1-in. keel blocks.

per cent sulphur added to weight per cent residual magnesium lost. Sulphur was added to a ladle of metal tapped at intervals from a large crucible which originally contained 0.085 per cent magnesium. The

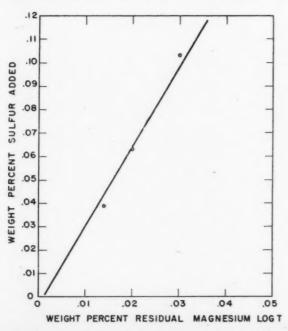


Fig. 28—Effect of sulphur added to the ladle on residual magnesium.

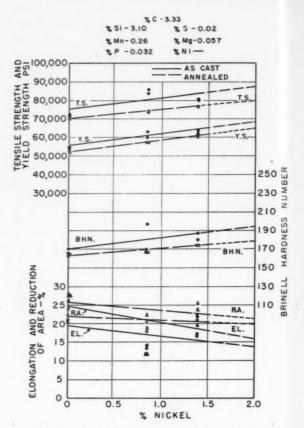


Fig. 29—Effect of nickel on mechanical properties of 1-in. keel blocks.

magnesium sulphide was skimmed from the first ladle before reladling for inoculation.

The effect of residual magnesium on microstructure is shown in Fig. 26. For increasing content the graphite particles become shorter and thicker until at about 0.04 per cent residual magnesium they become spherical. No further change in microstructure occurs to 0.10 per cent residual magnesium, when the iron becomes mottled and carbides or graphite stringers appear. This upper limit depends somewhat upon post-inoculation, chilling tendency of the iron, and section size of casting, and so is not a rigorous limit; it is known, however, the upper limit is very close to 0.10 per cent residual magnesium.

6. Effect of nickel—Figure 29 and Figure 30 show the effect of nickel is entirely one of solid solution in the ferrite. The strength and hardness increase and ductility drops slightly as nickel increases from 0 to 1.5 per cent.

Heat treatment lowers tensile and yield strength about 5,000 psi and raises ductility from about 20 to 25 per cent.

The effect upon microstructure is nil as the nickel is completely dissolved; this is shown in Fig. 30.

H. Effect of section size— The cooling rate of spherulitic cast iron determines the amount of pearlite and, within certain limits, the size of the spherulites. The amount of pearlite, in turn, effects the strength, hard-

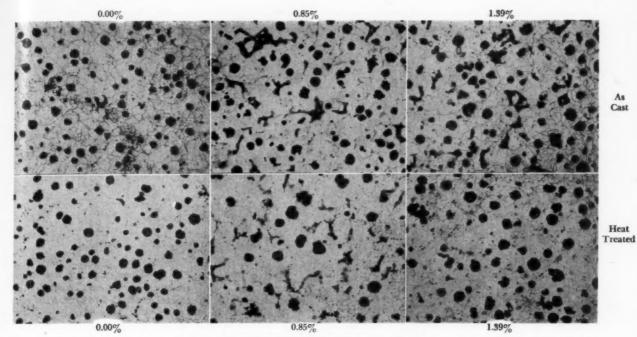


Fig. 30—Effect of per cent nickel on microstructure of as-cast and heat treated 1-in. keel blocks. 3.33% C, 3.10% Si.

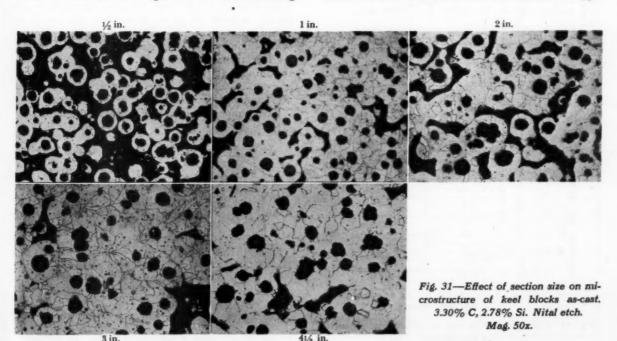
Nital etch. Mag. 50x.

ness, and ductility of the iron, while spherulite size, again within limits, is not particularly significant.

The photomicrographs of Figs. 31 and 32 show the decreasing proportions of pearlite in the microstructure as section size increases. This is a direct effect of cooling rate; the curves of Fig. 33 show a reduction in strength and hardness, and increase in ductility for specimens from bars 1/2 to 11/2 in. thick, and little or no effect for larger bars in which cooling

rate was slow enough to allow an almost completely ferritic matrix in the as-cast condition. The photomicrographs also show the relatively greater change (pearlite decomposition) in the smaller sizes due to heat treatment.

The properties of the 1-in. section are not in keeping with those of the other sizes, although all bars were poured from the same heat of metal; this is because the 1-in. bars were of the inverted "U" type,



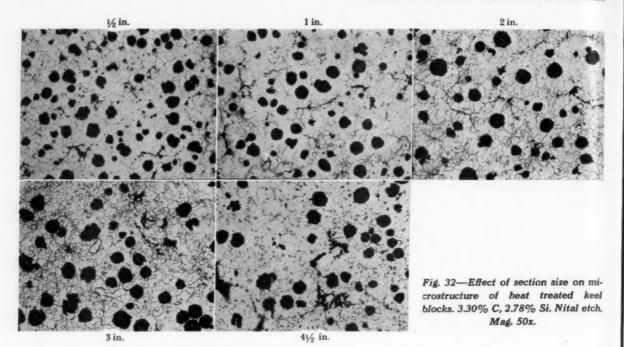


Fig. 34, and all others were "Y" shaped, Figs. 34, 35, and 36. Since the two legs of the keel block were so short and close together, the net effect was that of a

3.30 %SI - 2.78 % S - 0.016 % Mn-0.26 % Mg-0.051 % NI -0.890 % P - 0.030 100,000 DN 90,000 AS CAST ANNEALED STRENGTH STR TS 50,000 Y.S 250 8 40,000 230 30,000 220 SS OP RHN 170 ELONGATION AND REDUCTION OF AREA - % BHN 150 130 N 30 25 20 15 10 AND 5 0 0.5 1.0 1.5 2.0 2.5 3.0 3.5 4.0 4.5 5.0 SECTION - INCHES

Fig. 33—Effect of section size on mechanical properties of keel blocks.

much thicker section. No reason is evident why the microstructure and properties of this bar are so much better than for any of the other specimens.

The proportions of pearlite and the size and number of spherulites were determined by lineal analysis and plotted in Figs. 37 and 38 as a function of section size. These curves, when compared with mechanical property curves of Fig. 33, show clearly the effect of pearlite and spherulite size.

I. Effect of heat treatment— The effect of heat treatment on mechanical properties and microstructure of spherulitic iron has been discussed throughout the report; Figures 39 to 41 show how the hardness (which is taken as an index of rate of pearlite decomposition) varies as specimens are held at 1650 F for 1 hr, furnace cooled to 1275 F, and removed from the furnace at regular time intervals. The rate and amount of softening varies slightly with alloy and alloy content; in most cases 6 hr is adequate.

The compositions of base irons used for this in-

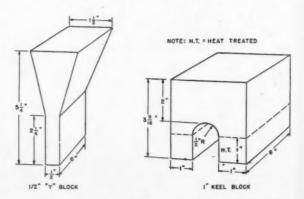


Fig. 34—Location of test bars in 1/2 and 1-in. castings.

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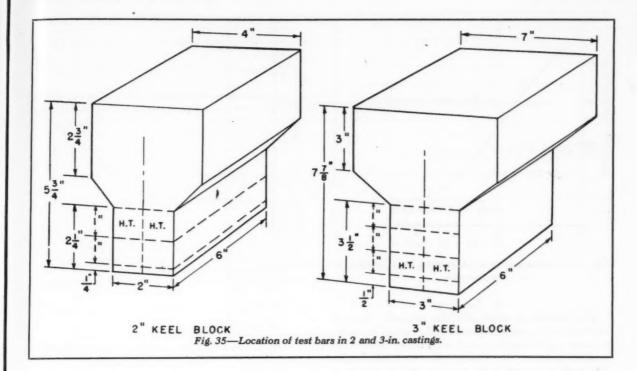
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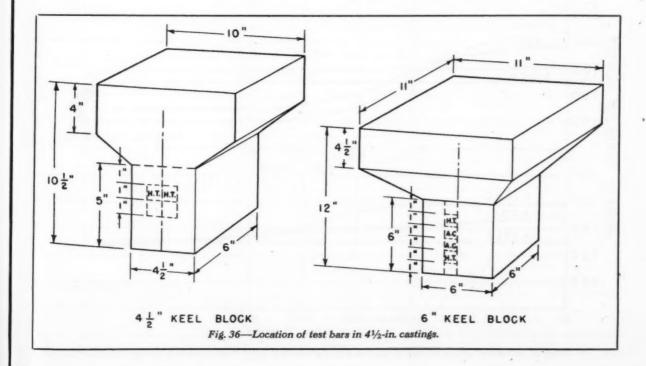
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vestigation were confined generally to those which would yield essentially a ferritic, or ferritic-pearlite matrix; and so the influence of heat treatment is not great, being confined to pearlite decomposition and some stress relief. Spherulitic graphite structures can be produced in martensitic, austenitic, acicular, and

other structures where the effect of heat treatment would be more pronounced; also, these irons can be supercooled by chilling to form spherulites of microscopic proportions which can then be grown to large size by annealing. Ductility would also be more markedly improved in small castings where stress relief would be a consideration.



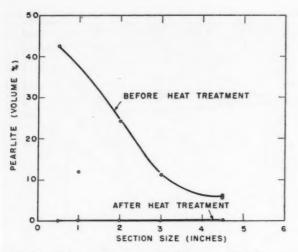


Fig. 37—Volume per cent of pearlite as a function of section size.

#### VI. Summary

The newness of spherulitic iron precludes rigorous appraisal of its potential future and position as an important engineering material; from all pertinent facts to date, however, this future looks extremely bright. The work reported herein is not extensive enough to provide all data for an intelligent estimate of the extent to which nodular iron can be used for military materiel; however, the following specific conclusions may be drawn.

1. As-cast tensile and yield strengths for ferritic spherulitic iron drop sharply from 93,000 and 70,000

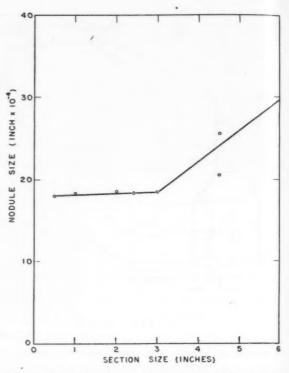


Fig. 38—Spherulite size as a function of casting section size, as-cast condition.

psi to 80,000 and 60,000 psi as section size changes from  $\frac{1}{2}$  to 1 in. The drop in these properties is then gradual to 70,000 and 53,000 psi for section  $4\frac{1}{2}$  in.

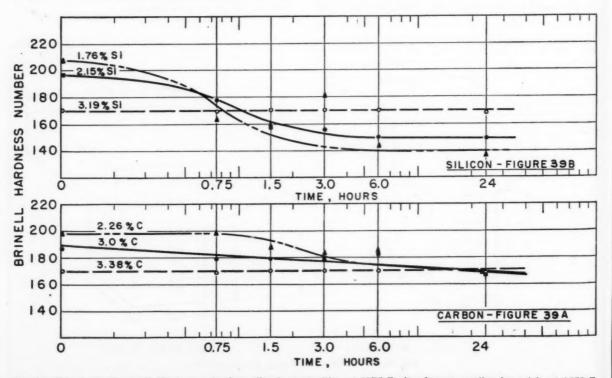


Fig. 39—Effect of carbon and silicon on rate of pearlite decomposition at 1275 F after furnace cooling from 1 hr at 1650 F.

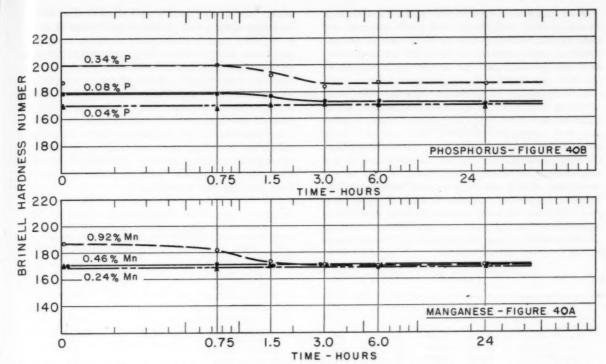


Fig. 40—Effect of manganese and phosphorus on the rate of pearlite decomposition at 1275 F after furnace cooling from 1 hr at 1650 F.

thick. Hardness follows strength properties closely and ductility increases initially from 5 per cent elongation and reduction of area in ½-in. sections to 15

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per cent in 4½-in. section. This reflects directly the trend in relative amounts of pearlite transformed during solidification and cooling.

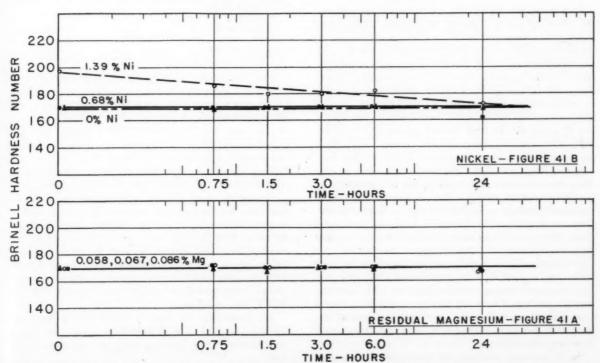


Fig. 41—Effect of residual magnesium and nickel on rate of pearlite decomposition at 1275 F after furnace cooling from 1 hr at 1650 F.

Heat treatment completes the transformation to a ferritic matrix and tensile and yield strengths are constant for all section sizes at 70,000 and 52,000 psi; hardness is also constant at 160 BHN, while ductility decreases from 22 per cent in the 1/2-in. section to about 15 per cent for the 41/2-in. specimens.

2. Comparison of photomicrographs and corresponding mechanical properties indicate that microstructure is a good index of quality. Table 6 shows the best properties obtained for pearlitic (high strength-low ductility) and ferritic (low strength-high ductility) spherulitic irons. It is obvious that any combination of strength and ductility can be obtained by varying the ratio of these structures. The "typical properties" column indicates values that could be expected conservatively from day to day shop operations. For optimum strength without alloying, an all pearlitic matrix is required and for optimum ductility an all ferritic matrix.

TABLE 6

	T.S., psi	Y.S., psi	EL., %	R.A., %	BHN
As-Cast					
Best Strength*	100,000	73,000	5	4	235
Best Ductility	73,000	56,000	21	27.5	170
Typical Prop.	80,000	60,000	15	15	185
Heat Treated					
Best Strength*	95,000	78,000	15	15	220
Best Ductility	78,000	60,000	24	20	175
Typical Prop.	70,000	55,000	20	20	160
Typical Prop.  * A higher strength coelement at a time—	ould be ob	tained by	varying	more th	

3. Figures 16, 19, 22, 25 and 27 show respectively the effects of carbon, silicon, manganese, phosphorus and residual magnesium on the tensile properties of spherulitic iron; curves are given for the as-cast and heat treated condition. The curves are easy to analyze and a concise verbal picture has been provided under discussion of results; so it is not considered necessary to re-analyze the data in detail here. Nickel, silicon, manganese, and phosphorus contribute to the mechanical properties of heat treated spherulitic irons by their solid solution effect on the ferrite. The effect on as-cast alloys depends jointly upon cooling rate (section size) and upon the particular effect of the alloy on formation and stabilization of carbides and pearlite; so it is not possible to predict the effect rigorously, except as found for the particular conditions of this research. However, it is felt the limits given for optimum amounts of each element are fair and are commercially applicable.

Although nearly any hypo- or hypereutectic cast iron can be used as a base for nodulizing, the preferred analysis for an average high ductility, ferritic, nodular iron is 2.8 to 4.2 per cent carbon, 1.5 to 4 per cent silicon, 0.25 to 0.75 per cent manganese, less than 0.25 per cent phosphorus, and less than about 0.06 per cent sulphur.

4. It was not necessary to investigate all possible magnesium alloys which could be used as inoculants for making spherulitic iron. Many experiments made

before and during initial tests showed that any inoculating agent which produced spherulites of carbon in a ferrite matrix would yield the same mechanical properties as any other; except, of course, for good or bad effects introduced by solution of an incidental element. For example, nickel, silicon or copper used as a carrier for magnesium would produce a solid solution effect corresponding to the alloy and the amount added; if the carrier inadvertently introduced some titanium, carbides might form and the graphite might not be satisfactorily spherulitic. Table 7 lists 23 magnesium alloys and their approximate efficiency. This list was compiled from the authors' data and data given in the literature.

TABLE 7—ALLOYING EFFICIENCY OF VARIOUS MAGNESIUM INOCULANTS

Alloy	Approximate Efficiency, %
1. Mg-Cu 50-50	10
2. Mg-Cu 30-70	13
3. Mg-Cu 20-80	15
4. Mg-Cu 10-90	23
5. Mg-Cu-Zn 50-35-15	5
6. Mg-Cu FeSi 6-10	32
7. Mg-Cu FeSi 13-10	24
8. Mg-Ni 50-50	8
9. Mg-Ni 20-80	40
10. Mg-Ni 10-90	50
11. Mg-Si 63-27	8
12. Mg-Fe-Si 11-43-43	25
13. Mg-Zn 50-50	<5
14. Mg-Zn 25-75	<5
15. Mg-Bi 50-50	<5
16. Mg-Bi 15-85	10
17. Mg-Al 55-45	15
18. Mg-Al 20-80	8
19. Mg-Li 90-10	6
20. Mg-Sb 20-80	<5
21. Mg-Si 80-20	10
22. Mg. airplane scrap	7
23. Cell Mg. 100	<5

The four typical magnesium alloys currently available are copper-magnesium, nickel-magnesium, iron-silicon-magnesium and nickel-silicon-magnesium. Of all these, iron-silicon-magnesium proved fully as good as any and, for purposes of this research, had the added advantage of not introducing an extra element not already required in cast iron.

5. A simple, fool-proof, method for treating gray iron to make spherulitic iron is described below.

Spherulitic iron of high quality can be made dependably by treating low sulphur gray iron with a magnesium-bearing inoculant. Sulphur content should be below 0.06 per cent for economical treatment and enough inoculant should be used to reduce sulphur below about 0.015 per cent and leave from 0.04 to 0.09 per cent residual magnesium in the iron. Specifically the technique which proved satisfactory at M.I.T. was as follows:

- (1) Superheat the iron to about 2850 F and hold 20 min, cool to 2750 F before tapping. (At higher temperature, this holding time is correspondingly less.)
- (2) Place the inoculant (which may be Fe-Si-Mg, Ni-Mg or Cu-Mg) on the bottom of a receiving ladle

and fill with metal.

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(3) Remove the slag which forms and collects on the surface of metal in the ladle.

(4) Pour the metal into a second ladle and add a post-inoculant (which may be 0.50-0.75 per cent silicon as 75 per cent ferrosilicon or 85 per cent ferrosilicon, and should contain about 0.06 per cent calcium-silicide) to the molten stream.

(5) This metal is now ready for use and may be poured into typical molds of green or dry sand.

The above method is more complicated than has been found necessary in practice where reladling is not essential. However, this method is excellent for laboratory use and gives dependable results.

(6) The chief requirements for melting are that a satisfactory base iron be used and that the metal be superheated to a high enough temperature and held at temperature long enough for complete solution of flake forming nuclei. 2850 F for 20 min proved a satisfactory combination of temperature and time. It is well known that metal melted in any type of furnace is satisfactory as a base for making spherulitic iron; the particular economic requirement in this respect is that sulphur be less than 0.06 per cent before treatment; this suggests a basic-lined cupola as the most economical melting unit.

#### VII. Conclusion

The authors wish to express their appreciation to the American Brake Shoe Co., the Electro Metallurgical Division of Union Carbide and Carbon, and the International Nickel Co. for their interest and cooperation. The authors are also grateful to the Watertown Arsenal for their sponsorship of this paper, and for permission to publish this material.

#### Bibliography

1. H. Morrogh, "Nodular Graphite Structures Produced in Gray Iron," A.F.S. Transactions, vol. 56, pp. 72-90 (1948).

2. T. H. Wickenden, "Nodular Graphite in Cast Iron," A.F.S.

Transactions, vol. 56, p. 88 (1948).

3. Vennerholm, Bogart and Melmoth, "Nodular Cast Iron," S.A.E. Quarterly Trans., vol. 4, no. 3, July 1950, pp. 422-437.

C. K. Donoho, "Producing Nodular Graphite with Magnesium," AMERICAN FOUNDRYMAN, Feb. 1949.

5. D. J. Reese, C. K. Donoho, G. Vennerholm, and R. G. Mc-Elwee, "Symposium on Nodular Graphite Cast Iron," A.F.S. Transactions, vol. 57, pp. 576-587 (1949).

 J. E. Rehder, "Effect of Phosphorus Content on Mechanical Properties of Nodular Cast Iron," A.F.S. Transactions, vol. 59,

pp. 501-508 (1951).

7. F. B. Rote, Discussion of "Influence of Silicon Content On Mechanical and High-Temperature Properties of Nodular Cast Iron," A.F.S. Transactions, vol. 59, p. 345 (1951).

8. Taylor, Rominski and Briggs, "The Fluidity of Ingot Iron and Carbon and Alloy Cast Steels," A.F.A. TRANSACTIONS, vol. 49, (1941).

9. Howard and Cohen, "Quantitative Metallography by Pointcounting and Lineal Analysis," AIME Tech. Publication No. 2215, Classes C and E, Metals Technology, August 1947.

10. "U. S. Air Forces Machinability Report, 1950," published by the Curtiss-Wright Corp.

#### APPENDIX

#### 1. Chemical Determinations

Carbon, silicon, manganese, phosphorus, sulphur, and nickel were analyzed by Dr. Walter Saunders,

Providence, R. I. Of these, only carbon requires special care if white iron samples are not available. To obtain the most representative sample, a single fractured piece of spherulitic iron should be analyzed. Carbon determinations for this investigation were taken from white iron samples. Magnesium was determined spectrographically by the Chicago Spectro Service Laboratory, Chicago.

#### 2. Mechanical Testing

Mechanical tests were conducted on a 200,000-lb Tate-Emery hydraulic testing machine using the 50,000-lb full-scale range. Standard 0.505-in. diam test bars were tested according to the ASTM procedure. Yield strengths were determined by Berry strain gage and divider method and both methods were found to be in close agreement. Therefore, unless otherwise stated, the divider method was used to determine the yield strength.

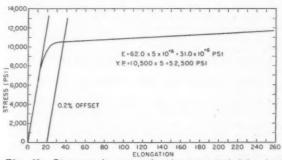


Fig. 42—Stress-strain curve for heat treated 2-in, keel block.

Figure 42 is a stress-strain curve of a typical spherulitic iron. The visual appearance of the tensile bar will be unchanged until the yield point is reached; thereafter, the entire gage length will show signs of working and will reduce uniformly, in contrast to the highly localized reduction shown by a steel tensile bar. The fracture may be (1) light, which is caused by the crystals failing by cleavage, (2) dark, which is caused by the crystals failing by shear, or (3) a combination of light and dark which is the usual condition.

#### 3. Machinability

An extensive and quantitative research on machinability has been done by Metcut Research Associates for the U.S. Air Force. The results show, for a gray cast iron and a spherulitic cast iron of the same hardness in the as-cast condition, a quantity of metal can be machined from spherulitic cast iron one-third faster than from the gray cast iron. The spherulitic cast iron had tensile strength of 84,700 psi and elongation of 17.5 per cent, while the gray cast iron had tensile strength of 35,000 psi and 0 pct elongation.

The machining of our 0.505-in. tensile bars was done on a 13-in. lathe using carbide-tipped tool without lubrication. It was possible to take a single roughing cut 0.007 in. per revolution to reduce the 1-in. by 1 in. original cross-section to a cross-section 3/4 in. in diameter.

#### Bibliography - Nodular Iron PART I*

1. A. Thum and H. Ude, "The Elasticity and Repeated Bending Strength of Cast Iron," Die Giesserei, vol. 16, 1929, pp. 501-547.

2. Meehanite Metal Corp., British Patents No. 292,164, August 22, 1929; 463,145, March 23, 1937.

3. O. von Keil, "Graphitization in Cast Iron," Archiv Eisenhuttenwesen, vol. 4, 1930, p. 245.

4. A. F. Meehan, British Patent 210,118, Jan. 28, 1944.

5. A. F. Meehan and Meehanite Corp., U.S. Patent 1,790,552, Jan. 27, 1931.

6. H. W. Gillett, "Heredity in Cast Iron," Materials & Alloys, vol. 5, 1934, p. 188.

7. A. von Keil, R. Mitsch, A. Gogat, and H. Trenklor, "On Magnesium Additions for Cast Iron," Archiv Eisenhuttenwesen, vol. 7, 1934, p. 579.

8. H. Hanemann and A. Schrader, "Nodular Graphite in Cast Iron," Atlas Metallographicus, Berlin, 1936.

9. H. Nipper, "Nodular Graphite in Cast Iron," Giesserei, vol. 22, 1936, p. 27.

10. H. Grober and H. Hanemann, "Nodular Graphite in Cast Iron," Arch. f.d. Eisenhuttenwesen, vol. 11, 1937, p. 199.

11. A. J. Krynitsky and C. M. Saeger, Journal of Research of National Bureau of Standards, 1942, vol. 28, Jan., pp. 73-94.

12. E. Piwowarsky, "Nodular Graphite in Cast Iron," Hochwertiges Gusseisen, Berlin, 1942, p. 245.

13. Linde Air Products Co., British Patent No. 622,419, March 21, 1947.

14. Anon., "Nodular Cast Iron," Machinery, vol. 70, April 1947, p. 420.

15. H. Morrogh and W. J. Williams, "Graphite Formation in Gray Cast Iron," Proceedings, Inst. of British Foundrymen, vol. 40, paper no. 875, 1947. Foundry Trade Journal, vol. 82, August 21, 1947, pp. 359, 363; August 28, 1947, pp. 391-6; Sept. 4, 1947. Engineering, vol. 164, Aug. 8, 1947, pp. 141-143; Aug. 15, 1947,

16. Anon., "Report of Annual General Meeting, British Cast Iron Research Association," Foundry Trade Journal, Dec. 11, 1947, p. 311.

17. J. G. Pearce, "Acicular Cast Irons," Engineering, vol. 164, Dec. 19, 1947, p. 596; Dec. 26, 1947, pp. 607-8.

18. H. Morrogh and W. J. Williams, "Graphite Formation in Cast Iron and in Nickel Carbon-Cobalt Carbon Alloys," Journal, Iron and Steel Inst., vol. 155, 1947, pp. 321-371; vol. 156, 1947, pp. 491-496; vol. 157, 1947, pp. 193-197.

19. H. Morrogh, "Nodular Graphite Structures Produced in Gray Cast Irons," AMERICAN FOUNDRYMAN, April 1948, pp. 91-106.

MOTE: 1) STEEL FOUNDERS SOCIETY MANDBOOK - MOST PROBABLE VALUES FOR MEDIUM CARBON STEELS - PP 242-243.
2) AVERAGE VALUE OF 2.5 TO 40 PERCENT CARBON NODULAR IRONS - FIGURE 17,
3) MALLEABLE FOUNDERS SOCIETY MANDBOOK - AVERAGE OF CLASS I AND CLASS : MOST PROBABLE VALUE - P 42. ELASTICITY FOR SECTIONS 1/2" TO 4 1/2",

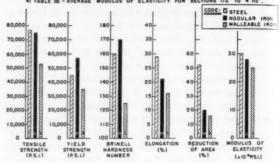


Fig. 43—Comparison of the mechanical properties of ferritic steel, spherulitic iron, and malleable iron.

20. J. G. Pearce, "Cast Iron, Nodular Graphite Structure," Soc. Chem. Ind. (London), Chem. Engrg. Group Adv. Copy, April 6, 1948.

21. A. Meyer and W. Westwood, "Cerium in Cast Iron," Foundry Trade Journal, April 15, 1948, p. 364.

22. E. K. Smith, "Nodular Graphite in Cast Iron," Iron Age, vol. 163, April 28, 1948, p. 84.

23. J. G. Pearce, "A New Cast Iron: I-Characteristics of the Nodular Graphite Structure; II-Examples of Chemical and Mechanical Properties," Chemical Age, vol. 58, May 1, 1948, pp. 616-18; June 5, 1948, pp. 607-8. 24. Anon., "Nodular Cast Iron," Foundry Trade Journal,

May 13, 1948, p. 463.

25. H. Morrogh and W. J. Williams, "Production of Nodular Graphite Structures in Cast Iron," Engineering, vol. 185, May 21, 1948, p. 494; Engineering, vol. 165, May 14, 1948, p. 475; Metallurgia, vol. 38, June 1948, pp. 98-100.

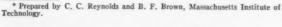
26. E. Piwowarsky, "Nodular Cast Iron," Die Neue Giesserei, vol. 1, July 1, 1948, p. 2.

27. H. Morrogh and J. W. Grant, "Nodular Cast Irons, Their Production and Properties," *Proceedings*, Inst. of British Foundrymen, vol. 41, A-29-53, 1947-8, paper no. 898, Foundry Trade Journal, July 8, 1948, pp. 27-34; July 15, 1948, pp. 51-57; July 22, 1948, pp. 81-86; July 29, 1948, pp. 105, 113. Canadian Metals and Metal Industries, vol. 11, 1948, pp. 18-22, 32.

28. J. G. Pearce, British Cast Iron Research Assoc., Metallurgia, vol. 38, Sept. 1948, pp. 273-5.

29. C. Adey, "On Nodular Cast Iron," Die Neue Giesserei, vol. 1, Sept. 1948, pp. 67-74.

30. Anon., "Properties of Nodular Cast Iron-A Bibliography,"



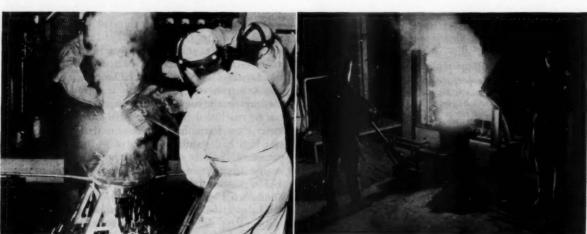


Fig. 44—Reaction of Fe-Si-Mg alloy with 70 lb and 300 lb of metal.

Iron Age, vol. 162, No. 25, 1948, p. 83. Bulletin, British Cast Iron Research Assoc., July 1948.

31. H. Morrogh and W. J. Williams, "Nodular Graphite, Production in Cast Iron," Iron and Steel (London), vol. 21, 1948,

pp. 208, 281; discussion pp. 266-8. 32. H. Morrogh and W. J. Williams, "The Production of Nodular Graphite Structures in Cast Iron," Journal, Iron and Steel Inst., vol. 158, 1948, p. 306; discussion, 160, 1948, p. 21.
 33. T. H. Wickenden, "Nodular Graphite in Cast Iron,"

TRANSACTIONS, American Foundrymen's Society, vol. 56, p. 88

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i,

- 34. Anon., "Magnesium Treatment for Nodulizing Graphite in Cast Iron," Dow Chemical Co., Bulletin No. D.M. 109, Jan.
- 35. J. E. Hurst, "Iron Founding and the Metallurgy of Cast
- Iron," Metallurgia, vol. 39, Jan. 1949, p. 129.

  36. Albert De Sy, "Belgian Research Advances Nodular Graphite Theory," AMERICAN FOUNDRYMAN, vol. 15, Jan. 1949,
- 37. M. M. Hallett, "Nodular Cast Iron," Ship and Boat Builder,
- vol. 2, Jan. 1949, pp. 233-241. 38. Albert De Sy, "Reports Latest Nodular Graphite Work," AMERICAN FOUNDRYMAN, vol. 15, Feb. 1949, p. 60.
- 39. C. K. Donoho, "Producing Nodular Graphite with Magnesium," American Foundryman, vol. 15, Feb. 1949, p. 30.
- 40. Anon., "Recent Developments in Cast Iron in the U.S.A.," Foundry Trade Journal, Feb. 10, 1949.
- 41. A. P. Gagnebin, K. D. Millis and N. B. Pilling, "Ductile Cast Iron: A New Engineering Material," Iron Age, Feb. 17, 1949.
- 42. Anon., "Magnesium Treatment for Nodular Graphite Cast Iron," Iron Age, Feb. 24, 1949, pp. 97-9
- 43. C. K. Donoho, "Magnesium Treatment for Nodular Graphite Cast Irons," Iron Age, Feb. 24, 1949, vol. 163, no. 8,
- 44. Anon., "New Ductile Cast Iron Promises to Fill Numerous Needs," Nickel Topics, March 1949, vol. 2, p. 1.
- 45. Anon., "Nodular Graphite Iron Should Widen Markets," The Foundry, vol. 77, March 1949, p. 67.
- 46. R. Bertschinger, "Nodular Graphite in Gray Iron," Schweizer Archiv fur Angewandte Wissenschaft und Technik,
- vol. 15, March 1949, pp. 75-84.
  47. H. W. Lownie, Jr., "Nodular Graphite in Cast Irons,"
  Metals Review, vol. 22, March 1949, pp. 5-8.
  48. J. G. Pearce, "Nodular Cast Iron," Mechanical Engineering,
- vol. 71, March 1949, pp. 236-8. 49. Anon., "Stronger Cast Iron: Nodular Iron," Business
- Week, March 5, 1949, pp. 48-49. 50. Anon., "Describes Development of Ductile Cast Iron,"
- The Foundry, April 1949, vol. 77, no. 4, pp. 111, 114, 116. 51. Anon., "New Ductile Cast Iron Available to Industry," Marine Engineering, vol. 54, April 1949, p. 60; Chemical Engineering, vol. 56, July 1949, p. 162; Automotive Industries, vol. 100, May 1949, p. 56.
- 52. Anon., "Nodular Cast Iron Now Being Produced in This Country," Material and Methods, vol. 29, April 1949, p. 85.
- 53. Anon., "Recent Development of Ductile Cast Iron," The Foundry, vol. 77, April 1949, p. 111.
- 54. C. O. Burgess, "Nodular Gray Iron-A Progress Report,"
- Gray Iron Industry News, vol. 22, April 1949, pp. 13-14. 55. C. K. Donoho, "Use of Magnesium and Magnesium Alloys," The Foundry, vol. 77, April 1949, p. 121.
- 56. T. C. Jarrett, "High Strength Cast Irons," The Foundry,
- vol. 77, April 1949, pp. 66-73.
  57. J. E. Rehder, "The Malleable Foundryman and Nodular Cast Iron," American Foundryman, vol. 15, April 1949, p. 144.
- 58. Anon., "Malleable Founders Group to Study Nodular
- Iron," Iron Age, vol. 163, April 7, 1949, p. 149. 59. Anon., "Nodular Cast Iron," Iron Age, April 7, 1949, p.
- 60. Anon., "Nodular Graphite Cast Iron Developments Producing Rapidly," Iron Age, vol. 163, No. 14, April 7, 1949, p. 149. 61. E. Piwowarsky and R. A. Wittmoser, "Hot Worked Cast Iron," V.D.I. Zeitschrift, vol. 91, April 15, 1949, pp. 183-5; The

Engineers Digest, X, Oct. 1949, p. 348. 62. Anon., "Nodular Iron," Gray Iron Founders' Society, Tech.

Bulletin, no. 2, April 18, 1949, pp. 1-2. 63. G. E. Holdeman and J. C. H. Stearns, "Nodular Graphite

Irons," Iron Age, 163, April 28, 1949, p. 82.

64. Anon., "Nodular Graphite Cast Iron as an Engineering Material-Correlated Review," Materials and Methods, May 1949, p. 45.

65. C. O. Burgess, "Progress in Nodular Field," Gray Iron Industry News, No. 23, May 1949, p. 13.

66. C. O. Burgess, "Progress Report on Nodular Iron," The Foundry, vol. 77, May 1949, p. 112.

67. H. Morrogh and J. W. Grant, "Nodular Cast Irons," Product Engineering, May 1949, pp. 143-145.

68. Anon., "Approves New Service on Nodular Graphite Iron,"

Iron Age, vol. 163, May 5, 1949, p. 138. 69. H. Morrogh, "Some Notes on the History of Nodular

Irons," Iron Age, May 19, 1949, p. 100.
70. W. L. Nelson, "Nodular Cast Iron," Oil and Gas Journal, vol. 48, May 19, 1949, p. 365.

71. O. Smalley, "Some Notes on the History of Nodular Irons," Iron Age, May 19, 1949, p. 100.

72. Anon., "Nodular Iron Study Begun," Steel, vol. 124, May

30, 1949, pp. 34, 36. 73. Anon., "Discuss Nodular Graphite Iron," AMERICAN

FOUNDRYMAN, vol. 15, June 1949, p. 66. 74. Albert De Sy, :"Belgian Research on Nodular Cast Iron," Metal Progress, vol. 55, June 1949, p. 838.

75. W. C. Jeffery, "Cast High Elongation Iron," AMERICAN

FOUNDRYMAN, vol. 15, June 1949, p. 68. 76. C. K. Donoho, "Producing Nodular Graphite with Magnesium," Foundry Trade Journal, June 2, 1949, p. 519.

77. H. W. Gillett, "Some Notes on the History of Nodular Irons," Iron Age, June 9, 1949, p. 30.

78. Anon., "Begin Ductile Iron Use," Steel, vol. 124, June 13, 1949, p. 69.

79. H. Morrogh, "Further Notes on Nodular Iron," Iron Age,

vol. 163, June 16, 1949, p. 64. 80. Anon., "Discuss Nodular Graphite Iron," Steel, vol. 124,

June 27, 1949, p. 40. 81. C. O. Burgess, "Progress Report on Nodular Cast Iron,"

The Foundry, vol. 77, July 1949, pp. 112-115.
82. R. E. Deas and L. T. Conradi. "Determination of Total

Carbon in Pig Iron, High Carbon Iron and Nodular Cast Iron," The Foundry, vol. 77, July 1949, pp. 68-9. 83. C. K. Donoho, "Nodular Graphite Cast Iron," Pig Iron

Rough Notes, vol. 109, Summer, 1949, pp. 3-7. 84. C. K. Donoho, "Nodular Graphite Cast Iron Symposium,"

Transactions, A.F.S., vol. 57, p. 580 (1949). 85. R. G. McElwee, "Nodular Cast Iron, A Symposium,"

Transactions, A.F.S., vol. 57, p. 585 (1949). 86. C. N. Offenhauer and J. F. Collins, "The Effects of Zirconium in Cast Iron," Canadian Mining and Metallurgical Bulletin, July 1949.

87. D. J. Reese, "Symposium: Nodular Graphite Cast Iron," Transactions, A.F.S., vol. 57, p. 576 (1949).

88. G. Vennerholm, "Nodular Cast Iron, A Symposium," Transactions, A.F.S., vol. 57, p. 581 (1949).

89. O. Smalley, "Nodular Cast Iron-The Patent Situation," Iron Age, vol. 164, July 14, 1949, p. 68.

90. Anon., "New Ductile Cast Iron Developed," Skillings Mining Review, vol. 38, July 23, 1949, p. 6.

91. Anon., "The Production of Massive Castings of Ferritic Nodular Cast Iron," La Fonderie Belge, May-August 1949, pp. 31-7.

92. Anon., "List of Foundries Making Nodular Cast Iron," The Foundry, Aug. 1949, p. 156.

93. W. W. Braidwood and A. D. Busby, "Spheroidal Graphite Cast Iron," Foundry Trade Journal, vol. 87, Sept. 15, 1949; Nickel Bulletin, vol. 22, August-September 1949, pp. 126-130.

94. G. E. Holdeman and J. H. Stearns, "Variables in Producing Nodular Cast Iron," AMERICAN FOUNDRYMAN, vol. 16, Aug. 1949, pp. 36-37.

95. Anon., "Plans Making of Nodular Iron," The Foundry, vol. 77, Sept. 1949, p. 98.

96. Max Kuniansky, "Problems in Producing Ductile Iron," The Foundry, vol. 78, Jan. 1950, pp. 62-64; The Iron Worker, Fall, 1949, p. 24.

97. C. More, "Nodular Cast Iron," Metallurgie, vol. 81, Sept. 1949, pp. 15-17, 27.

98. J. G. Pearce, "British Cast Iron Research Assoc. Announcement of Developments," Metallurgie, vol. 40, Sept. 1949, pp.

266-7

99. J. G. Pearce, "Progress in the Study of Graphite Forma-Technisch-Wissenschaftliche Beihafte Metallkunde und Giesseiwesen, no. 1, Sept. 1949, pp. 17-21.

100. E. Piwowarsky, "On Cast Iron with Nodular Graphite," Technisch-Wissenschaftliche Beihefte Metallkinde und Giessei-

wesen, no. 1, Sept. 1949, pp. 23-31.

101. J. E. Rehder, "Magnesium Additions and Desulphurization of Cast Irons," AMERICAN FOUNDRYMAN, vol. 16, Sept. 1949, p. 33.

102. E. K. Smith, "Foundries Find Fertile Field in Nodular

Iron Castings," Western Industry, vol. 14, Sept. 1949, p. 38.103. Anon., "Nodular Cast Iron," Steel, Sept. 5, 1949, p. 53. 104. E. T. Myskowski and R. P. Dunphy, "Economic and Safety Advantages Seen for Improved Nodulizing," Steel, vol. 125, Sept. 5, 1949, p. 82.

105. E. T. Myskowski and R. P. Dunphy, "A Graphite Nodulizing Alloy," Iron Age, vol. 164, Sept. 8, 1949, pp. 78-9.

106. Anon., "Are You, Producing Nodular Cast Iron?" Steel, Sept. 12, 1949, p. 152.

107. Anon., "Nodulizing Gray Iron with Dilute Magnesium

Alloy," Iron Age, Sept. 29, 1949, p. 65.
108. Anon., "The Materials Outlook," Materials & Methods, Oct. 1949, p. 3.

109. Wm. W. Austin, Jr., "What's in a Name?", AMERICAN FOUNDRYMAN, vol. 16, Oct. 1949, p. 38.

110. Albert De Sy, "What's in a Name?", AMERICAN FOUNDRY-

MAN, vol. 16, Oct. 1949, p. 35. 111. C. K. Donoho, "Nodular Graphite; Producing Spheroidal

Structures with Magnesium," Iron and Steel, vol. 22, Oct. 1949,

112. C. K. Donoho, "What's in a Name?", AMERICAN FOUNDRY-MAN, vol. 16, Oct. 1949, p. 36.

113. A. P. Gagnebin, "What's in a Name?", AMERICAN FOUND-

RYMAN, vol. 16, Oct. 1949, p. 37.

114. G. Halbart, "Nodular Graphite May be Due to Self-Inoculation by Magnesium Silicide," Fonderie Belge, no. 21, Sept.-Oct. 1949, pp. 5-6.

115. W. R. Kennedy, "Determination of Magnesium in Cast

Iron," The Foundry, Oct. 1949, p. 80. 116. E. O. Lissell, "What's in a Name?", AMERICAN FOUNDRY-

MAN, vol. 16, Oct. 1949, p. 36. 117. E. T. Myskowski and R. P. Dunphy, "New Graphite

Nodulizing Alloy," *The Foundry*, Oct. 1949, p. 72. 118. J. E. Rehder, "What's in a Name?", AMERICAN FOUNDRY-MAN, vol. 16, Oct. 1949, p. 34.

119. E. K. Smith, "What's in a Name?", AMERICAN FOUNDRY-

MAN, vol. 16, Oct. 1949, p. 38. 120. G. Vennerholm, "What's in a Name?", AMERICAN FOUND-

RYMAN, vol. 16, Oct. 1949, p. 35.

121. Anon., "Mechanite Tells Foundries Nodular Iron is Ready," Iron Age, Oct. 13, 1949, p. 100.

122. Anon., "Nodular Graphite Meehanite," Steel, Oct. 24,

1949, p. 46. 123. C. O. Burgess, "Advances in Gray Iron Are Discussed at ASTM Symposium," Materials & Methods, Nov. 1949, p. 45.

124. Albert De Sy, "Some Results of Belgian Research on Nodular Cast Iron," The Foundry, Nov. 1949, p. 128.

125. D. E. Krause, "What's in a Name?", AMERICAN FOUNDRY-MAN, vol. 16, Nov. 1949, p. 45.

126. D. E. Krause, "Nodular Iron," AMERICAN FOUNDRYMAN, vol. 16, Nov. 1949, p. 32.

127. R. B. Kropf, "Ductile Cast Iron," Metals Review, Nov. 1949, p. 16.

128. W. W. Levi, "What's in a Name?", AMERICAN FOUNDRY-MAN, vol. 16, Nov. 1949, p. 46.

129. R. B. Melmoth, "Nodular Cast Iron at Ford Motor Co.,"

AMERICAN FOUNDRYMAN, vol. 16, Nov. 1949, p. 32. 130. R. L. Moncrief, "Ductile Cast Iron Shows Interesting Strength Properties," Metals Review, vol. 22, Nov. 1949, p. 16. 131. H. Morrogh, "What's in a Name?", AMERICAN FOUNDRY-MAN, Nov. 1949, p. 44.

132. J. E. Rehder, "More on Magnesium Treatment of Iron," AMERICAN FOUNDRYMAN, vol. 16, Nov. 1949, p. 63.

133. J. C. H. Stearns, "What's in a Name?", AMERICAN FOUND-RYMAN, vol. 16, Nov. 1949, p. 46.

134. Anon., "Makes Ductile Iron Castings," Steel, Nov. 28, 1949, p. 28.

135. Anon., "Nodular Graphite Cast Iron-How Will it Affect Australian Foundrymen?", Australian Foundry Trade Journal, vol. 1, Dec. 1949, pp. 4-9.

136. Anon., "Spheroidal Graphite Cast Iron Patents in the U.S. and Great Britain," Metallurgia, vol. 41, Dec. 1949, p. 102.

137. F. R. Bryan, G. A. Nahstoll, and H. D. Veldhirs, "Spectrographic Methods for Determining Magnesium in Nodular Iron," Bulletin, ASTM, Dec. 1949, p. 69.

138. A. B. Everest, "Progress in High Duty and Alloy Cast Iron," Metallurgia, vol. 41, Dec. 1949, p. 84.

139. C. L. Rohrbaugh, "Ductile Iron Comparison Shows How Properties and Structure Are Altered," Metals Review, Dec. 1949,

vol. 22, no. 12, p. 11. 140. E. K. Smith, "Experiments in Nodular Iron," AMERICAN FOUNDRYMAN, vol. 16, Dec. 1949, pp. 45-47.

141. J. C. H. Stearns, "Need More Reliable Formulas," AMERI-CAN FOUNDRYMAN, vol. 16, Dec. 1949, p. 64.

142. C. R. Austin, "Some Engineering Aspects of Nodular Cast Iron," Iron Age, Dec. 1, 1949, p. 79.

143. T. E. Eagan and J. D. James, "A Practical Evaluation of Ductile Cast Iron," Iron Age, vol. 164, Dec. 8, 1949, pp. 75-79; Dec. 15, 1949, pp. 77-82.

144. J. T. Tozsa, "Spectrographic Analysis of Ductile Cast Iron," Iron Age, Dec. 22, 1949, p. 73.

145. Anon., "Symposium on Nodular Graphite Cast Iron," Transactions, American Foundrymen's Society, vol. 57, p. 576 (1949).

146. Albert De Sy, "Producing As Cast-Ferritic Nodular Iron in Heavy Sections," Metal Progress, vol. 57, no. 1, Jan. 1950, p.

147. C. K. Donoho, "Nodular Graphite Cast Iron," Gray Iron Inst. News, Jan. 1950, p. 15. 148. M. G. Fontana, "Corrosion," Industrial and Engineering

Chemistry, vol. 42, Jan. 1950, pp. 67A-68A.

149. A. P. Gagnebin, K. D. Millis, and N. B. Pilling, "Engineering Applications of Ductile Cast Iron," Machine Design, vol. 22, Jan. 1950, pp. 108-114.

150. J. B. Guillaumin, "Spheroidal Graphite Cast Iron," Métaux et Industries, no. 1, 1950, pp. 11-13.

151. W. C. Jeffery, "More Names for New Iron," AMERICAN FOUNDRYMAN, Jan. 1950, p. 74.

152. G. Joly, "More Names for New Iron," AMERICAN FOUNDRY-MAN, Jan. 1950, p. 74.

153. A. E. LaRochelle and G. A. Fournier, "Magnesium in Iron Determined by Mercury Cathode Method," FOUNDRYMAN, Jan. 1950, p. 65. 154. R. W. Lindsay, "More Names for New Iron," American

FOUNDRYMAN, Jan. 1950, p. 74.

155. H. W. Maynor, "More Names for New Iron," AMERICAN FOUNDRYMAN, Jan. 1950, p. 74.

156. D. J. Reese, "Factors Affecting Development of Ductile Iron." The Foundry, vol. 78, Jan. 1950, pp. 58-62, 216-217.

157. J. F. Rehder, "Does Nodular Iron Resist Zinc?", AMERI-CAN FOUNDRYMAN, Jan. 1950, p. 62.

158. R. Schneidewind, "New Names for New Iron," AMERICAN FOUNDRYMAN, Jan. 1950, p. 93.

159. R. L. Collier, "Popularity of Ductile Irons will Expand Rapidly," Steel, Jan. 2, 1950, p. 173. 160. H. A. Schwartz, "Caution Urged in Using Nodular Cast

Iron," Steel, Jan. 2, 1950, p. 172.

161. E. B. Sherwin, "Grey Iron Still Very Useful Material," Steel, Jan. 2, 1950, p. 172.

162. F. H. Allison, Jr., "Spheroidal Iron—A New Engineering Material," *United Effort*, Jan.-Feb. 1950.

163. U. von Hummel and E. Piwowarsky, "The Influence of Graphite, Especially Shape on the Modulus of Elasticity and the Displacement of Neutral Phases by Transverse Bending Tests," Technische Wissenschaftliche Beihefte, no. 2, Feb. 1950.

164. R. G. McElwee, "Nodular Cast Iron Is Newest Star on Foundry Horizon," Metals Review, Feb. 1950.

165. H. Morrogh, "Grey Cast Iron with Nodular Graphite," Technische Wissenschaftliche Beinhefte, no. 2, Feb. 1950, pp. 65-74.

166. H. Morrogh, "Nodular Graphite Structures," British Cast Iron Research Assoc., Journal of Research and Development, vol. 3, Feb. 1950, pp. 251-298.

167. N. Nacken and E. Piwowarsky, "On the Structure of Spherulitic Graphite," Technische Wissenschaftliche Beichefte, no. 2, Feb. 1950, pp. 75-76.

168. Anon., "Chemical Analysis: Nodules in Nodular Cast Iron," Quarterly Report, Canada, Dept. of Mines and Technical

Surveys, Jan.-March 1950, pp. 8-9.
169. M. Hatherly and L. E. Samuels, "Radial Structure of Graphite Nodules," Metal Progress, vol. 57, no. 3, March 1950,

170. V. I. Laverty, "Meehanite Nodular Iron Castings," Cana-

dian Metals, vol. 13, no. 3, March 1950, pp. 30-31.
171. F. R. Morral, "Nodular Iron," The Foundry, vol. 78, no.

3, March 1950, p. 135.

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172. H. Thyssen, P. Coheur, and L. Habraken, "Examination of Nodular Iron with the Electron Microscope," Revue Universelle des Mines de la Metallurgie, March 1950, p. 62.

173. H. Thyssen and F. Gaty, "Production of Nodular Iron with Mg or Ce with Inoculation," Revue Universelle des Mines

de la Metallurgie, March 1950, p. 57.

- 174. A. Portevin, R. Cabarat, and L. Guillet, "Influence of Graphite Shape on Internal Friction of Cast Iron," Comptes Rendus de l' Acadime des Sciences, vol. 230, March 6, 1950, pp.
- 175. G. Vennerholm, H. Bogart, and R. Melmoth, "Nodular Cast Iron," Foundry Trade Journal, vol. 88, March 9, 1950, pp. 247-56, no. 1749.
- 176. O. Carlsson, "Nodular Cast Iron," Foundry Trade Jour-
- nal, March 23, 1950, vol. 88, no. 1751, p. 317 (Abstract). 177. H. Morrogh, "Nodular Cast Iron," Foundry Trade Jour-

nal, vol, 88, March 23, 1950, p. 321. 178. C. K. Donoho, "Nodular Graphite Cast Iron," Product

Engineering, vol. 21, April 1950, pp. 140-144.

179. J. W. Grant, "Notched and Unnotched Fatigue Tests on Flake and Nodular Cast Irons," British Cast Iron Research Assoc., Journal of Research and Development, vol. 3, April 1950, рр. 333-354.

180. G. Vennerholm, H. Bogart, and R. Melmoth, "Nodular Gray Iron Compared with Other Cast Ferrous Metals," Mate-

rials & Methods, April 1950, vol. 31, no. 4, pp. 51-55.
181. Tsung Ko, "Growth of Nodular Cast Iron," Foundry

Trade Journal, April 20, 1950, p. 433, vol. 88, no. 1755.
182. T. A. Heppenstall, "High Duty Cast Irons," Chemistry

and Industry, vol. 16, April 22, 1950, pp. 312-313.

183. A. L. De Sy, "Further Results of Belgian Nodular Cast Iron Research," AMERICAN FOUNDRYMAN, vol. 17, no. 5, May 1950, pp. 75-83.

184. J. E. Rehder, "Nodular Cast Iron," Canadian Metals, vol. 13, no. 5, May 1950, pp. 16-20, 46; vol. 13, no. 6, June 1950.

pp. 16-18, 21, 46-47.

185. D. J. Reese, "Producing Magnesium-Contained Iron," The Foundry, vol. 78, no. 5, May 1950, p. 120.

186. A. P. Gagnebin, "INCO Writes Specs. for Ductile Iron,"

Iron Age, vol. 165, May 4, 1950, p. 89.

187. F. R. Bryan, G. A. Bahstoll and H. P. Veldhuis, "Determination of Magnesium in Nodular Iron," The Foundry, vol. 78, no. 6, June 1950, pp. 102, 244-249.

188. A. L. De Sy, "The Core of Graphite Spherules in Nodular Cast Iron," Metal Progress, vol. 57, no. 6, June 1950, pp. 774-775.

189. C. K. Donoho, "Mechanical Testing and Properties of Gray Iron," The Foundry, vol. 78, no. 6, June 1950, pp. 96-99,

190. W. B. Sohns and J. L. Yarne, "Magnesium Determination in Nodular Iron," American Foundryman, vol. 17, no. 6, June 1950, pp. 33-35.

191. Anon., "Revolution in Casting," Modern Industry, vol.

19, June 15, 1950, pp. 38-43.

192. Anon., "Current Applications of Ductile Iron Castings," Nickel Topics, vol. 3, no. 7, July 1950.
193. F. R. Morral, "Nodular Iron—A Bibliography," AMERICAN

FOUNDRYMAN, vol. 18, July 1950, pp. 61-64.

194. E. M. State and B. L. Stott, "Some Experiences In Producing Ductile Iron," The Foundry, vol. 78, no. 7, July 1950, pp.

195. P. A. Russell, "Where is Cast Iron Going To?", Foundry Trade Journal, vol. 88, no. 1765, June 29, 1950, pp. 697-702; vol. 89, no. 1766, July 6, 1950, pp. 7-10.

196. Anon., "Nodular Iron's Record-Two Years Later," Steel, July 17, 1950.

197. A. B. Everest, "Spheroidal-graphite Cast Iron-Some Properties and Applications," Foundry Trade Journal, vol. 89, no. 1768, July 20, 1950, pp. 57-64; vol. 89, no. 1769, July 27, 1950, pp. 95-102. 198. C. D. Galloway III, "Ductile Iron for Heavy Machinery,"

Iron Age, vol. 166, no. 5, Aug. 3, 1950, pp. 75-78.

199. G. N. Gilbert, "Cast Iron at Sub-atmospheric Temperature," Foundry Trade Journal, vol. 89, no. 1771, Aug. 10, 1950; vol. 89, no. 1772, Aug. 17, 1950, pp. 179-189. 200. E. S. Laufer, "Ductile Iron in Light Sections," Iron Age,

vol. 166, no. 6, Aug. 10, 1950, p. 79. 201. A. L. De Sy, "Nodular Cast Iron Produced with Li, Ca, Ba, Sr, Na," Metal Progress, vol. 58, no. 3, Sept. 1950, p. 357.

202. A. B. Everest, "Some Properties and Application of Spheroidal Graphite Cast Iron," Inst. of British Foundrymen, Advance Copy, 1950, Paper no. 963, 11 pp.

203. R. A. Flinn and R. W. Kraft, "Improved Test Bars for Standard and Ductile Grades of Cast Iron," Transactions, A.F.S., vol. 58, p. 153 (1950).

204. J. E. Rehder, "An Introduction to the Annealing of Nodular Iron," Transactions, American Foundrymen's Society, vol. 58, p. 298 (1950).

205. N. A. Ziegler, W. L. Meinhart and J. R. Goldsmith, "Dilatometer Studies of Nodular Cast Iron," Transactions, American

Foundrymen's Society, vol. 58, p. 687 (1950). 206. Cerium Metals Corp., "The Production of Nodular Cast Irons with Cerium-Some Practical Details," 153 Waverly Place, New York 14, New York (1950).

207. A. F. Meehan, "Cast Iron and Making Same," U.S. Patent no. 1,683,087, Sept. 4, 1928.

208. A. F. Meehan, "Cast Iron and Making Same," U.S. Patent no. 1,683,086, Sept. 4, 1928.

209. A. F. Meehan, "Heat Treating Iron," U.S. Patent no. 1,731,346, Oct. 15, 1929.

210. F. Bernauer, "Gedrillte," Kristalle, Berlin, 1929.

211. British Patent no. 463,145, March 23, 1937.

212. W. Baukloh and H. Meierling, "The Effect of Cerium on Cast Iron," Die Giesserei, vol. 27, 1940.

213. O. Smalley, "Method of Manufacturing Cast Iron," U.S. Patent no. 2,364,922, Dec. 12, 1944.

214. H. Morrogh, "Nodular Cast Irons," Journal of Res. and Dev., B.C.I.R.A., vol. 2, pp. 107-140, 1947.

215. E. C. Dickinson, "Recent Technical Developments in the Production of Iron Castings," Foundry Trade Journal, vol. 84, April 8, 1948, pp. 341-344; April 15, 1948, pp. 369-370.

216. G. R. Woodward, "Production and Properties of Nodular Cast Iron (Nodulite): A Review of the Literature," Bulletin, British Cast Iron Res. Assoc., no. 7, July 1948, pp. 200-204.

217. T. Tyberg, "Cast Iron with Nodular Graphite," Gieteriet,

vol. 38, Aug. 1948, pp. 114-117; disc. pp. 118-119. 218. G. Moseley, "Is Nodular Cast Iron New?", Iron and Steel, 21 Nov. 1948, pp. 470-1.

219. Anon., "Grey Cast Iron," Canada, Dept. of Mines and Resources, Quarterly Report, 1948, Oct.-Dec., p. 13.

220. O. Smalley, "Nodular Cast Iron," Iron Age, vol. 162, Dec. 30, 1948, pp. 18-19.

221. A. De Sy, "Inoculation and Graphite in Grey Irons," Fonderie, no. 37, Jan. 1949, pp. 1433-1446.

222. M. Singh, India Patent no. 40,209, June 15, 1949.

223. B.C.I.R.A. and H. Morrogh, "Improvements Relating to Cast Iron," British Patent no. 626,178, July 11, 1949.

224. Dayton Malleable Iron Co., "Improvements in or Relating to Cast Iron and Method of Producing Same," British Patent Application no. 19744/49, July 27, 1949.

225. C. O. Burgess, "Suitable Raw Materials for Production of Nodular Irons," Gray Iron Industry News, no. 27, Sept. 1949, pp. 7-8; no. 28, Oct. 1949, p. 10.

226. Anon., "Grey Irons with Nodular Graphite Obtained by the Addition of Magnesium," Fonderie, no. 46, Oct. 1949, pp.

227. N. Hajto, "Nodular Graphite in Cast Iron," Banyaszati es Kohaszati Lapok, Oct. 1949, vol. 4, pp. 425-433. 228. Mond Nickel Co., "Improvements Relating to Cast Iron,"

British Patent no. 630,070, Oct. 5, 1949.

229. Mond Nickel Co., "Improvements Relating to Alloys," British Patent no. 630,099, Oct. 5, 1949.

230. Anon., "Nodular Cast Iron," Fabrimetal, Oct. 10, 1949, p. 801.

231. K. D. Millis, A. P. Gagnebin and N. B. Pilling, "Cast Ferrous Alloy," U.S. Patent no. 2,485,760, Oct. 25, 1949.

232. K. D. Millis, A. P. Gagnebin and N. B. Pilling, "Grey Cast Iron Having Improved Properties," U.S. Patent no. 2,485,-761, Oct. 25, 1949.

233. B.C.I.R.A. and H. Morrogh, "Nodular Cast Iron and the Manufacture Thereof," U.S. Patent no. 2,888,511, Nov. 15, 1949. 234. B.C.J.R.A. and H. Morrogh, "Cast Iron," U.S. Patent no. 2,488,512, Nov. 15, 1949.

235. B.C.I.R.A. and H. Morrogh, "Production of White Cast

Iron," U.S. Patent no. 2,488,513, Nov. 15, 1949.

236. Anon., "Ductile Iron Represents Century's Greatest Advance in Cast Iron Field," Nickel Topics, vol. 2, no. 11, Nov.-Dec. 1949, pp. 1, 9.

237. J. E. Fifield, "Ductile Iron Said Not Competitive to Grey Iron," Metals Review, no. 22, Dec. 1949, p. 17.

238. H. Pieper, "Nodular Iron," Die Neue Giesserei, vol. 36,

1949, pp. 293-6. 239. Anon., "Ductile Cast Iron Labeled Major Foundry Ad-

vance," Metals Review, Jan. 1950, p. 8. 240. C. H. Gerould, "Electron Metallography of Cast Irons."

Journal of Applied Physics, vol. 21, Jan. 1950, pp. 68-69. 241. Anon., "Chemical Analysis: Nodules in Nodular Cast Iron," Quarterly Report, Canadian Dept. of Mines and Tech.

Surveys, Jan.-March 1950, pp. 8-9. 242. M. M. Hallett, "Practical Experiences in Producing Nodu-

lar Cast Iron," Foundry Trade Journal, Jan. 4, 1950, vol. 90, pp. 3-9; disc. pp. 9-12, 16.

243. Anon., "Risering Nodular Iron," AMERICAN FOUNDRYMAN.

vol. 17, Feb. 1950, p. 59. 244. C. O. Burgess, "Foundry Facts for the Operating Executive-1: British Foundry Practice, 2: Nodular Iron Development," Gray Iron Industry News, no. 32, Feb. 1950, pp. 7-8.

245. N. Chvorinov, "Controlling the Shape of Graphite in Cast Irons," Supplement to Hutnicke Listy, no. 2, 1950, pp.

246. I. litaka, "A Theory Globular Graphite Formation in Cast Iron," Journal of the Casting Inst. of Japan, vol. 22, no. 2, 1950, pp. 1-5.

247. J. Plachy and J. Chenicek, "Studying Nodular Cast Iron," Supplement to Hutnicke Listy, no. 2, 1950, pp. 44-49.

248. F. D. Widner, "Nodular Cast Iron is Newest Star in Foundry Horizon," Metals Review, vol. 26, Feb. 1950, p. 7. 249. H. Morrogh, "Modern Metallurgy of Cast Iron," Engi-

neering, 169, Feb. 3, 1950, pp. 136-138; Feb. 10, 1950, p. 166. 250. G. L. Cox, "Uses of Ductile Iron Range from Plowshares to Gears and Crankshafts," Metals Review, 23, March 1950, p. 16. 251. D. E. Krause, "Nodulation Process for Cast Iron," Metals

Review, 23, March 1950, p. 10.

252. B. M. Thall and B. Chalmers, "Discussion on Nodular Iron," Journal, British Inst. of Metals, 71, March 1950, pp. 96-97.253. P. P. Petrosyan and S. E. Shou-Shakhbudagyan, "Influ-

ence of Certain Elements in the Formation of Nodular Graphite in Gray Cast Iron," Doklady Akademii Nauk, USSR, 71, March

1, 1950, pp. 123-125.

254. J. W. Grant, "A Note on the Hot Rolling of Cast Iron and an Examination of Some Rolled Cast Iron Samples of German Origin," B.C.I.R.A. Journal of Res. and Dev., April 1950, pp. 371-375.

255. N. Hajtó, Bavyászati es Kohaszati Lapok, 1950, vol. 5,

April, pp. 268-279.

256. A. De Sy, "Production and Properties of Nodular Cast Iron." Report of Researches under the Auspices of IRSIA and CRIF. Societe Royale Belge des Ingenieurs et des Industriels, Bulletin Series A, April 1950, no. 1, pp. 3-9.

257. Anon, "Chanbersburg Engineering Develops New Ductile

Cast Iron," Steel, 126, April 10, 1950, p. 86.

258. Anon., "Ductile Cast Iron," Metallurgia, May 1950, p. 368. 259. C. O. Burgess, "Foundry Control is Greatest Problem in Nodular Iron," Metals Review, 23, May 1950, p. 14. 260. J. Plachy and j. Chenicek, "Research on Nodular Cast

Iron," Hutnicke Listy, vol. 5, 1950, pp. 44-49.

261. G. Vennerholm, H. N. Bogart and R. B. Melmoth, "Nodular Iron Finds Favor with Ford," SAE Journal, 58, May 1950, pp. 31-34.

262. N. Chvorinov, "On Controlling the Shape of Graphite in Cast Iron," Hutnicke Listy, vol. 5, June 1950, Supplement no. 2,

263. M. M. Hallett, "Modern Cast Iron," Journal of Birmingham Metallurgical Society, vol. 30, June 1950, pp. 42-62.

264. I. Iitaka, "The Effect of Various Elements Upon the Form of Graphite in Cast Iron," Bulletin of the Inst. of Physical and Chemical Res., vol. 22, no. 6, pp. 571-576, 1950.

265. W. L. Larson, "Formation of Spherulitic Graphite in

Cast Irons," Doctor's Thesis, M.I.T., June 1950. 266. E. Piwowarsky, K. Kramer, W. Patterson and M. Krichel, "Production of Cast Iron with Nodular Graphite Formation by Treatment with Magnesium Alloys," Technisch-Wissenschaftliche Beihefte, no. 3, June 1950, pp. 91-103.

267. J. Plachy and J. Chenicek, "Investigation of Nodular Cast Iron," Hutnicke Listy, vol. 5, June 1950, Supplement no. 2,

pp. 44-49.

268. K. D. Millis, U.S. Patent no. 2,516,524, July 25, 1950. 269. H. F. Taylor, "Industrial Place for Ductile Iron," Metals

Review, vol. 23, July 1950, p. 11.

270. Vennerholm, Bogart and Melmoth, "Nodular Cast Iron," SAE Quarterly Trans., vol. 4, no. 3, July 1950, pp. 422-437. 271. A. G. Zima, "A Cast Iron That Bends," Western Machin-

ery and Steel World, vol. 41, July 1950, pp. 74-75. 272. A. L. De Sy, "La Contribution Belge dans l'étude des fontes nodulaires," Revue Universelle des Mines et de la Metallurgie, August 1950.

273. W. Westwood and R. Presser, "The Determination of Magnesium in Cast Iron," B.C.I.R.A. of R.&D., vol. 3, Aug.

1950, pp. 515-519.

274. J. E. Rehder, "Nodules and Nuclei, Nodular Cast Iron," Canada, Dept. of Mines and Technical Surveys, Mines Branch, Aug. 24, 1950.

#### PART II*

1. Anon., "On the Corrosion Resistant Properties of Ferritic and Pearlitic Ductile Iron," Nickel Topics, Sept. 1950, p. 8.

2. C. E. Herrington, "Another View of Nodular Iron," Product Engineering, vol. 21, Sept. 1950, pp. 214, 216, 218.

3. J. E. Rehder, "Desulphurizing Gray Iron with Magnesium," AMERICAN FOUNDRYMAN, vol. 18, no. 3, Sept. 1950, pp. 43-49.

4. A. Wittmoser, "On the Present Status of Cast Iron," Stahl und Eisen, vol. 70, Sept. 14, 1950, pp. 813-828.
5. S. F. Carter and C. K. Donoho, "Experiences with the Basic

Lined Cupola in Nodular Iron Production," Pig Iron Rough Notes, no. 112, Autumn 1950, pp. 3-15.

6. R. Y. Scapple, "Nodular Cast Iron: Heat Treatment and Physical Properties," *Iron and Steel*, vol. 23, Sept.-Oct.-Nov. 1950,

pp. 379-382.

7. H. N. Acker, "Report on the All-Canadian Conference of the A.F.S.," Canadian Metals, vol. 13, no. 10, Oct. 1950, pp. 20-

8. G. L. Cox, "Progress Report on Ductile Iron Gives New Data," Metals Review, vol. 26, Oct. 1950, p. 4.

9. J. G. Hague and J. I. Shultz, "Determination of Magnesium in Cast Iron," Foundry, vol. 78, no. 10, Oct. 1950, pp. 92-93;

10. J. F. Jordan, "Method of Controlling the Residual Magnesium Content of Nodular Graphite Cast Iron," U.S. Patent no.

2,527,498, Oct. 24, 1950. 11. J. T. MacKenzie, "Ten Years of Advances in Ferrous Foundries," Metal Progress, Oct. 1950, p. 499.

12. H. F. Scobie, "MIT Nodular Iron Symposium," AMERICAN

FOUNDRYMAN, vol. 18, no. 4, Oct. 1950, pp. 46-48. 13. O. Smalley, "Method of Producing Nodular Cast Iron,"

U.S. Patent no. 2,527,037, Oct. 24, 1950.

14. B. S. Srikantiah and B. R. Nijhawan, "Nodular Cast Iron Literature Report no. 10," Journal of Scientific and Industrial Research, vol. 9, Oct. 1950, pp. 364-366.

15. Anon., "Ductile Iron Replaces Steel and High Test Gray Iron in Shafting," Nickel Topics, vol. 4, no. 11, Nov.-Dec. 1950,

17. K. L. Clark, "Applications of Ductile Iron Shows Its Adaptability," Metals Review, vol. 23, Nov. 1950, p. 11.

18. B. Dixon, "Profits from Nodular Iron," Canadian Metals, vol. 13, no. 11, Nov. 1950, pp. 14-16; 43.

19. T. E. Eagan, "Notch Sensitivity of Various Cast Materials," AMERICAN FOUNDRYMAN, vol. 18, Nov. 1950, pp. 22-4.

20. H. Morrogh, "Circular Graphite," Metal Progress, vol. 58, Nov. 1950, pp. 734-735.

^{*} Prepared by C. C. Reynolds, Massachusetts Institute of Technology.

21. E. M. Stein, "Nodular Iron Experiments," Foundry, vol. 78, no. 11, Nov. 1950, pp. 96-99; 214, 216, 217, 218, 220.
 22. Anon., "Machinability Research Program Reported," Iron

Age, Nov. 9, 1950, pp. 99-102.

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23. Anon., "Magnesium for Production of Nodular Iron," Steel, Nov. 20, 1950, p. 71.

24. Anon., "Proves Machinability Depends on Microstructure," American Machinist, vol. 94, Nov. 27, 1950, pp. 109-124. 25. A. Wittmoser, "On Ladle Additions to Cast Iron of

Magnesium-Bearing Alloys," Die neue Giesserei, vol. 37, Nov. 30, 1950, pp. 533-540.

26. T. E. Egan, "Comparative Growth of Gray Iron vs. Nodular Iron (when cyclically heated to 1650° F)", The Foundry,

vol. 78, Dec. 1950, pp. 96-99, 203-4. 27. E. Piwowarsky, "Present Status, Current Developments and Future Possibilities of Gray Cast Iron with Nodular Graph-Technisch-Wissenschaftliche Beihefte, no. 4, Dec. 1950, 175-180.

28. A: E. Egan, "Nodular Iron for Stressed Parts," Iron Age,

vol. 166, Dec. 7, 1950, p. 162.

29. J. F. Kahles and R. Goldhoff, "Anneal Ductile Iron for Better Machinability," Iron Age, vol. 166, Dec. 14, 1950, pp.

30. R. Collette and A. De Sy, "Methods of Producing Nodular Cast Iron," Foundry Trade Journal, vol. 89, no. 1789, Dec. 14, 1950, pp. 495-498.

31. T. E. Egan, "Nodular Iron for Mechanical Parts," Steel,

vol. 127, Dec. 25, 1950, p. 72.

32. C. M. Schwetter, "Two New Gear Materials," Machinery,

Dec. 1950, Jan. 1951.

33. K. P. Bunin and G. I. Ivanstov, "On the Crystallization of Cast Iron With Nodular Graphite," Doklady Akademii Nauk,

USSR, vol. 72, 1950, pp. 1051-3. 34. O. Carlsson, "Experiences with Nodular Cast Iron," Gjuteriet (Stockholm), vol. 40, 1950, pp. 31-37. Henry Brutcher

35. E. E. Cheburkova, "Determination of Magnesium in Cast

Iron," Zavodskaya Laboratoriza, vol. 16, 1950, pp. 663-666. 36. J. F. Kahles, N. Zlatin, and R. B. Kropf, "High Machinability and Productivity of Ductile Iron," Metal Progress, vol. 59, no. 2, Feb. 1951, pp. 238. Complete in "U.S. Air Forces Machinability Report, 1950," Curtiss-Wright Corp.

37. W. Patterson, "Spherulitic Cast Iron-A New Material,"

Werkstatt u. Betrieb, vol. 83, 1950, pp. 18-20.

38. H. A. Schwartz, Foundry Science, 1950, pp. 163-4, Pitman

Publishing Co., New York.

39. G. Vennerholm, H. Bogart and R. Melmoth, "Section Size Relationships in Nodular Iron," Transactions, A.F.S., vol. 58, pp. 174-183 (1950).

40. F. R. Morral, "Nodular Iron-A Bibliography," AMERICAN

FOUNDRYMAN, vol. 19, no. 1, Jan. 1951, pp. 43-44.

41. M. M. Hallett, "Practical Experiences in Producing Nodular Cast Iron," Foundry Trade Journal, vol. 90, no. 1792, Jan. 4, 1951, pp. 3-12. 42. C. M. Offenhauer, "Method of Producing Ductile Cast

Iron," U.S. Patent no. 2,538,263, Jan. 16, 1951.

43. A. Wittmoser, "Cast Iron with Nodular Graphite as an Engineering Material," Zeitschreft des Vereines Deutscher Ingenieure, vol. 93, Jan. 21, 1951, pp. 49-57.
44. Anon., "On the Production of Gray Cast Iron with Nodu-

lar Graphite by Treatment with Cerium," Zeitschrift des Vereines Deutscher Ingenieure, vol. 93, Jan. 21, 1951, pp. 65-66

45. Anon., "Ductile Iron Proves Good Pump Casing Material" and "Novel Automatic Clutch Developed," Nickel Topics, vol.

4, no. 2, 1951, p. 1 and p. 8.
46. N. Croft, "Producing Ductile Cast Iron. Some Foundry Experiences," Iron and Steel, vol. 24, Feb. 1951, pp. 45-50.

- 47. A. L. De Sy, "Eliminate Second Inoculation in New Nodular Iron Process," American Foundryman, vol. 19, no. 2, Feb. 1951, pp. 41-45.
- 48. A. L. De Sy, R. Collette and J. Vidts, "Production of Nodular Cast Iron by Calcium or by the Combined Effects of Calcium-Magnesium, Calcium-Lithium, and Calcium-Lithium-Magnesium," Fonderie Belge, Feb. 1951, pp. 34-42.
- 49. W. S. Owen and B. G. Street, "The Crystal Structure of Graphite in Cast Iron," Journal of the Iron and Steel Inst., vol. 167, Feb. 1951, pp. 113-116.
- 50. C. K. Donoho, "Letters to the Editor-Calcium Nodular Irons," AMERICAN FOUNDRYMAN, vol. 19, no. 3, March 1951, pp.

51. E. Franke, "The Corrosion Process in Ferritic and Pearlitic Cast Iron with Nodular Graphite. A Critical Survey," Werkstoffe and Korrosion, vol. 2, March 1951, pp. 101-103.

52. J. E. Rehder, "Letters to the Editor—Calcium Nodular Irons," American Foundryman, vol. 19, no. 3, March 1951, pp.

89, 91,

53. Stauss, Von Batchelder and Salkovitz, "Structure of Spherulites in Nodular Cast Iron," Journal of Metals, vol. 3, no. 3, March 1951, p. 249.

54. H. W. Gronegress, "Flame Hardening Cast Iron," Stahl

und Eisen, vol. 71, March 1, 1951, pp. 246-252. 55. J. F. Jordan, "Process for Adding Magnesium to Cast

Iron," U.S. Patent No. 2,543,853, March 6, 1951.

56. G. Derge, "Deoxidation in Production of Nodular Cast

Iron," Foundry, vol. 79, April 1951, pp. 122-123.
57. W. Westwood and R. Presser, "Chemical Determination of Magnesium in Cast Iron," Analyst, vol. 76, April 1951, pp. 191-

199; disc. pp. 199-200.58. W. W. Austin, Jr., "A Practical Appraisal of Nodular Iron and Its Testing," vol. 167, April 19, 1951, pp. 86-89,

Malleable Iron Facts, no. 40.

59. Anon., "Ductile Iron Makes Good in Sintering Grates" and "Large Ductile Iron Gear Cast Abroad," Nickel Topics, vol. 4, no. 5, 1951, p. 2.

60. A. B. Everest, "Cast Iron Crankshafts, with Special Reference to Acicular and Spheroidal Graphite Cast Irons," Congres International des Moteurs, Paris, May 1951, Group IV-1, 20 pp.

61. S. L. Gertsman and B. F. Richardson, "Magnesium Alloys Aid Desulphurization," Canadian Metals, vol. 14, no. 5, May 1951, pp. 20-23.

62. A. Guedras, "Improvement of the Characteristics of Cast Iron by Modification of the Graphite," Metallurgie, vol. 83, May

1951, pp. 351, 353, 355.63. J. E. Rehder, "Finds Magnesium will Dephosphorize Molten Iron When Added in Ladle," AMERICAN FOUNDRYMAN,

vol. 19, no. 5, 1951, p. 95. 64. E. Longden, "Notes on Cast Iron—Past and Present," Foundry Trade Journal, vol. 90, no. 1810, May 10, 1951, pp.

65. J. D. Sheley, "Ductile Iron Replaces Alloy Gear Castings, Forgings," Iron Age, vol. 167, May 10, 1951, pp. 99-100.

66. M. Grandpierre and H. De Bouvier, "Influence of Phosphorus on Ductile Iron," Master's Thesis, M.I.T., May 18, 1951. 67. Anon., "Ductile Cast Iron," Engineer, vol. 191, May 25, 1951, pp. 684-685.

68. Anon., "Heavy Duty Clutches Now Produced in Ductile

Iron," Nickel Topics, vol. 4, no. 6, 1951, p. 7.
69. A. L. De Sy, "Spherulitic Formation in Nodular Cast Iron," Metal Progress, June 1951, p. 798.

70. J. Keverian, "Growth of Graphite Spherulites in Cast Iron," Master's Thesis, M.I.T., August 1951, to be published:

71. W. P. Fishel and R. C. Bramlette, "Some Experiments in Preparing Nodular Iron," American Foundryman, vol. 20, no. 3, Sept. 1951, pp. 55-56.

72. F. H. Buttner, H. F. Taylor and J. Wulff, "A Note on the Spherulization of Graphite in Cast Iron," AMERICAN FOUNDRY-MAN, Oct. 1951, p. 49.

73. J. F. Jordan, "Method of Purifying Molten Iron," U.S. Patent no. 2,530,368, Nov. 21, 1951.

74. Fuchs, "Determination of Magnesium in Cast Iron," Association Technique de Fonderie, 24th Foundry Congress, 1951, Preprint no. 6, p. 7. 75. Office of Naval Research (London): "Nodular Iron Re-

search of Ghent," Technical Report ONRL-17-51.

76. J. E. Rehder, "Effect of Phosphorus Content on Mechanical Properties of Nodular Cast Iron," TRANSACTIONS, American Foundrymen's Society, vol. 59, p. 501 (1951).

77. E. Piwowarsky, "Status, Development and Future of Nodular Gray Cast Iron," Die Neue Giesserei, vol. 195, no. 4, Dec. 1950, pp. 175-180.

78. L. F. Spencer, "Ductile Iron Progress Reported at Hartford," Metals Review, vol. 24, no. 3, March 1951, p. 10.

79. W. P. Fishel and R. C. Bramlette, "Some Experiments in Preparing Nodular Iron," AMERICAN FOUNDRYMAN, vol. 20, no. 3, Sept. 1951, pp. 55-56.

80. W. H. White, L. P. Rice, and A. R. Elsea, "Influence of Silicon Content on Mechanical and Heat Treated Properties of Nodular Cast Iron," Transactions, American Foundrymen's Society, vol. 59, p. 337 (1951).

### AIR POLLUTION AND THE CUPOLA

By

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During the past 15 years the problem of air pollu tion has increased in importance to man and his everyday environment. Admittedly, during this period there has never been major concern regarding the disposal of various stack effluents into the limitless air reservoir around us. However, recent localized conditions producing both irritation and health damage have spurred on scientific action in controlling these various sources of contamination. The recent experience in Donora, Pa., and the conditions in Los Angeles are primary examples of the problem.

The foundry cupola is a melting unit used largely by gray iron foundries for the purpose of reducing iron and steel to the molten condition prior to pouring into molds to make castings. To obtain the high temperatures necessary to melt these materials it is necessary to have practically complete combustion of the coke, which is the normal fuel used for this purpose. To obtain this good combustion it is necessary to supply a forced draft through the "tuyeres" located in the lower part of the cupola. This draft can blow finely divided particulate matter and fumes out of the top of the cupola and into the surrounding air.

This particulate matter and metal fumes are the major parts of the problem of air pollution from a cupola. Admittedly, certain gases are also produced which should be considered in the overall aspect of pollution. Another point to be considered is smoke. Smoke does not exist specifically at the time of "lighting off" the cupola. The extent of this smoke due to wood, oily rags, and other methods of igniting is a part of the problem and should be reviewed.

Open Charging Door Cupola. The open charging door cupola admits the coke and charging metal through a side charging door (the dimensions of which depend on the metal to be charged), and the gases pass out a vertical open-top stack. During the melting operation, dilution air enters through the charging door. This dilution air provides a cooling effect to the gas. This dilution also supplies additional oxygen to burn any carbon monoxide and other carbonaceous materials in the combustion gases.

Closed Charging Door and Closed Top Cupola. The closed charging door cupola is one in which all air supplied originates from the tuyere blower. The charging door is so designed that the charge passes through a double door air lock, thus excluding the dilution air. This arrangement is such that only minor secondary

combustion takes place above the charge and the entire system is under pressure during the blast period. The depth of the charge in the cupola is greater than in conventional cupolas (depth between tuyeres and charging door), and is maintained at a specific minimum level. This increased depth of charge tends to provide cooler stack gas temperatures since more heat is absorbed in the charge. The low temperature of the stack gases and absence of secondary combustion means that carbon monoxide is retained unburned in the stack gases,

In evaluating the entire problem there seem to be three major questions which need answering. First, how extensively does cupola dust, fumes and smoke contribute to air pollution? Secondly, what are the various means available for control of cupola emissions? Thirdly, what is the cost and what improvement can reasonably be expected? A complete answer to these three questions would be very valuable information. In part some of the facts leading to these

answers are available.

#### Sampling Technique

Figure 1 illustrates the fundamental equipment used in sampling for particulate matter. A stainless steel nozzle is attached to a machined thimble holder which is connected to the dry-gas meter and thence to a source of suction. On the dry-gas meter a thermometer and U-gauge are attached respectively to the inlet and exhaust sides of the meter. All stacks where this sampling probe is used are fitted with a 21/2-in. nipple and plug. The plug can be removed and the entire sampling head placed inside the stack. A plug with a hole the diameter of the sampling probe pipe extension is used to insure a tight fit.

Initially a pilot tube is inserted through the sampling hole and a velocity traverse of the stack is made. From these results the size of sampling nozzle is determined in relation to the suction capacity of the pump. Normally, the suction pump used is one which gives a 1 cu ft per min flow with this type of system.

The thimble used in the thimble holder head of the probe is ceramic. It is dried at 105 C (221 F) in the laboratory in a weighing bottle to a constant weight. After sample collection it is again dried to constant weight. The difference indicates the total dust load collected during the sampling period.

A standard pyrometer is used to determine stack gas temperatures. The pyrometer can be inserted into the stack through the same or a different sampling port.

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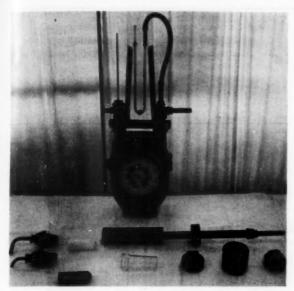


Fig. 1-Equipment for sampling particulate material.

Gas analyses are of the grab type and are run with standard orsat gas analysing equipment. A stainless steel sampling probe is attached to this unit to obtain the sample. The same sampling port used for the particulate matter sampling is used in this sampling.

#### What is the Air Pollution Problem

A series of stack samples were collected on one of our number 8 size cupolas which has a 78-in, diam shell and a 60-in, inside diameter.

This cupola is an open top continuous charge, through open charging door, type. It has a capacity of 85-90 tons per 8 hours. This cupola is charged with 5500 cfm of air through the "tuyeres" and the material shown in Table 1.

TABLE 1

Silicon Briquettes	
(40% SiO ₂ , 45% Fe, 16% CaO)	3%
Basic Pig Iron	35%
Black Scrap	52%
Cast Iron Briquettes	10%

Normally this cupola will be ignited with oily rags and wood. In approximately 3 hr the metal is ready for tapping. During this 3-hr warm-up period there is some discharge of black smoke and red hot particles, especially when the air blast is put on for the first time. This black smoke discharge will last for approximately 6 min at that time. Thereafter, the discharge is a gray or reddish-gray in appearance.

Results obtained during light up and at the initial air burning operation were as shown in Table 2.

TABLE 2

Operation	Temp. in	Grain Loading at 500°F in gr/cu ft	% CO.	%	% CO
Preheat	254	0.038	- 2		0
Preheat	290	0.172			
Preheat	200	0.134	0.3	20.6	0

Following this preheating operation the air blast is turned on and within a few minutes tapping is started and the cupola is in full operation for the day. Table 3 illustrates the dust and gas loadings during the operation.

TABLE 3

Tem Operation	p. in °F	Grain Loading at 500°F in gr/cu ft.	% CO ₂	% O ₂	% CO
Begin Melt Down	914	0.284	6.0	14.4	0
Begin Melt Down	904	0.454			
Begin Melt Down	670	0.369	4.9	15.5	0
Normal Operation '	748	0.672			
Normal Operation 8	368	0.701			
Normal Operation 9	947	0.754			
Normal Operation 1	076	1.867	7.6	12.8	0

Data from other sources would seem to confirm the figures shown. In open top cupolas with open charging doors the average of a series of gas analyses from different size cupolas by personnel of the Air Pollution Control District, County of Los Angeles showed 12.9% CO₂, 7.6% O₂, and 0% CO. Other data from that source indicated grain loadings at standard temperature from 1.09 to 1.32 grains per cubic foot. Conversion of grain loading figures in Table 3 to standard temperature will bring these two sets of results into close agreement.

Tests of a closed door charging cupola by others shows a gas mixture of 9.9% CO₂, 12.9% CO and 2.7% oxygen. Obviously, closed charging doors reduce greatly the air intake with a corresponding increase in carbon monoxide.

Thus, the question of what degree of pollution a gray iron cupola contributes to the general atmosphere can be stated. In general, with open charging doors carbon monoxide is not a major constituent of the cupola stack. However, with closed charging doors carbon monoxide can be present in an approximate concentration of 13 per cent. Thus, there is a problem here. The smoke from a cupola seems to be blackest for a short period of time immediately following the addition of air through the tuyeres. The degree of pollution at this time is slight when expressed in grain loadings up to a maximum of 0.2 grains per cubic foot at 500 F. During normal operations loadings from 0.6 grains to 1.9 grains (both expressed at 500 F) have been established. Obviously, during normal operations of open top and open charging door cupolas an air pollution problem due to particulate matter exists.

#### Survey of Means of Control and Cost

The Gray Iron Foundry Smog Committee of Los Angeles County conducted tests on collection equipment divided into four categories:

- (1) Wet Scrubbers
- (2) Centrifugal Dry Collectors
- (3) Cloth Filters (baghouse)
- (4) Electrical Precipitators

The two wet scrubbers tested by this group attained efficiencies of 57 per cent and 70 per cent on a weight basis. Even this efficiency would not be adequate to

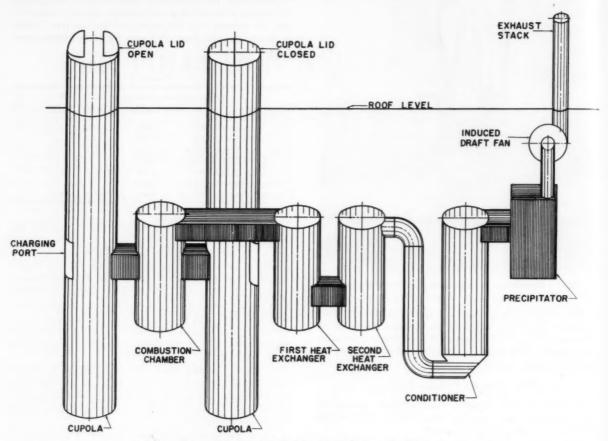


Fig. 2-Diagramatic sketch of electrical precipitator.

satisfy most air pollution codes in the country if an average stack effluent approaching 1.0 gr/cu ft is considered.

Many references are available in the literature pertaining to water washers over a cone type shield in the top of cupola stacks or simply as sprays in this area of the cupola. Few if any of these references cite exact figures regarding efficiency other than to state that the effluent is reduced below the local fly ash ordinance limit, or that visibly there is good control. Costs of these installations are estimated anywhere from \$2,000 to \$10,000, according to the size of the cupola, type of nozzles, method of sludge return, etc.

One dry type unit of the louvered cone principal is referred to in the literature. In this reference costs are cited at from \$0.20 to \$0.30 per cubic foot of gas. In addition it is stated that the unit is efficient in controlling dust and will take particulate matter out of the effluent gases down to about 1 micron in size. Our experience with this type of unit on other than cupola dusts and fumes is generally good. More specific facts though are needed regarding this installation.

Other dry type collectors have been referred to in the literature with the statement, "No tests, but visual observations favorable. Very little dust emitted to atmosphere. This installation is of very recent date; full operating information, therefore, not yet available." Thus, the dry collectors may still be a factor to be considered.

A dry centrifugal collector was tested by the Gray Iron Foundry Smog Committee, but it also failed to meet the requirements.

Three types of glass cloth filtering units were tested and each removed sufficient dust and fumes to come within the Los Angeles County regulations.

A test was conducted by the Air Pollution Control District of Los Angeles County on a sonic wet agglomeration collection system (pilot scale model) on cupola stack gas emissions and this unit was found to be nearly 100 per cent efficient. However, the possibility exists that full scale operation may possess inherent characteristics, unlike pilot scale operation, which could produce considerably lower efficiencies, and thus render this system unsatisfactory. No full scale tests have ever been made.

Various pilot scale tests have been conducted on electrical precipitators and, as suspected, these units can be more than 90 per cent efficient on cupola stack emissions

Work done in Los Angeles has indicated that bag filtering systems where precooling of the gas and preburning of the carbon monoxide are carried out is satisfactory. Glass wool bags will filter a gas at a temperature of 500 F satisfactorily. In addition, orlon bags will work well at 275 F or less. In either system a

gas burner is required at the charging door level to ignite the unburned carbon monoxide and any miscellaneous oil vapors in the gases. On the basis of a number of proposed evaporative cooling and glass wool bag filtering systems now under construction in the Los Angeles area the cost of all equipment exclusive of the cupola itself, is approximately \$1.80 per

cfm of gas entering the baghouse.

Unpublished information made available to the writer indicated that the first orlon bag type collector installed in the Los Angeles area has been accepted by the local authorities. There are no specific data yet available on this installation. This cupola had 5,000 cfm of air supplied at the tuyeres at S.T.P. and handled 22,000 cfm of air at the collectors at a temperature of 275 F. The cost of this entire installation was estimated to be \$1.60 per cfm at 275 F including exterior painting of all of the equipment.

Tests and research conducted, as of this writing, indicate that only two types of collection equipment, and possibly a third, are capable of collecting the suspended solids in cupola stack gases with an efficiency of 90 per cent or more. These three types are the cloth, glass or orlon type filtering unit, the electrical precipitator and the sonic wet agglomeration system.

Sonic Wet Agglomeration System. Basically, the sonic collection system is installed on a closed top cupola with spray nozzles to cool the gas by evaporative cooling to a temperature of approximately 200 F, a sonic agglomeration chamber and a water wash collector. As stated before a pilot model has shown efficiencies in excess of 99 per cent. In this pilot model part of the gas from the top of the cupola stack was drawn through a 3-in. diam duct and then passed through the collector. It is possible that the high efficiencies attained by the pilot scale model may not actually exist in a full size installation. No installation of this type has yet been made or is under contract, as of this writing, on a gray iron cupola.

The first cost of a sonic wet agglomeration collection system is somewhat greater than for a cloth filtering system and may be roughly estimated at \$6.00 per cfm. This cost estimate is based upon contractor's estimates of two proposed installations in Los Angeles County. These proposed installations were abandoned

because of excessive power requirements.

The Electrical Precipitator. The electrical precipitator may be used for collection equipment for cupola stack gases and is best adapted to the larger sized cupolas. Because of the high first cost of electrical precipitators, such equipment is a possible solution only for the medium and large sized foundries.

One large installation of electrical precipitators has been made on an 84-in. inside diameter cupola. The diagramatic sketch (Fig. 2) illustrates the equipment used in this unit. The approximate installation cost for this unit was \$2.80 per cfm at 300 F for the coolers and precipitators. This cost will increase slightly for

larger cupolas.

Certain preliminary tests were made on this precipitator using the sampling system discussed earlier. Due to limitations on accessability only effluent samples from the precipitator were obtained. These results showed a grain loading of 0.037 grains per cubic foot

at 500 F during normal operations. Assuming an inlet loading approaching 1.00 gr/cu ft 500 F this would indicate an efficiency of 96 per cent. Extensive sampling is being carried out on this installation, and a report will be published regarding this unit.

#### Summary

In iron cupola operations the exhaust gases from open top cupolas contain particulate matter loadings varying between 0.6 and 2.0 grains per cubic foot of gas expressed at 500 F. These grain loadings are in excess of practically any known air pollution code. With open charging doors there does not seem to be any large amount of carbon monoxide present in the final effluent due to this added air and the further oxidation of carbon monoxide to the dioxide form. In closed charging door operations there is carbon monoxide present probably to the extent of 13 per cent.

Attempts have been made with most types of collection devices for controlling emissions of particulate matter from cupolas. No accurate figures seem to be available of grain loading efficiencies for dry type collectors or sprays and conical washer types. However, in all of the literature references to these units, their visual efficiency seems to be good. Wet scrubbers have been tested and shown to be from 57 to 70 per cent efficient. Further work being done on wet collectors may yet bring them within an acceptable range.

The use of bag filters of the glass wool and orlon type have been investigated in pilot scale work in the Los Angeles area, and they have been found to be very efficient. Possibly cotton can also be used but more gas cooling will be needed in such installations. Certain glass wool and orlon bag collector installations are being made, but no efficiencies on actual installations are yet available. Pilot scale tests of sonic agglomerators have been made, and are known to be efficient. Electrical precipitators have been installed, but again conclusive reports are still lacking. Theoretically, electric precipitators in pilot scale models control the emissions within any known air pollution code. Preliminary tests on one operating installation indicate that these units will be well within any known code.

The cost of the various types of collectors varies from \$0.20 per cubic foot of gas in dry collectors to over \$6.00 per cubic foot of gas in sonic agglomerators.

#### **Bibliography**

1. "Cupola Dust Suppression," by R. L. Collier, Dec. 1949. A survey report prepared and distributed by the Gray Iron Founders' Society, Inc.

 "Los Angeles Foundrymen Test Air Pollution Control Equipment," T. L. Harsell, Jr., Western Metals, vol. 8, no. 26, March, 1950.

3. "Control of Cupola Stack Emissions," by J. F. Drake, T. G. Kennard and W. A. Saylor, *Iron Age*, April 7, 1949.

4. "Foundry Cupola Dust Collection," by Wm. N. Witheridge, Heating and Ventilating, December, 1949.

 "Contaminants in the Air." Special Report from Industry and Power, 1951.

"Cupola Research Committee Reports," by American Foundrymen's Society, 1951.

7. Personal Correspondence, R. T. Pring, American Wheel-abrator Company.

## STANDARDS FOR ROUGH CHIPPING AND REMOVING WELDS

By Dean Van Order*

■ The most variable operation in the foundry cleaning room is the removal of excess metal and welds. Generally speaking, the chipping and hand grinding operation must put the casting in a condition satisfactory for shipment to the customer. These operations must also correct any defective workmanship of production departments which may have resulted in a poor casting, besides performing the normal work of removing excess metal.

The normal work of chipping and grinding rough castings would consist of removing excess metal in the form of heads or gates that could not be ground on swing grinders, (vents, brackets, fin, nails, chills and similar operations). This type of work when uniform is comparatively easy to measure. The corrective part of the chipping and grinding operation consists of removing burned in sand, extra heavy fin, scabs, welds, swells and other abnormal elements. This part of the operation is one which can never be completely controlled by the production department and is the major item contributing to the variability of the chipping and grinding work.

#### **Previous Methods of Measuring**

Several methods of measuring both the simple and complex parts of this operation were tried before the present system was adopted. Originally, the chipping and grinding operation was a group set-up with each of the men sharing equally in the earnings of the group. This method had its merits in that it was rather easy to maintain and did not require any complex check of production; on the other hand, the degree of incentive was low since the individual did not profit in proportion to his own productivity but rather from the productivity of the group as a whole. Under the group set-up, only the chipping and grinding of rough castings was covered by standards; the removal of welds was not measured in any way but was paid for as day work.

Following the elimination of the group set-up, standards were set for individual piece work. At this time rough and finish chipping was combined, and each

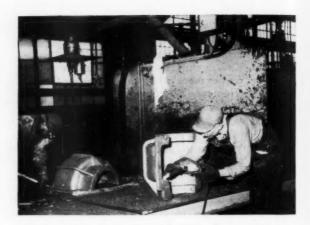


Fig. 1 . . . Front view of chippers bench. Notice that each operator is equipped with air grinder, other items.

operator was equipped with an air grinder, electric grinder, chipping hammer, and individual hoist (Fig. 1 and 2). Standards were also set on the removal of welds. The area of the weld was measured and then classified as Spot, Small, Medium or Large. Time values were set for each of these classifications with an additional allowance for handling depending upon the weight of the casting. Although this method of measurement was a great improvement over the original group set-up and also gave a greater coverage, there were still disadvantages. Checkers had to be hired to measure welds and record pieces and pattern numbers. The accuracy of checking and measuring welds depended entirely upon the type of individual hired for the job and became increasingly apparent with labor turnover. Many castings reverted back to a daywork basis because of unmeasurable amounts of burned in sand and extra heavy fins. Consequently, an improved method was necessary to simplify the checking and still get an accurate coverage of all classes of work on an individual basis.

After considerable research, the electrical department was able to develop a meter recorder which would show the total time that the air grinder, electric grinder, and chipping hammer were operating

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over the day. The writer would like to emphasize that this is a recording of the actual running time of the equipment and not contact time with the castings. Although this meter recorder is rather simple in design, the final assembly and adjustment is of a very delicate nature. The valve that measures the flow of air is composed of an inlet valve and seat, two magnets, a contact screw, an adjusting nut and counterweight, and an outlet opening for the air. The pressure of air on the incoming valve plunger opens the valve and allows the air to flow to the chipping hammer and air grinder. The machining of the valve plunger and valve seat is very important. The angle of the seat must be maintained at a certain degree to allow the air to flow through at constant pressure. The use of the air tool causes air to flow through the air valve which raises magnet A allowing magnet B to drop, thus breaking the contact between the contact

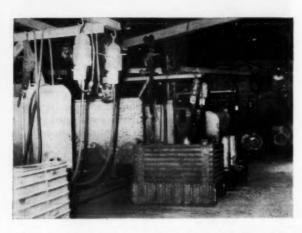


Fig. 2 . . . General view of back of chipping line. The arrangement of hoists and air grinders is conspicuous.

screw and valve body which de-energizes the relay allowing the minute counter to run (Fig. 3). It is very important to see that the air pressure in the lines is kept constant (80 to 90 lb) and free of moisture, rust, and dirt. Any obstruction that would lodge in the valve or valve seat could result in an improper reading at the minute counter. The valve and meter assembly should be inspected and cleaned regularly. A light is hooked up in relay with the minute counter and used to indicate to the operator and inspector that the equipment is working properly.

#### **Centrally Located Counters**

A meter assembly was installed in a dust proof metal case at each work station (Fig. 4). The minute counter for each meter assembly was installed on a single panel in the foreman's office where the amount of minutes produced by each operator could be seen at a glance (Fig. 5). Although each work station is equipped with a meter recorder and minute counter, they are only used for means of payment when the operator is removing welds, or very excessive burnt sand or fins not covered by the chipping table.

The removal of welds (which had always been costly to check and inspect and was only at best mod-

erately accurate in the method of measurement) could now be placed on individual piece work with a minimum of checking and inspecting. Studies were taken on all types, classes and weights of welded castings. An allowance for handling castings (placing casting on work bench either by hand or with hoist, turning casting over, removing casting when completed), was determined depending on the weight of the casting. An additional allowance of 40 per cent was added to the minutes taken from the minute counter for each operator. The additional allowance is composed of 30 per cent for incentive earnings and fatigue, and 10 per cent for looking for welds and a visual inspection by the operator of the weld when finished.

#### Method of Weld Removal

Almost all the welds are removed with the electric grinder unless they are not accessible to the 8-in.

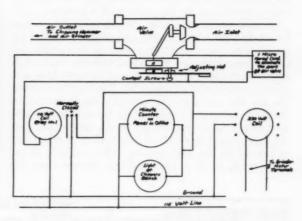


Fig. 3... Diagram of layout and wiring of minute meter recorder, attached to grinders and hammers.

grinding wheel, in which case they may be chipped or removed with a small spindle or cone wheel. Since the electric grinder, air grinder and chipping hammer are all metered through the minute counter, the operator may use whichever tools he chooses. Studies were also taken on chipping out cracks before welding. This type of work is in the minority and is performed only on certain alloy castings and special jobs; however, an additional allowance had to be established. Working strictly with a chipping hammer is much more fatiguing than operating an electric or air grinder, and also builds up less minutes on the minute counter. For this reason, an additional allowance of 110 per cent was added to the minutes taken from the minute counter.

#### Composition of Allowance

This additional allowance is composed of exactly 50 per cent for incentive and fatigue, 30 per cent for handling the chipping hammer, and 30 per cent for locating cracks and inspecting to see if the crack has been completely chipped out. An additional allowance of five minutes per hour on standard is added to both removing welds and chip for welding, for sharpening chisels and changing tools. Ten minutes per day is

added for getting out equipment in the morning and putting it away at night (Fig. 6 and 7 are reproduc-

tions of forms demonstrating this).

Rough chipping had previously been on an individual piece work basis, so the same principles were applied in establishing new standards. Studies were taken on various types of castings and compiled on a standard data sheet which emphasized the fact that fin and sand were the variables that were going to contain the greatest degree of fluctuation. Fin or parting was broken into four classifications: (1) edge regular and spoke fin, (2) edge core fin, (3) flat regular fin, (4) flat core fin. The definitions for these classifications are more or less self-explanatory. Edge regular fin is formed where the parting line is on the corner of a casting. Edge core is formed where a core meets the corner of a casting. Flat regular fin is formed where the parting line is in the center of the casting or on a flat surface. Flat core fin is formed where two cores come together on the flat surface of a casting.

In addition to classifying the types of fin, it was found that the actual length and thickness of the fin varied very closely with the weight of the casting. All studies were taken on the actual amount of fin that appeared in each classification, and then the time was plotted against the potential fin. This makes it possible for the rate setter to measure the potential fin from the pattern. Castings made from loose patterns are only a small percentage of the work that goes through the chippers, yet studies had to be taken and values set for fin and parting. The majority of the loose work is of a more difficult nature than the



Fig. 4 . . . The installed meter assembly. Note light.

mounted work, as the studies showed. To cover this difference, it was found that the fin and parting standards had to be doubled over that for mounted fin.

Tables were made up from the studies on miscellaneous items, such as removing brackets, heads or gates, nail, vent, core wire, chill prints, etc. (Table 2). The allowances for handling were based on the weight of

								AND PA	RTING TA			E				
Length in		0-200				201-4	00#			401-8	00#			801-	Up QU	
Inches	Mig Reg				Edg Reg			Flat	Indg Reg			Flat	Indg Reg		Flat	Flat
	Spoke			Core	Spoke		Reg.	Core	Spoke		Reg.	Core	Spoke	Core	Rege	Core
0~ 10	.14	.27	-41	.53	.14	.29	•43	•56	.24	.48		.93	.29	.57	.86	1.10
11- 20	.21	•41	.62	•78	.19	-47	.70	.89	.42	.84	1.26	1.59	.42	1.12	1.68	2.10
21- 30	.28	•55	.83	1.04	.32	.64	.97	1.21	.60	1.20	1.80	2.25	.84	1.67	2.51	3.11
31- 40	.35	.69	1.04	1.29	.41	.82	1.23	1.53	.78	1.56	2.34	2.90	1.11	2.22	3.33	4.11
41- 50	.42		1.25	1.55	.50	1.00	1.50	1.86	.96	1.92	2.88	3.56	1.39	2.76	4.15	5.11
51~ 60	.49		1.46	1.80	.59	1.18	1.77	2.18	1.14	2.28	3.42	4.22	1.66	3.31	4.97	6.11
61- 70	.56	1.11	1.67	2.06	.68	1.35	2.03	2.50	1.32	2.64	3.96	4.88	1.93	3.86	5.79	7.11
71- 80	-63	1.85		2.31	.77	1.53	2.30	2.83	1.50	3.00	4.50	5.54	2.21	4.41	6.62	8.12
81- 90	.70	1.39		2.57	•86	1.71	2.57	3.15	1.68	3.36	5.04	6.19	2.48	4.95	7.44	
91-100	.77	1.53		2.82	.95	1.89	2.85	3.48	1.86	3.72	5.58	6.85	2.76	5.50		10.12
101-110	.84	1.67		3.08	1.03	2.06	3.10	3.80	2.05	4.08	6.13	7.51	3.05	6.05		11.12
111-120	.91	1.81		3.33	1.12	2.24	3.36	4.13	2.23	4.44	6.67	8.17	3.31	6.60		12.12
121-130	.98	1.95		3.59	1.21	2.42	3.63	4.45	2.41	4.80	7.21	8.83	3.58	7.14		
131-140	1.05		3.14	3.84	1.30	2.60	3.90	4.77	2.59	5.16	7.75	9.48	3.86	7.69	11.55	14.13
141-150	1.12	2.23		4.10	1.39	2.77	4.16	5.10	2.77	5.52		10.14	4.13	8.24	12.37	15.13
151-160	1.19	2.37		4.35	1.48	2.95	4.43	5.42	2.95	5.88		10.80	4.40		13.19	
161-170	1.26	2.51		4.61	1.57	3.13	4.70	5.75	3.13	6.24		11.46	4.68	9.33	14.01	17.13
171-180	1.33	2.65		4.86	1.66	3.31	4.96	6.07	3,31	6.60		12.12	4.95	9.88	14.84	18.14
181-190	1.40	2.79		5.12	1.75	3.48	5.23	6.39	3.49		10.45		5.23		15.66	
191-200	1.47	2.93		5.37	1.84	3.66	5.50	6.72	3.67		10.99				16.48	
201-220	1.57		4.71	5.76	1.97	3.93	5.90	7.20	3.94		11.80				17.71	
221-240	1.71	3.42		6.27	2.15	4.28	6.43	7.85	4.30		12.88			12.89	19.36	23.65
241-260		3.70		6.78		4.64	6.96	8.50	4.66		13.96				21.00	
261-280			5.97	7.29		4.99	7.50	9.15			15.04		7.56	15.08	22.64	27.65
281-300		4.26		7.80		5.35	8.03	9.80			16.12				24.29	
301-320		4.54		8.31		5.70		10.44			17.21			17.27	25.93	31.66
321-340		4.82		8.82		6.06		11.09			18.29		9.21	18.37	27 .58	33.67
341-360		5.10		9.33		6.41		11.74			19.37		9.76	19.46	29.22	35.67
361-380		5.38		9.84			10.16				20.45		10.31	20.56	30.86	37.67
381-400	2.83	5.66	8.49	10.35	3.57	7.12	10.69	13.04	7.19	14.34	21.53	26.26	10.86	21.65	32.51	39.68

- On all loose floor work double fin and parting values. -

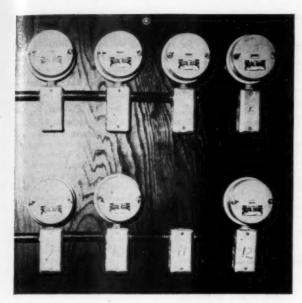


Fig. 5 . . . Minute counters are mounted on one panel.

the casting and were the same as that used for handling welded castings. The removal of burnt sand and extra heavy fin was the most difficult to study and classify. The time for removing any burnt sand or scabs was recorded separately in the study by the time study observer. The time for removing extra heavy fin was also kept separate. The question of what was heavy fin

over and above the normal amount expected was lett up to the judgment of the time study man and the head of the inspection department.

After the allowances in the study for heavy fin had been pace rated and fatigue allowances applied, the normal allowance for the given length of fin was deducted and the difference was plotted. After the curves were drawn, a definite ratio was seen to exist between the amount of extra fin as against the total potential fin and the weight of the casting. To simplify the fin allowance table, a factor was used to classify the fin into one of three classes. This factor was determined by dividing the total potential fin on the casting by the casting weight. Any light casting with a large amount of potential fin would have a higher factor than a heavy casting with a small amount of potential fin and would in return receive a higher fin allowance if one was needed. The writer wishes to point out the very few castings require an additional fin allowance and if they do, it is authorized only by the standards department. Whenever any extra fin allowances are added, the standards department examines all pattern equipment to see if improvements could be made and the fin allowance removed on the next run. The allowances for removing burnt sand and scabs were pace rated and fatigue allowances added, and then they were plotted in the same manner as the extra fin allowances. The burnt sand that appeared did show some relation to the casting weight, but on the whole, it was caused by variables such as casting design, pouring temperatures, gating and heading, wet cores, etc. Four classes of sand allowances were determined under

		oketa												Fin Al		e Facto
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	0-1-	H	0-8-	14	Area	Reg.		Weight	Hal.				BY-HVY			
0- 1-	.23	•27	.20	-26	0- 1ª	•45		0- 25			.42		1.55	•63	2.16	3.69
1.1- 2"	.48	.74	-41	.65	1.1- 2	.94	1.32	26- 50	•56		.84	1.47		.88	2.46	4.04
2.1- 3"	.74	1.16	.62	.98	2.1- 3"	1.44	2.06	51- 75	.77	•60	1.26	2.11	3.71	1.14	2.76	4.40
3.1- 4"	1.00	1.59	.83	1.53	3.1- 4"	1.93	2.81	76- 100	.98		1.69	2.49	4.78	1.38	2.95	4.76
4.1- 5"	1.25	2.01	1.05	1.68	4.1- 5	2.43	3.55	101- 125	1.18		2.10	3,38	5.86	1.63	3.37	5.11
5.1- 6"	1.52	2.44	1.24	2.03	5.1- 6	2.92	4.30	126- 150	1.39		2.52	4.02		1.88	3.68	5.47
6.1- 7"	1.77	2.86	1.45	2.38	6.1- 7	3.48	5.04	151- 200	1.69		3.14	4.96		2.25	3.98	
7.1- 8	2.03	3.29	1.66	2.73	7.1- 8	3.91	5.79	201- 300	8.51		4.40		11.75	3.00	5.04	
8.1- 9"	2.29	3.71	1.86	3.08	6.1- 9ª	4.41	6.53	301- 400	3.14		6.08		16.05	4.00	6.25	
9.1-10	2.55	4.14	2.07	3.43	9.1-10"	4.90	7.28	401- 500	3.96				20.35	5.00		9.93
10.1-11	2.80	4.56	2.28	3.78	10.1-11"	5.40	8.02	501- 600	4.79	4-45			24.65	6.00		11.36
11.1-12"	3.06	4.99	2.49	4.13	11.1-12"	5.89	8.77	601- 800	6.03				31.10			13.50
12.1-13"	3.32	5.41	2.69	4.48	12.1-13*	6.39	9.51	801-1000	7.68			23.45			12.93	
13.1-14"	3.58	5.84	2.90	4.83	13.1-14"	6.88	10.26	1001-1250	9.53	8.99	19.10	29.19	49.38	11.75	15.66	19.58
14.1-15"	3.83	6.26	3.11	5.18	14.1-15"	7.38	11.00	1251-1500	11.59	10.96	23,30	35.56	60.13	14.25	18.69	23.15
15.1-16"	4.09	6.69	3.32	5.53	15.1-16"	7.87	11.75	1500-Up	13.66	13.93	29.60	45.13	76.25	18.00	21.75	26.72
16-1-17"	4.35	7.11	3.52	5.88	16.1-17"	8.37	12.49	*Use for								
17.1-18"	4.61	7.54	3.73	6.23	17.1-18	8.86	13.24	handling		* Fact	or =	Tota	al Pote	mtial F	in	
18.1-19"	4.86	7.96	3.94	6.58	18.1-19"	9.36	13.98	welded				We	ight of	Castin	-	
9.1-20		8.39	4.15	6.93	19.1-20	9.85	14.73	castings								
					Over 200	Reg-20	plus									
	Nails					(.495	I area)	1	Blade 1	Vents						
Butto	ns & F	Rods			Over 20*	Irreg	20 plus	Leng	th	Widt	h					
Rd or	Sig	Stde	1				x ares)			1 1	- 8	1. 1	r. o. a	hills -	(9m -	.25)ea.
Scab		.05						0	-1/2	.07	.10		0 00		(Lg -	.50)ma.
1 to	3/8	.10		unch H	oles	Chill	Prints	5/8-1		.09	.15	2. 0	Chip &			
a to	5/8	.25	Rd	or S	q. Std.	Perim.	Stde	1-1/8-1-	-1/2	.12	.22					50% Ad
1 to		.50		0-1		0-10	-20	1-5/8-2		.18	.29		itional			
1- to		.75	1	-1/8-1	10 1	11-20	-51								at Ste	m 2ª
	1-1/2		1	-5/8-2	.13	21-30	.81					,				
			2	-1/8-8	.16	31-40	1.11		lore W	ires						
				-5/8-3					la.	Std.						
			-					1	8	.20						
								1	4	.30						
								Į į	10	.40						

Table 2, accessory standards covering such items as removing brackets, heads, or gates, core wire, chill prints, etc.

PORM	1000 MEV.	. 200d 1-S	1 660		PREMIUM	TIME	TIC	ŒT							
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				Walds 4	/30282		6		7.68			0	2.56		1/2
	130			-	/02329		1		6.03			40	41	1	R
/30		_		Chip For (	113867	-	6	-	214			30			W
				Wald 1	112419		2		9.33			1			7
-	330			(	102038		6		2.3/			9	32,146		-
_		-					-	-	134-	)		17	32		
												1			
				Meter str.	45150 (	ا عرم	5.4	34R				1			
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				566 . 70.	75 - B2.17									%58	
-	_	-	-	7						-		57			Hoe
60	8							480		50		9		3	5

Fig. 6 . . . Finished check sneet snowing method of payment to operator for removing welds and chipping out for weld. Numbers at top of vertical columns represent the following: 1, Start; 2, Stop; 3, Work Day; 4, Delay; 5, Type of Work; 6, Pattern Number; 7, Part Number; 8, Pieces; 9, On Standard; 10, Unit Standard; 11, Exp. Units; 12, Plus Standard.

each weight bracket. The classification of light sand, medium sand, heavy sand, and extra heavy sand were urawn so as to cover all the existing burnt in sand and scap conditions encountered in the studies.

The use of the sand allowance and fin allowance table is determined by the Time Study Department only. It is up to the time study engineer to keep in contact with all questionable castings and decide which sand allowance and fin allowance is needed if any. He must follow further and see if the condition of the casting can be improved so that the extra fin and sand allowances can be removed. Any casting with more extra work than can be covered by the sand or fin allowance table will not be chipped on direct standards. Instead, the casting will be handled as a welded casting with payment determined from the meter reading and a handling allowance. However, castings that can not be covered by the chipping tables are in the minority, and everything possible is done to correct the condition when it does occur. Following castings through the cleaning room to determine their condition requires on the average from two to three hours work every day by the time study engineer, but it is time well spent as it keeps the chipping operations practically

100 per cent on direct standard and also enables the time study department to recommend methods improvements.

Although many methods were tried and many hours of labor were computed before the final results were obtained for determining standard for rough chipping and removing welds, the numerous advantages are a definite asset.

- (1) Operators are rewarded in direct proportion to their individual effort.
- (2) The cleaning room foremen can control and allocate their backlog of work.
- (3) Chipping operations are no longer the bottleneck of the cleaning department.
- (4) Actual direct chipping costs and weld removal costs per pound can be tied directly to each casting.
- (5) Progress on new operators can be seen at a glance.

(6) Poor methods and costly operations can be detected easily and steps taken to correct them.

Today, more than ever, the modern foundry needs a tool that will accurately measure operator effort with a minimum of supervision and checking if it is to take advantage of any improvement in costs.

Fig. 7 . . . Finished check sheet showing method of payment to operator for rough chipping of castings. Code numbers at top of vertical columns represent the same as in Fig. 6, at top of page. 1, Start; 2, Stop; 3, Work Day; 4, Delay; 5, Type of Work; 6, Pattern Number; 7, Part Number; 8, Pieces; 9, On Standard; 10, Unit Standard; 11, Exp. Units; 12, Plus Standard. Although all columns have not been filled in, the premium time ticket gives a fairly accurate picture of what was accomplished by this worker during his work day. Tickets of this nature are almost a necessity in obtaining maximum benefit from a standards determination program of the sort outlined here.

		/32		Occu	PREMIUM				G-4		Date	12	-17-	-51	
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					111066		13		6.89			M	26		1
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					F5-15		2		2.29			1	1		Total Wage
					OH 28/174		1		9.80			7			3
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#### DISCUSSION

Chairman: J. E. HYLAND, John Deere Spreader Works, East

Co-Chairman: C. J. PRUETT, McWane Cast Iron Pipe Co., Birmingham, Ala.

Recorder: C. J. PRUETT.

H. W. ARENDES (Written Discussion)1 The author's Company has succeeded in installing a workable incentive system on an operation involving numerous variables and problems. This paper shows that a thorough analysis of numerous timestudies was made before the final answer was reached. Mr. Van Order indicates that the several basic requirements of a good incentive plan such as increased production, financial reward to workers in direct proportion to effort, decreased costs, etc., have been met. The organization should be complimented for this achieve-

This incentive plan, like all good plans, has been tailored to meet the needs and particular conditions of the shop involved. A number of the features, no doubt, would work satisfactorily in almost any plant, while others would probably only fit the particular conditions existing at the author's plant.

Individual incentives, in this particular case, worked out better than group incentives. This is the usual experience. At one time, we attempted the installation of an incentive plan with Chippers working in groups of eight. The problems of work division for various casting designs was a headache as was the problem of getting the men interested in an incentive plan that used the productivity of the group to determine incentive bonuses. We had to resort to the use of individual incentives before results were attained.

While group incentives have their advantages and are desirable on certain operations, work such as chipping and grinding should be placed on a direct incentive plan if at all possible.

The use of a minute counter for measuring work on extremely variable items such as chipping and grinding welds, gouging out defects and removing burnt sand is one way of handling a difficult work measurement problem. The minute counter has the advantage of paying on the basis of tool usage time and should be an incentive for the operator to stay on the job. However, perhaps because I do not have full knowledge of this method of work measurement, I have not been able to figure out as to what incentive there would be for a man to put any pressure on the grinding tool and what incentive there would be to get the job done as quickly as possible. It seems that power consumption on an electric grinder (the type mentioned as chief tool used for this work) would be another possible means of work measurement.

Another way that this matter has been handled in some shops is to measure welded areas on a sampling basis and pay the amount found to all castings chipped and ground of the same design. When it becomes apparent that conditions have changed, another sampling is taken and new value paid on the basis of the latest sampling.

The method of determining the time standards, particularly for chipping fins was reviewed with considerable interest. The establishment of fin chipping standards on a potential or possible fin basis has numerous advantages such as, (1) ability to set standards prior to production, (2) consistent standards no matter who sets the rate, and (3) the possibility of using the data for cost estimates. Potential fin is a very desirable basis of work measurement whenever the analysis of the timestudy data indicates that sound standards can be set by this method.

We have had favorable experience in the use of potential fin on truck bolster, side frame and other so-called specialty castings. Since these are production items, a standard data formula, using the potential fin method has been established for each type of product rather than by weight groups.

The author's Company follows the desirable principle of paying excess allowances for excess work such as heavy fins above normal, heavily blocked pin holes, etc. These items can best be handled as special allowances. When a special payment is made to pay for a corrective condition, it comes to the attention of all concerned and steps can be taken immediately to correct the excess condition.

The Timestudy Group determines when to pay the excess allowance and determines the amount to allow, which is in line with good incentive practice. Determining when and how much excess allowance to pay can become a field for argument with the worker in the shop. If the situation gets out of hand, the entire incentive system can fail. We have found it necessary to use timestudies and measurements of areas chipped to prove whether conditions are beyond the normal range of work requirements or not.

There are three questions I would like to ask regarding this

incentive plan:

1. When grinding welds and performing other types of work recorded and measured on the meter recorder, does the worker have any incentive other than to keep the tool operating as large a portion of the time as possible?

2. How does the Timestudy Department determine when a fin condition is bad enough to warrant use of an additional fin

3. Workers often do a better job than necessary when being timestudied. Work in excess of say 15 per cent over minimum inspection requirements on operations as variable as chipping and grinding is hard to detect yet if the worker gets by with it, the resultant time standard will be loose by this amount. What steps were taken during the obtaining of time studies used to formulate the standard data, to prevent workers from doing a higher quality job than is absolutely required?

MR. VAN ORDER (Reply to Mr. Arendes): In answer to Mr. Arendes' first question, the operator does have an additional incentive in that the number of pieces chipped or ground is recorded and a handling standard is paid to the operator depending on the weight of the piece. These handling standards can amount to as much as \$4 or \$5 over the day's pay, therefore, it is a definite incentive to the operator to turn out as many pieces as he can. So far, it has worked out very well.

In answer to the second question, the Timestudy Department must see the casting in question and then determine if a fin allowance is needed based on their past experience. If the quantity on order or other conditions warrant it, a check study may be taken. Every effort is then made to see that the fin is eliminated so that the extra allowance may be removed.

In answer to the third question, before any casting was studied, the inspector in charge explained to both the operator and the timestudy man just what work was necessary. Even after this precaution was taken, the operator did no more work than was necessary at times. It was then up to the timestudy man to use his judgment and compensate for this extra work in quoting the study. The timestudy man must have considerable experience with the company's inspection requirements on various classes of work to be able to do this.

A. E. OPPENHEIMER (Written Discussion): 2 This standard appears to be built on a sound basis. In building this standard for the type of work involved, the author has apparently overcome the common obstacle which is in effect in many standards today, in that the method for measuring production is cumbersome and involved. In many cases the value of a good job in building standard data is lost because of measurements incorporated which are not too practical to apply.

It is fairly well proven from experience in standards development, factors and classifications must be clear and specific and can not be left to any great amount of interpretation in daily use. This interpretation must be settled in the development of the standard.

Rough chipping castings by no means falls in the simple class of work on which standards can be developed and used for a long period of time without a periodic follow-up to insure that no changes have taken place in the basic work on which th. standard was originally established. If improvements are made in prior operations, such as molding, which affect chipping time and for which standard does not provide for the corrected change, this, of course, will lead to difficulty because the standard will become unsound.

Using air-flow meters to record the relative time standard as in this case, should in the writer's opinion be practical, if the operator can not manipulate the air gun to use air while the gun is not in actual use. Some companies provide that a check should be made of the work involved in any standard data periodically to insure the standard does not become unsound.

The author of this standard is to be commended for the apparent fine job done.

¹ Ind. Engr. Supervisor, American Steel Foundries, East St. Louis, Ill.

² Manager, Ind. Engr. Research Division, Deere & Co., Moline, Ill.

H. L. SHELTON (Written Discussion): ² This paper indicates that considerable research work was done on a difficult operation. Most encouraging to the foundry industry is to have persons in management willing to expend effort and money for development.

In formulating and developing Standard Data, just as in installing physical improvements, each foundry must determine which is best for its particular problem and which meets the requirements of optimum economic value. Just as in buying a car, each individual selects the car which meets his needs and his pocket book, so each foundry must consider what it can economically afford to install.

The development of Standard Data has been so well covered by Mr. Van Order it will not be necessary to repeat here. The application of Standard Data for any operation, not only chipping, is almost as individual as the selections previously mentioned.

The characteristics of each foundry's employees have a direct bearing on the application of Standards. At what pace; for how much incentive will your employees work? The reaction of the Supervisors; will their sentiment be with the workers or the Timestudy Department? There are many others, which each foundry will have to discover for themselves.

The Standard Data for Rough Chipping which are most familiar to this writer, have been divided into two sections—Rough Chipping and Defect Chipping, with Weld and Finish chip predicated on the Rough Chip standard time.

We use a Standard time computation sheet. This sheet is started from the molding computation sheet. Many of the elements involved in rough chip can be determined from the molding sheet. Therefor, in our organization when the Timestudy Engineer has computed his molding rate, the computation sheet is passed on to the Timestudy Engineer who has the responsibility of establishing the chipping rates.

The partially prepared rough chip sheet is held in file until the job shows up at the Sandblast operation, where all new jobs are held aside after the rough sandblast operation. The Timestudy Engineer then completes measuring of fins, brackets, etc., to complete the factoring of the Rough chip operation standard time.

When the Engineers made the timestudies for establishing Standard Data, they segregated the defect chipping time. When compared with the performance rating of the workers this enabled the Engineers to establish percentages of Rough Chip time for Defects, Weld and Finish Chip, which are according to classification of the castings shown in Table A. Provision is made for inserting these percentages into the completed standard time.

In connection with this method of handling a serious variable, additional provision is made in the computation of earnings. Under this Incentive Plan, we start paying incentive at 81 per cent performance. That is, if a worker produces 6.5 hr of work in 8 hr on standard, he starts to earn incentive. For any performance under 133 per cent the plan provides for allowance of Variable Expense Allowance according to a predetermined table. Hence, if a worker makes 6.6 hr in 8, the variable expense allowance table provides 1.6 hr variable expense which is added to the workers 6.5 hr production making 8.1 hr pay for 8 hr work. This foregoing is the explanation of the usual 80 per cent chart, but in the case of the Rough chippers we use a 50 per cent chart to compensate for the variable in molding quality, which means that a chipper who produced 6.5 hr in 8, would receive 2.5 hr

TABLE A

De	fect_Weld_	-Finish C	hip_Per	centages	
Bases:		Class 1			lass 2
	(1)	hin Section	ns)		1 Sections
Defect		.71			.61
Weld		.60			.32
Finish		.38			.24
		I.	E	Group "B	
	Class 1	Class 2	E	Group "B Class 2	Class 3
	Thin	Normal		Normal	Heavy
D. C.			Sections		Sections
Defect	.60	.49	.45	.42	.35
Weld	.56	.40	.35	.21	.17
Finish	.35	.30	.24	.15	.11
Sh	oes—Sprocke D	ets and Si		stings:	
	W	Veld	.05		

Bases:		Class 1		Class	2	
	Th	in Section	1S	Normal S	ections	
Defect		.28		.17		
Finish		.19		.12		
Miscellaneous	Castings					
	Grou	ıp "A"	Group "B"			
	Class 1	Class 2	Class 1	Class 2	Class 3	
	Thin	Normal	Thin	Normal	Heavy	
	Sections	Sections	Sections	Sections	Section	
D-6	.26	.19	.16	.10	.08	
Defect	+000 47					

variable expense making a total earnings of 9.0 hr in 8 actual on standard.

The variable Expense is provided on the theory that poor service or variable quality was responsible for the worker not reaching 133 per cent earnings. Therefor, the Company makes the allowance. This feature is particularly advantageous when changing a department over from piece work or day work to incentive, because it cushions the standard time, and expires automatically when the workers reach 133 per cent earnings.

automatically when the workers reach 133 per cent earnings. Where Mr. Van Order has provided an electro-mechanical means of checking excess work on a casting, we have provided an additional Standard Data chart for use when the worker cannot "make-out" against the rate. The timekeeper keeps track of the chippers time on large difficult castings by Rough time and Defect time. When the chipper cannot "make-out" on chipping defects the Timestudy Engineer measures the volume of each defect chipped out, which is then assigned to a time value to determine the Standard, which is compared with the Defect time previously allowed, and submitted to the Foreman for transmittal to the Union Steward, then changes in the rate are made accordingly.

In the case of burned-in sand and such other defects which cannot be measured by our present methods, after the rough and defect chipping has been completed, a time check is made on burned-in sand or in some instances excess weld chipping, then a Plus-standard allowance is made for the extra time, which represents a 100 per cent performance entitling the chipper to 120 per cent earnings on that time. The Plus-standard allowance is made for any operation which has a non-recurring element that cannot be included into the standard time for that operation, or for any extra work involved on a single casting.

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³ Chief of Timestudy, Continental Foundry & Machine Co., Chicago.

## FACTORS AFFECTING FLUIDITY OF CAST IRON

#### By

#### Lew F. Porter and Philip C. Rosenthal*

#### ABSTRACT

The running or flowing quality of metals is controlled by a large number of variables which can be grouped into two major classifications, namely, variables determined by the mold, and variables dependent on the metal. In the spiral fluidity test, those variables determined by the mold conditions are held as constant as possible, and adjustments made in metal composition, temperature, or melting practice to determine the effect of variables dependent on the metal.

In this paper the effect of temperature and composition was determined for cast irons in the range of 2.0 to 3.7 per cent carbon, 1.0 to 2.4 per cent silicon, 0.3 to 1.5 per cent manganese, and 0.18 to 1.18 per cent phosphorus; and with sulphur of approximately 0.10 per cent. A spiral test mold, modified to give a mean deviation in duplicate tests of less than 2.0 per cent,

Part of the tests was in the form of a "factorial" experiment, which gives information not only on the effect of each element individually, but on the interrelated effects of the elements as

Considering the effects of carbon, silicon and phosphorus on fluidity, with one exception, these elements altered fluidity in direct proportion to their effect on the freezing range of the iron, with the highest fluidity occurring at the eutectic composition. Fluidity decreases drastically on the hypereutectic side. When hypoeutectic irons were compared on the basis of pouring with a fixed degree of superheat above the start of freezing, carbon added 4 in., silicon 1 in., and phosphorus 2 in. to the fluidity spiral for a one per cent increase of the respective element. Manganese gave erratic results, but always negative, with the average effect being a reduction of 11/2 in. for a one per cent increase in manganese.

In general the effects of the elements are additive. Only one interrelated effect was noted. In a low carbon iron, the combination of high silicon and phosphorus gave a greater fluidity than would be predicted from the sum of the fluidity effects of the individual elements.

The study of the effect of temperature on fluidity showed that temperatures greater than 350 F superheat affected fluidity as a straight-line function of pouring temperature, equal to 5 in. on the fluidity spiral per 100 F rise in temperature regardless of composition. As the pouring temperature approached the liquidus, however, the function deviated more and more from a straight line, with the fluidity spiral length approaching zero at approximately 30 F above the liquidus temperature, regardless of composition.

The information obtained has led to a tentative formula and chart which make it possible to predict the relative effects of variations in analysis and pouring temperature on the fluidity of the iron.

Misruns, blows, and the difficulty of pouring extremely thin sections are problems continuously present in the gray iron foundry. These problems are intimately related to the ability of the metal to fill a mold, and therefore it is obvious that a study of the relative importance of the various factors affecting this property is of interest to foundrymen. This ability of a metal to fill a mold is often referred to as "fluidity," and it is in this sense that the term "fluidity" will be used herein. This term should not be confused with the fundamental property of liquids also referred to as "fluidity" and which is the converse of viscosity. Thus dual usage of the term "fluidity" has led many foundrymen to the misconception that the variations which are found in the ability of various molten alloys to fill a mold are primarily dependent on the viscosity of the molten metal. Actually a large number of factors are involved in determining the fluidity, and viscosity is a relatively minor one.

Before investigating the effect of composition and pouring temperature on fluidity, a survey of all factors affecting fluidity and their probable relative importance is in order.1 The variables which influence fluidity may be divided into two major classifications:

Factors related to the mold;

Factors dependent on the metal.

The factors related to the mold may be classified as follows:

- 1. Static and velocity head;
- 2. Size and shape of the mold;
- 3. Composition of the mold;

4. Mold temperature.

The static and velocity head are largely dependent on the gating conditions and the speed of pouring. Naturally, the greater the static and velocity head of molten metal, the greater will be the speed at which a mold will be filled, and therefore the fluidity of the metal will be increased. However, other factors besides fluidity must be considered in the design of the gating system, such as the elimination of entrapped slag, sand, and gases. These factors normally limit the static and velocity heads allowable.

The size and shape of the mold affect the ability of the metal to fill the mold in that a compact heavy-

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^{*} Research Engineer and Professor of Metallurgy, respectively, in the Department of Mining & Metallurgy, University of Wisconsin

section mold is readily filled with metal which would be too sluggish to fill a thin-section mold. The size and shape of the mold are factors which depend on the casting being poured and therefore are not alterable.

The composition of the mold is a variable which can greatly affect fluidity. Such things as mold surface, mold permeability, moisture content, bond strength, gas evolution and heat capacity may have considerable effect on the fluidity of the metal. It has been shown² that normal variations in moisture content are relatively unimportant. Bond strength of the sand must be such that washing is prevented, as particles of sand in the molten metal can markedly affect the fluidity. Mold surfaces are said to have an important influence. In general, the smoother the surface of the mold the greater the fluidity. Naturally, if gas is evolved in an impermeable mold the flow of the metal will be retarded.

#### Influence of Mold Atmosphere

The atmosphere created by the mold or by the mold washes can be a factor. If the atmosphere is oxidizing, surface oxide films may form and may seriously restrict the flow of metal. Conversely, a reducing atmosphere might promote the filling out of the mold. The oxide films also result from metal conditions, as discussed later.

The mold temperature can affect fluidity since the rate at which heat is removed from the molten metal determines the time available for flow before the metal solidifies. High mold temperatures will increase fluidity. However, normal variations at room temperature may be expected to have relatively minor effects on fluidity.

It can be seen that normal molding practice, in general, fixes those factors controlling fluidity that are related to the mold. Aside from these, there are those factors dependent on the metal which also can cause variations in the fluidity. The factors dependent on the metal are:

- 1. The viscosity of the molten metal;
- 2. Surface tension of the molten metal;
- 3. The formation of surface oxide films;
- The formation of non-metallic inclusions in the molten metal;
- The degree of superheat (temperature above the start of freezing);
- 6. The mechanism of freezing.

Measurement of the viscosity of molten metals, made just above their melting points, shows that in general the viscosity of molten metals is of the same order of magnitude as that of water so that, when taking into consideration the weight of the metal, molten metals should flow much more readily than water. The change in viscosity with composition or temperature is not great enough to account for the observed variations in the fluidity of metals.¹

The surface tension of metals is high. For instance, the surface tension of cast iron measured at 2372 F is 1150 ergs per square centimeter, while the surface tension of water measured at room temperature is about 70 ergs per square centimeter. The higher the surface tension the greater will be the pressure

necessary to force a stream of metal into a given channel. However, once the necessary initial pressure has been exceeded flow will occur in the channel and differences in surface tension will not greatly affect the fluidity.

The formation of surface oxide films has been shown to have a marked effect on fluidity. Certain alloys under certain conditions of atmosphere, composition, and temperature form a solid brittle film on the surface of the melt. During casting, the presence of this film on the advancing metal surface, and as broken solid pieces in the molten metal, can markedly hinder flow. This is a factor which may lead to erratic and inconsistent fluidity measurements since variations in melting practice can result in the formation or absence of such a film. Film formation is also governed by mold conditions as already discussed.

A second factor leading to erratic fluidity results is the formation of non-metallic inclusions in the molten metal. During the cooling of molten metals, particularly cast irons and steels, non-metallic particles precipitate in the liquid metal. Examples of particles forming in this manner are aluminum and silicon oxides or silicates and manganese and iron sulphides. Such particles can conceivably affect fluidity by hindering the normal flow of the metal. Their importance is largely dependent on the time at which they form and the amount involved.

#### Non-Metallic Inclusions Reduce Fluidity

Particles forming after part of the alloy has started to freeze would not be expected to alter the fluidity because the solid particles of metal would have already virtually stopped flow. A few particles forming just as the molten metal started to cool would flow along with the metal stream without materially hindering flow. On the other hand, a large amount of non-metallic inclusions freezing out of the molten metal shortly before the liquidus temperature is reached would be expected to reduce the fluidity noticeably.

The degree of superheat, i.e., the amount the alloy is heated above its liquidus temperature, when poured, is an extremely important variable. The metal is capable of flowing until freezing begins. The higher the pouring temperature is above the temperature for the start of freezing the longer it will take for the metal to cool to the point where freezing begins, and the higher will be the fluidity.

The manner in which an alloy freezes has been found to have a major influence on the fluidity of the alloy. Portevin and Bastien³ first presented the theory indicating the importance of the mechanism of freezing in 1932. They investigated a number of low melting binary and ternary alloy systems and arrived at the conclusion that the major factor influencing the change in fluidity in any alloy system was the mechanism of freezing. In turn, the mechanism of freezing was defined as a combination of two variables:

- (a) the process of solidification (solidification range);
- (b) the crystallization (crystal faces and speed of crystallization).

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More specifically, the fluidity varies with the primary solidification range. The greater the solidification range, the lower is the fluidity. Therefore, maximum fluidity occurs when congruent (constant temperature) freezing takes place such as in the case of pure metals and alloys of eutectic composition. Minimum fluidity occurs at the maximum limits of solid solubility. This is true whether tests are made using a constant degree of superheat above the liquidus or a constant pouring temperature.

It is believed that the dependence of fluidity on the solidification range is primarily related to the manner in which solidification proceeds. When alloys freeze through a solidification range, dendrites form throughout the entire cross section of the metal as soon as the liquidus temperature is reached. The greater the range of primary solidification, the greater will be the amount of solid present at a given tem-

perature increment below the liquidus.

Eutectic freezing, on the other hand, occurs as a skin of metal forming at the mold wall and progressing slowly in toward the center of the metal cross section. In the first case, flow is stopped quickly by the mesh of dendrites forming throughout the cross section, while in the case of eutectic freezing metal continues to flow through the center of the solid tube which forms as freezing progresses. An excellent discussion of the freezing characteristics of gray iron has been presented by Dunphy and Pellini.⁴

With regard to the mode of crystallization, when compact polyhedral crystals are the first to form fluidity is better than when dendritic crystals are the first to separate. The dendritic type of freezing occurs

in ferrous alloys.

The generalizations concerning the effect of alloy systems on fluidity are nicely illustrated in Fig. 1 in which the variations in fluidity are plotted on top of the binary phase diagrams for the systems antimony-lead and antimony-cadmium.

Points "A" show the relatively high fluidity of the pure metals as compared with the alloys; points "B" show the minimum fluidity which occurs at the limit of solid solubility; point "C" shows the high fluidity which occurs at eutectic alloy compositions; and point "D" shows the extremely sharp maximum which occurs under the favorable polyhedral mode of solidification of the intermetallic compound SbCd.

With this brief survey of factors affecting fluidity, it is felt that the specific problem, now to be discussed, of gray cast iron fluidity as affected by composition and temperature variations will be more clearly defined.

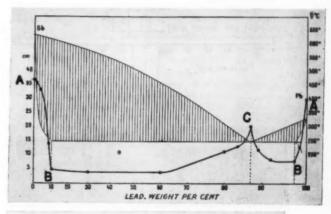
#### Fluidity Test Mold

In order to study the effects of composition and temperature a fluidity test is necessary which will maintain the factors related to the mold essentially constant, while producing a sensitive measure of the variations in fluidity due to variations in the metal. A number of such fluidity tests have been devised. The most popular test is to pour the metal into a mold having a spiral channel of small cross-sectional area. The spiral is long enough to extract sufficient heat to cause solidification of the metal before it

reaches the end of the channel. Thus, the length of the metal spiral obtained is a measure of the fluidity.

Such a fluidity spiral designed by Saeger and Krynitsky⁵ of the Bureau of Standards is in general use, and therefore was adopted for preliminary investigation. The pattern was molded in a naturally bonded green sand containing 6 per cent moisture since this mold composition is one which is normally used in pouring gray cast iron. Four 16-lb induction furnace heats of a standard gray iron composition were made. Two Saeger fluidity spirals were poured from each heat. Although each spiral was poured with metal of the same composition and from the same temperature, the spiral lengths varied from 29 to 35 in. This is a deviation of 9.4 per cent from the mean and was not of sufficient accuracy to produce reliable information for the contemplated program.

Observations made during the test indicated that it was difficult to maintain a steady flow of metal and a constant metal head. It will be remembered from the introduction that this is one of the factors related to the mold which can affect fluidity. Berger and Pering in 1934 used a different gating



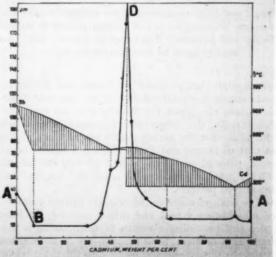
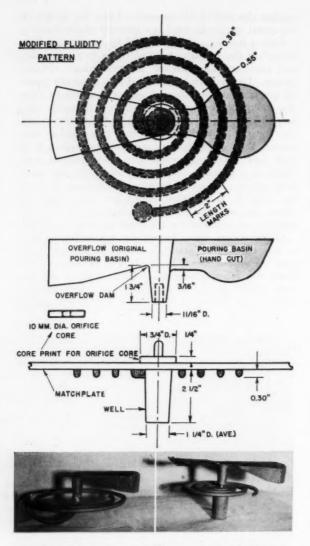


Fig. 1—The fluidity curves superimposed on the binaryphase diagrams indicate the close relationship between fluidity of the metal and the process of solidification.²



Figs. 2 and 3—Scale drawing of the modified fluidity pattern, and photographs of test castings poured in original Saeger and Krynitsky fluidity spiral (above, left), and fluidity spiral as modified (above, right).

system than that proposed by Saeger and Krynitsky, and claimed a reproducibility of one per cent deviation from the mean for fluidity tests on cast iron. Accordingly, the Saeger and Krynitsky pattern was altered so that the gating system agreed more closely to that of Berger and Perin. Figure 2 shows the altered pattern, and Fig. 3 is a photograph of the fluidity test castings poured using the original and the altered patterns.

The changes made in the original pattern consisted of substituting a well and orifice core for the horn gate, and providing a means of maintaining a constant level of metal head through use of an overflow basin. In theory, the first metal into the mold begins to fill the well. This allows time to fill the pouring basin and downspout to the height of the overflow before the metal enters the spiral. Thus, as the well fills and the metal begins to flow into the spiral the

rate of metal flow is controlled by the constant head and the size of the orifice core.

Tests performed using the altered fluidity mold proved this method to be much more consistent. Considering all tests that have been made to date it appears that a mean deviation of approximately 2.0 per cent is now obtained.

#### Procedure and Results

The experimental work on the effect of composition and temperature on the fluidity of cast iron was divided into three major phases. The first phase was an investigation of the effect of carbon on the fluidity of high phosphorus iron poured at constant temperature. In the second phase a factorial experiment was designed to show the main and interrelated influences of the four elements, carbon, silicon, phosphorus, and manganese on the fluidity of cast iron. The third phase was an evaluation of the effect of temperature on fluidity. A companion paper to this one covers the effects of sulphur on the fluidity of gray cast irons near eutectic composition.⁷

#### Fluidity Variations With Carbon and Temperature

The base metal used in this study had the following composition: Total carbon, 3.1 per cent; Mn, 0.30; Si, 2.2; P, 0.70; S, 0.10. Standard melting procedure consisted of melting a constant weight of the base metal scrap in an induction furnace, adjusting the carbon content to the desired value, and then regulating the temperature to that desired. Temperatures were measured by a platinum-platinum 10 per cent rhodium thermocouple in a silica protection tube immersed 21/2 in. into the melt. When the temperature was stabilized at the desired pouring temperature, the couple was removed, the power turned off, and the fluidity mold poured as quickly as possible.

The first variable to be studied was the effect of carbon on the fluidity of the base iron poured at 2600 F. To do this it was necessary to change the carbon content of the iron without changing the rest of the analysis. For carbon contents higher than the normal content of the scrap this was accomplished by adding powdered graphite to the melt. It was necessary to superheat the metal to approximately 2800 F to effect rapid solution of the graphite. Once in solution the temperature was lowered to 2600 F and the fluidity mold was poured.

For carbon contents below that of the normal content of the scrap a somewhat unusual method was used. Related work in the department laboratories had shown that holding the melt at 2700 F caused the carbon to drop while the rest of the analysis remained practically unchanged. Furthermore, this work showed that it is possible to gage the holding time necessary to lower the carbon any given amount. Figure 4 is a graph of this effect of holding time at 2700 F on the carbon and silicon content of gray iron melted under an atmosphere of air in a magnesia crucible.

Using the information shown in Fig. 4 it was a simple matter to reduce carbon content to any given value in the range studied.

The effect of varying carbon on the fluidity of gray

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iron poured at 2600 F was studied by pouring 30 separate fluidity tests. The carbon content was varied from about 2.2 to about 3.7 per cent.

The influence of pouring temperature on the shape of the curve was also determined by pouring low, intermediate and high-carbon irons at 2400 F and at 2800 F.

The results of the tests are shown in Fig. 5 where the carbon content has been plotted against the length of the fluidity spiral. Curves are given for 2400 F, 2600 F, and 2800 F. The sharp break in the curves was definitely established at 3.4 per cent carbon for the curve at 2600 F. Theoretically, this point

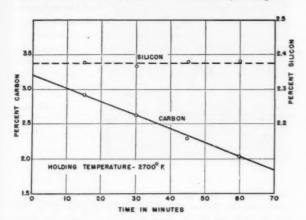


Fig. 4—Variation of silicon and carbon contents as a function of holding time at 2700 F.

corresponds to the eutectic composition of the iron. While this point was not definitely established for the curves at 2400 F and 2800 F, it was assumed to occur at the same carbon content. One test heat having a carbon content in excess of 3.48 per cent did show a reduced fluidity when poured at 2400 F and at 2800 F. It was on this basis that all the curves were drawn with the sharp inflection at 3.48 per cent carbon.

The following points are of interest:

1. Carbon has a great effect on the fluidity of this iron. Between the limits tested from 2.2 per cent carbon to the eutectic composition, fluidity is a straight-line function of carbon content.

2. The eutectic composition has the highest fluidity, but once the eutectic is passed fluidity falls with disastrous rapidity. It was noted that in the hypereutectic compositions graphite separated out of the liquid melt as cooling began. The drastic drop in fluidity is believed due to the formation of kish which, because of its flake-like nature, is very detrimental to the flow of the metal. This agrees with the principles laid down by Portevin and Bastien.³

3. Temperature has an important effect on fluidity. In general, an increase of 200 F results in an increase in fluidity of from 10 to 11 in.

# Factorial Experiment to Determine Main and Interrelated Influences of Carbon Manganese, Silicon and Phosphorus

The "factorial" experimental design is of special interest in research investigations of a large number

of variables since it yields the maximum amount of reliable useful information for the least expense.⁸ In this study the effect of the variables carbon, manganese, silicon and phosphorus on the fluidity of hypoeutectic cast iron was determined. In order to minimize the number of tests necessary only two levels of each of the variables were decided upon.

Specifically, this experiment determined the effect of increasing the carbon from a level of 2.0 to 3.1 per cent, manganese from 0.3 to 1.5 per cent, silicon from 1.0 to 2.4 per cent, and phosphorus from 0.18 per to 1.18 per cent. When all of the variables are at a low level an initial composition is obtained from which all of the remaining compositions can be produced by adding the proper amounts of the specific elements. The initial composition was that of a large supply of uniform malleable iron scrap.

The symbol "I" will be used to denote this initial composition. The symbols C, Mn, Si and P denote carbon, manganese, silicon, and phosphorus, respectively. Whenever a composition contains the higher value of a particular element, the symbol for the elements appears in the symbol for the composition. All possible combinations of the two levels of each of the four variables were included in the experiment.* Furthermore, each composition was tested in duplicate. A list of the symbols and the compositions they represent is shown in the first five columns of Table 1.

All other probable variables were held as constant

[•] If a low-carbon iron is made with a low- and a high-silicon content, and if a high-carbon iron is also made with a low- and a high-silicon content,  $2 \times 2 = 4$  heats will be required. For four elements, each at two composition levels, there will be  $2 \times 2 \times 2 \times 2 = 16$  heats.

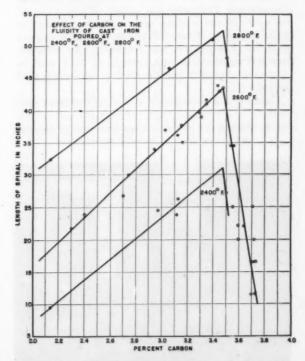


Fig. 5—Curves showing effect of carbon contents on fluidity of cast irons at various pouring temperatures.

as possible. The variable of pouring temperature was particularly important in this study because it determines the degree of superheat (i.e., temperature above the start of freezing). As the composition of the iron changes the temperature at which the alloy begins to freeze changes. Therefore, to separate superheating effects from the actual effects of the composition in fluidity, each composition was poured from a fixed temperature interval above the point at which that composition started to freeze.

The start of freezing was determined by recording the temperature of the molten alloy while it was being slowly cooled and noting where the cooling curve showed a marked reduction in cooling rate. The temperature was measured and recorded by using a platinum platinum 10 per cent rhodium thermocouple in conjunction with a photoelectric potentio-

meter recorder.

A standard melting practice was used in making the various compositions. The melting procedure followed was to charge 8 lb total of scrap having the initial (I) composition to which was added alloying elements to make the composition desired.

Alloys and quantities used for the additions were: Mn, 71 grams 80 per cent ferromanganese; Si, 105 grams 50 per cent ferrosilicon; P, 167 grams 25 per cent ferrophosphorus; C, 50 grams of graphite.

Except for the graphite, all alloy additions were placed in the bottom of the induction furnace under the scrap addition. Graphite was added to the clean surface of the molten iron after the melt had reached a temperature of 2600 F. It usually took about 15 min for the graphite addition to go into solution in the iron at which time the metal temperature had reached about 2800 F. After all additions had gone into solution, the power was turned off and a recording made of the cooling curve. At the instant the metal started to freeze, the power was again applied until 350 F superheat was reached. This temperature was maintained long enough (3 to 5 min) to guarantee constancy, and the fluidity casting was then poured.

Figure 6 is a photograph of the arrangement used

Fig. 6—Photograph of the arrangement used to measure and record start of freezing temperature.



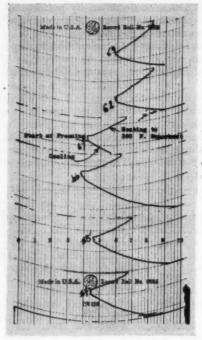


Fig. 7 — Reproduction of a typical graph obtained on the high-speed recorder.

to measure and record the start of freezing temperature and to obtain an accurate 350 F superheat. Directly in front of the operator may be seen the fluidity test mold in place before the induction furnace. Leading from the top of the induction furnace is a silica tube housing the platinum-platinum 10 per cent rhodium thermocouple. The couple junction is at the bottom of the silica tube which extends about  $2\frac{1}{2}$  in. into the molten metal. On the table is the high speed recorder, and alongside of it a portable potentiometer. The potentiometer and recorder units are connected in series with the thermocouple.

Figure 7 is a reproduction of a typical graph obtained on the recorder. The graph shows the curve resulting from the cooling of the molten metal, the break which occurs when the start of freezing is reached, and the curve obtained during the heating of the metal to 350 F superheat prior to pouring the fluidity test.

The fluidity and liquidus temperatures obtained for each composition are listed in Table 1. The usual method of analyzing data from a factorial experiment is through a system of mathematical formulas whereby it is possible to determine the main effect of each variable being studied and all interrelated effects.

For instance, to determine the main effect of carbon on the fluidity, find the average fluidity of all of the heats which contain carbon at the high level and subtract from this the average fluidity of all of the heats which contain carbon at the low level.* The excess of the first average over the second will be the main effect of carbon. In a similar manner the main effect of manganese, silicon and phosphorus can also be determined.

By proper rearrangement of the mathematical relationships, so-called first, second and third-order interactions can be determined. In a first-order interaction we might determine such effects as the influence of increasing the phosphorus content on the effect of carbon.* In a second-order interaction we might determine the influence of increasing the phosphorus content on the influence of silicon on the effect of carbon. A third-order interaction would determine the influence of phosphorus on the influence of silicon on the influence of silicon on the influence of carbon.

Furthermore, it is found that rearranging the order of variables does not affect the results. Thus, the influence of increasing carbon content on the effect of phosphorus is the same as the influence of increasing phosphorus content on the effect of carbon. In this experiment there are four main effects, six first-order interactions, three second-order interactions, and one third-order interaction.

The results of the mathematical analysis of these data are as follows:

#### Main Effects

d

Main effect of carbon = +4.3 in. for a 1.1 per cent increase in carbon content.

Main effect of manganese  $\pm$  — 1.9 in. for a 1.2 per cent increase in manganese content.

Main effect of silicon  $=+1.5\,$  in. for a 1.3 per cent increase in silicon content.

Main effect of phosphorus  $\pm + 2.0$  in. for a 1.0 per cent increase in phosphorus content.

#### First-Order Interactions Involving

Manganese and Carbon	+	0.4 in.
Phosphorus and Carbon	-	0.5 in.
Silicon and Carbon	+	0.2 in.
Phosphorus and Manganese	+	0.1 in.
Silicon and Manganese		0.0 in.
Phosphorus and Silicon	+	1.0 in.
Second-Order Interactions Involving		

#### 

It is seen that the values of all interactions except that between silicon and phosphorus fall within  $\pm 0.5$  in. of fluidity. The range of  $\pm 0.5$  in. is probably within the limits of experimental error for this experiment and, therefore, it may be assumed that except for phosphorus and silicon no interactions exist and that the main effects of the four elements studied are additive and not dependent on what other elements are present. Furthermore, when the results are confined strictly to hypoeutectic cast iron, previous experiments as well as results reported in the literature indicate that the main effect of each of the

elements on fluidity varies linearly with the amount of the element present. Thus it is concluded that:

Carbon increases the fluidity of the iron approximately 4 in. per a one per cent increase in carbon. Silicon increases the fluidity of the iron approximately one in. per a one per cent increase in silicon.

Phosphorus increases the fluidity of the iron approximately 2 in. per a one per cent increase in phosphorus.

Manganese decreases the fluidity of the iron approximately  $1\frac{1}{2}$  in. per a one per cent increase in manganese.

It was found that the data obtained in the factorial experiment can be analyzed in another manner. A plot of the temperature for the start of freezing against the fluidity of the various compositions gave the results shown in Fig. 8.

This graph shows that in general fluidity is inversely proportional to the start of freezing temperature with two exceptions. First, increasing the manganese content causes a somewhat erratic decrease in fluidity without a corresponding change in the start of freezing temperature. Second, the combination of the high level of silicon and phosphorus in a composition low in carbon appears to have a unique first-order interaction which results in a fluidity greater than the additive effects of the individual elements

TABLE 1—COMPOSITION, FLUIDITY, AND LIQUIDUS TEM-PERATURES OF IRONS USED IN FACTORIAL EXPERIMENTS

		Compo	sition,	%	Fluidity,	Liquidus
Symbol	C	Mn	Si	P	in.	Temp., F
I	2.0	0.3	1.0	0.18	211/4	2442
					21	2447
C	3.1	0.3	1.0	0.18	243/4	2210
					243/4	2210
Mn	2.0	1.5	1.0	0.18	183/4	2450
					20	2470
Si	2.0	0.3	2.4	0.18	213/4	2422
					211/2	2422
P	2.0	0.3	1.0	1.18	221/4	2370
					221/4	2372
C Mn	3.1	1.5	1.0	0.18	24	2215
					231/2	2190
C Si	3.1	0.3	2.4	0.18	27	2128
					253/4	2181
CP	3.1	0.3	1.0	1.18	261/2	2110
					261/2	2112
Mn Si	2.0	1.5	2.4	0.18	181/2	2400
					193/4	2400
Mn P	2.0	1.5	1.0	1.18	201/2	2357
					191/4	2356
Si P	2.0	0.3	2.4	1.18	25	2322
					251/2	2317
C Mn Si	3.1	1.5	2.4	0.18	231/4	2230
					233/4	2178
Mn Si P	2.0	1.5	2.4	1.18	221/4	2305
					22	2312
C Si P	3.1	0.3	2.4	1.18	271/4	2050
					281/2	2035
C Mn P	3.1	1.5	1.0	1.18	241/2	2122
					241/4	2150
C Mn Si P	3.1	1.5	2.4	1.18	28	2022
					281/2	2065

Spot check analyses were made for each of the elements after their addition in the amounts indicated in the text. The values given above represent the average of these determinations.

^{*} Thus, the main effect of carbon would be obtained from fluidity measurements handled as follows: [(C + C Mn + C Si + C P + C Mn Si + C Si + C Si + C Mn + C Mn Si + C Mn + C Si + C Mn + C Mn +

and greater than the corresponding change in the temperature for the start of freezing. This point was checked with two additional heats and the results were verified.

The graph in Fig. 8 supports the contention that the freezing mechanism is one of the major factors controlling fluidity. In the case of manganese an additional factor appears to be exerting an influence. Manganese sulphide particles forming under the proper condition could be responsible for the drop in fluidity without a change in the liquidus temperature.

The inconsistent first-order interaction between silicon and phosphorus when carbon is low is not so

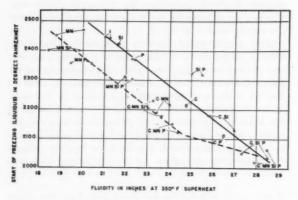


Fig. 8—Relationship between liquidus temperature and fluidity at 350 F superheat.

readily understood. Fortunately, this effect is of minor importance since the combination of high silicon and phosphorus with low carbon does not appear to occur in commercial irons.

A dashed line has been included in Fig. 8 to show the effect of increasing the manganese content from 0.3 to 1.5 per cent. It will be noted that as the iron approaches the eutectic composition the increased manganese rather abruptly ceases to decrease fluidity and the iron in this region exhibits the same fluidity at high-manganese contents as at low-manganese contents. A number of additional heats were made containing 1.5 per cent manganese which gave further verification of the above-mentioned effects.

#### Effect of Temperature on Fluidity

If the results of the factorial experiment could be combined with the influence of temperature on the fluidity of cast iron, it might be possible to derive an expression which would indicate the fluidity of any composition poured from any temperature. The phase of the experimental work dealing with the effect of carbon on fluidity indicated that a temperature increase of 200 F results in an increase in fluidity of from 10 to 11 in. However, these tests were not extensive enough to determine whether this relationship holds at all pouring temperatures and also for compositions in the malleable iron range. Therefore, additional tests were conducted in which a malleable iron composition and a gray iron composition were

poured at 100 F intervals from about 90 F above the liquidus temperature to 2850 F.

The analysis and liquidus temperatures of the two compositions used are given below:

Co	mposi	ition,	Per Cen	t	Liquid	115
C	Si	Mn	P	S	Temp.	F

Malleable Composition . . 2.48 1.24 0.30 0.12 0.15 2360 Gray Iron Composition . . 3.26 2.40 0.35 0.76 0.10 2057

Eight-pound charges of the iron were melted in the induction furnace as rapidly as possible. The liquidus temperature was determined in the same manner as in the factorial experiment, and the metal was reheated to the desired pouring temperature as measured by the platinum-platinum 10 per cent rhodium thermocouple. After adjusting the temperature the power was turned off and the fluidity spiral poured. The data obtained are shown in Fig. 9.

A number of points of interest are illustrated by Fig. 9. First, it is seen that at temperatures higher than 300 F above the liquidus, in the case of the malleable iron, and 350 F in the case of the gray iron, the curves are essentially straight lines. Furthermore, the slope of the straight-line portion is the same regardless of the composition, and is equal to 5 in. of fluidity per 100 F change in pouring tem-

As the pouring temperature approaches the liquidus the curves deviate more and more from a straight line. The extrapolated curves cross the line of zero fluidity at approximately 30 F above the liquidus temperature in both cases. In order for this to occur, the deviation of the gray iron curve from a straight line must be much greater than that of the malleable iron curve. Thus, the improvement in fluidity garnered by adjusting the composition of the iron can be partially or completely lost by pouring at too low a temperature.

#### Discussion of Results

The information obtained from the factorial experiment when coupled with the information concerning the effect of temperature on the fluidity of cast irons makes possible a determination of the relative fluidity of any set of hypoeutectic cast irons poured from any temperature.

The first phase of the work dealing with the effect of carbon on fluidity indicated that the fluidity increases an average of 5 in. per 100 F increase in temperature. This figure was substantiated by the later tests made to determine the effect of temperature, but is valid only at temperatures in excess of 350 F superheat.

The factorial experiment indicated that one per cent carbon increased fluidity 4 in., while 1 per cent silicon increased fluidity 1 in., and 1 per cent phosphorus increased fluidity 2 in. Since there are no interactions present it is concluded that the effects are additive.

This information can be written as a simple formula providing the pouring temperature is above 350 F superheat. The formula resulting is as follows:

Fluidity, inches = 4C + 2P + 1 Si +  $\frac{5 \times \text{Superheat in °F}}{100}$ 

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where the content of each element is given in percentages.

In applying the formula results of calculated fluidity are consistently 6 in. above the experimentally determined fluidity. This is due to the fact that the formula given does not make allowance for a base fluidity value which is independent both of composition and degree of superheat. Thus, the factor of 6 in. can be considered a base correction value, and the fluidity formula then becomes:

Fluidity, inches = 
$$4C + 2P + 1$$
 Si  $+ \frac{5 \times \text{Superheat in } ^{\circ}\text{F}}{100} - 6$ 

In order to apply this formula to a given iron, the amount of superheat must be known. To determine the amount of superheat, the liquidus temperature of the iron must be known. Figure 8 indicates that the carbon, silicon and phosphorus are the elements which determine the liquidus temperature, and that they affect the liquidus temperature in the same way that they affect the fluidity of the metal. Therefore, the liquidus temperature is inversely proportional to the value of the composition factor,  $C + \frac{1}{2}P + \frac{1}{4}Si$ .

This relationship should not be confused with the commonly used criterion for determining carbon equivalent  $= C + \frac{1}{3}$  (Si + P). The composition factor used in the present case is related to the liquidus temperature and must necessarily be different from the relation for carbon equivalent since two irons of different analysis can both have the same carbon equivalent and yet may have different liquidus temperatures. For instance, a eutectic iron of high-phosphorus content would have a lower liquidus than a eutectic iron of low-phosphorus content.

The formula for the liquidus temperaure is: Liquidus Temp.,  ${}^{\circ}F$ , = 2981-218 F, where F is the composition factor (C +  $\frac{1}{2}$  P +  $\frac{1}{4}$  Si). By substituting the formula for the liquidus line in the formula for fluidity, a general expression is obtained which makes it possible to calculate the fluidity of an iron from the composition and pouring temperature without reference to the temperature for the start of freezing. The general expression is as follows: Fluidity, inches = 14.9 F + 0.05 T - 155, where F is the composition factor (F = C +  $\frac{1}{2}$  P + 1/4 Si) and T is the pouring temperature in °F. It must be remembered that the expression will not hold when the pouring temperature is within 350 F of the liquidus because of the deviation of the temperature dependency of fluidity from a straight line in this region.

Since a formula of this type is hard to remember and awkward to use, and since it would be an advantage to be able to determine the relative fluidity of irons poured within 350 F of the liquidus, it has been found convenient to construct a chart based on the data from the factorial experiment and the temperature investigation. The chart is presented in Fig. 10. Using the chart it is possible to estimate the relative fluidity and the liquidus temperature of any set of commercial hypoeutectic irons from malleable compositions through eutectic gray irons, providing the composition of the irons and the pouring temperatures are known.

The formula and chart have definite limitations which should be thoroughly understood if the formula and chart are-to be used intelligently. Some of the limitations result from the assumptions made in designing the chart, and some result from a lack of information concerning the effects of factors which have not been investigated.

Four major assumptions have been made. First, it has been assumed that each element affects the fluidity linearly as it increases in amount. The factorial experiment as used for this investigation is not capable of indicating this, and so it is necessary to justify the assumption on the basis of other observations. The first portion of the experimental work showed that carbon affected the base iron in a linear relationship, and the work of Berger and Perin⁶ indicates that for carbon, silicon, phosphorus alloys, in the commercial range of compositions, the effects of the three elements are linear. Therefore, this first assumption seems reasonable.

Second, since manganese has been found to have an erratic effect, reducing fluidity in the malleable iron range while not affecting it as the eutectic compositions are approached, manganese has not been included in the composition factor. This means that the formula and chart hold only when manganese is in the neighborhood of 0.3 to 0.6 per cent. Fortunately, this is almost universally the manganese range of commercial irons.

Third, the effect of sulphur has not been consid-

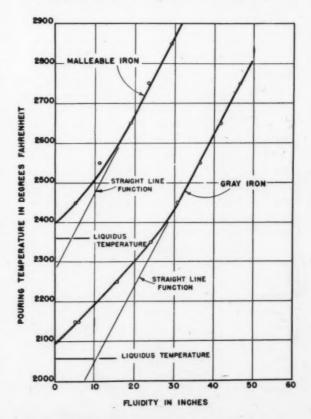


Fig. 9—Effect of pouring temperature on fluidity of malleable and gray cast iron.

ered. There is much information available which indicates that high sulphur decreases fluidity. This point has been investigated and information concerning its effect is presented in another paper. However, at this time it must be recognized that the chart and formula apply only to low-sulphur irons preferably in the neighborhood of 0.10 per cent sulphur.

Fourth, it will be recognized that we have assumed that no interactions occur between the elements carbon, silicon, and phosphorus. In doing this the interaction found between phosphorus and silicon at low-carbon content has been ignored. This is justified on the basis that the combination of high phorphorus and silicon with low carbon is not a normal com-

mercial composition.

A careful examination of the data from the factorial experiment as presented in Fig. 8 will also show that phosphorus and silicon appear to have a greater effect on fluidity when carbon is at the high level than when carbon is at the low level (compare I vs. P with C vs. CP; and I vs. Si with C vs. Si). This fact is not brought to light in the factorial analysis of the data. The effect was carefully investigated and it was decided that to include this possible interaction in the composition factor is not warranted by the only slight increase in the accuracy of the chart and formula so obtained.

The above-mentioned assumptions therefore limit the use of the chart and formula to commercial compositions of normal, low manganese and sulphur contents. For instance, extrapolation to an iron, carbon, phosphorus alloy without silicon would not be justified.

There is also no guarantee that the acutal fluidity measured for heats made in an induction furnace will agree with fluidities obtained for metal made in a cupola, air furnace or by duplex melting. As a matter of fact, work on a related project9 has already shown that heats melted under atmospheres of pure hydrogen or nitrogen have lower fluidity than heats melted under air or carbon monoxide atmospheres. It is probable that the atmosphere used in the experimental work is near enough to that found in a cupola that the fluidities will agree, but only experience with commercial melting units will settle this point. Nevertheless, it is believed that the relative changes in fluidity of irons of different compositions and pouring temperatures made in the same melting unit can be determined from the chart.

An indication of the reliability of the fluidity chart is obtained by comparing the fluidity measured on 42 heats made since the chart has been established. The heats cover the entire composition range from a factor of 2.4 to 4.4, and a pouring temperature range from 2200 to 3000 F. A probability curve for the 42 heats is given in Figure 11. It will be noted that in spite of the assumptions made in construction of the chart the fluidity of all heats falls within 3 in. of the calculated fluidity, and that over 85 per cent fall within 2 in. The average deviation is found to be 1.08 in. Thus, it appears that the accuracy of the chart is satisfactory providing its limitations are recognized and observed.

#### Conclusions

1. It has been shown that normally the mechanism

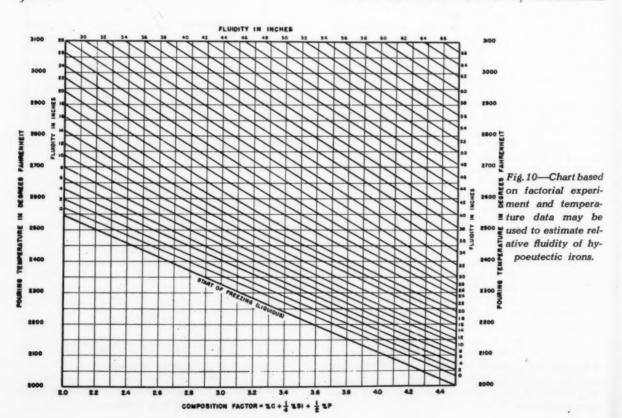


Fig. 11-Probability curve for 42 heats covering the entire range of composition and temperature of the fluidity chart.

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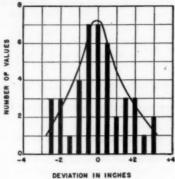
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of freezing determines the effect of composition on the fluidity.

2. Carbon has the greatest influence on increasing the fluidity of hypoeutectic cast iron. Phosphorus is half as effective as carbon and silicon one fourth as effective as carbon.

3. Manganese reduces the fluidity of cast iron for hypoeutectic compositions but this has little or no effect near the eutectic.

4. Above 350 F superheat, fluidity increases rapidly on a straight-line function as the temperature is increased. As pouring temperatures approach the liquidus temperature, the fluidity deviates more and more from a straight-line function, decreasing to a zero measurement on the fluidity spiral slightly above the liquidus regardless of composition.

5. A formula and chart relating the effects of composition and temperature to fluidity have been pro-

#### Acknowledgments

The authors are indebted to Eugene Lange for his assistance in developing the fluidity test and determining the effect of carbon and temperature on fluidity; to Mark Hanson, David Zimmerman, and Charles Armitage, students in the Department of Mining and Metallurgy, for their help in conducting the tests; to Professors G. J. Barker and R. W. Heine of the Department of Mining and Metallurgy, and Professor K. F. Wendt, Associate Director of the University of Wisconsin Engineering Experiment Station, for their cooperation and encouragement; and to the sponsors of the project for furnishing the funds which made the study possible.

#### References

1. C. W. Briggs, "Fluidity of Metals," ASM Metals Handbook, 1948 Edition, pp. 199-203.

2. H. W. Dietert and E. E. Woodliff, "Sand Affects Physical Properties of Gray Iron," TRANSACTIONS, American Foundrymen's Society, vol. 48, pp. 393-424 (1940).

3. A. Portevin and P. Bastien, "Fluidity of Ternary Alloys," Journal, Institute of Metals (1934).

4. R. P. Dunphy and W. S. Pellini, "Solidification of Gray Iron in Sand Molds," Transactions, American Foundrymen's Society, vol. 59, pp. 425-434 (1951).

5. C. M. Saeger, Jr., and A. I. Krynitsky, "A Practical Method for Studying the Running Quality of a Metal Cast in Foundry Molds," Transactions, American Foundrymen's Society, vol. 39, pp. 513-540 (1931).

6. R. Berger and S. Perin, "A Contribution to the Study of Fluidity of Cast Iron," TRANSACTIONS, American Foundrymen's Society, vol. 42, pp. 589-632 (1934).

7. L. F. Porter and P. C. Rosenthal, "Effect of Sulphur on Fluidity of Gray Cast Iron," American Foundrymen's Society, Transactions, vol. 60, pp. 611-616 (1952).

8. M. K. Barnett, "The Factorial Experiment in Engineering

Research," Metals Technology, AIME, June, 1947, T.P. 2161.

9. E. A. Lange and R. W. Heine, "Effects of Melting Furnace Atmosphere on Fluidity, Hot Tearing Tendency, and Other Properties of Malleable Iron," American Foundrymen's Society, Transactions, vol. 60, pp. 169-181 (1952).

#### DISCUSSION

Chairman: J. T. MACKENZIE, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: T. J. Wood, American Brake Shoe Co., Mahwah, N. J.

Recorder: H. W. LOWNIE, Jr., Battelle Memorial Institute, Columbus, Ohio.

E. R. Evans (Written Discussion): 1 The authors are to be congratulated for dealing systematically with the main factors affecting the fluidity of molten cast iron. There are, however, certain limitations to their interpretation of the results and it will be of interest to discuss these since the work reported in this paper has followed similar lines to that carried out by the writer at the British Cast Iron Research Association.*

Apparently, the authors first determined the liquidus temperature for irons of each composition used, then measured the fluidity at a superheat temperature of 350 F. In this way the relative effects of the elements carbon, silicon and phosphorus were determined at this superheat temperature and it was noted that the fluidity increased by approximately 4 in. of spiral for a 1 per cent carbon increase; approximately 1 in. for an increase of 1 per cent silicon; and approximately 2 in. for an increase of 1 per cent phosphorus.

To choose this method of calculating the relative effects of these elements thus invited the magnification of small errors of measurement, whereas, if fluidity had been measured at a constant temperature, similar errors would have been proportionally much less. It had already been shown that at constant temperature a change of 1 per cent carbon produced a corresponding change of about 18 in. fluidity; thus the proportionate error in the case chosen by the authors is 41/2 times as much as it would have been if results had been obtained at constant

Si . P temperature. The composition factor  $C + \frac{1}{4} + \frac{1}{2}$  was calcu-

lated from the results obtained and the authors claim that fluidity is directly dependent on this factor.

In work carried out at the British Cast Iron Research Association calculations were based on fluidity measured at a definite temperature using a slightly more sensitive spiral test (1 per cent carbon change giving about 24 in. change of spiral length).

A composition factor of similar form to that mentioned above Si was derived and expressed as a formula CEF  $\equiv$  C + - + where CEF is an abbreviation for "Carbon Equivalent based on Fluidity.

The limitations of the authors' interpretation is illustrated by replotting their results. It is significant that a satisfactory straight line relationship between fluidity at 350 F superheat temperature and composition factor is obtained with either of the above formulae and with the usually accepted carbon Si P

equivalent expression  $CE = C + \frac{1}{3} + \frac{1}{3}$ . However, when

fluidity at a constant temperature is plotted this confusion of results does not exist. Fluidity at a constant temperature was not measured by the authors but results from my own work may be used to illustrate the point. Figure 1X shows the type of graphs obtained when carbon equivalent calculated from Si P

the formula  $CE = C + \frac{1}{3} + \frac{1}{3}$  is plotted against (a) fluidity at constant temperature, (b) liquidus temperature and (c)

¹ British Cast Iron Research Association, Birmingham, England.

* E. R. Evans, "The Fluidity of Molten Cast Iron," Journal of Research and Development of the British Cast Iron Research Association, 1951, vol. 4, no. 2, October, pages 86-139.

fluidity at 150 C (270 F) superheat temperature. The significance of the symbols in Fig. 1X is in the method by which the carbon equivalent value was altered; for example, P means that the carbon equivalent was varied by varying the phosphorus content of the iron and CP means that it was varied by varying the carbon content in a phosphoric iron. Part (c) of Fig. 1X corresponds to the authors' methods of obtaining results and a satisfactory relationship is illustrated whereas this is not the case in part (a). Figure 2X shows the same results replotted

using the formula CEF = C +  $\frac{\text{Si}}{3}$  +  $\frac{\text{P}}{2}$  with parts (a), (b) and

(c) of the Fig. 2X corresponding to the similar parts of Fig. 1X. In this case a satisfactory relationship is still exhibited at 150 C (270 F) superheat temperature, but a satisfactory relationship is also illustrated with fluidity at constant temperature. It is suggested that these results justify he use of the CEF formula.

Returning to the authors' composition factor 
$$C + \frac{Si}{4} + \frac{P}{2}$$

this has been derived from approximated figures. The main effect of carbon is an increase of 4.3 in, for a 1.1 per cent increase of carbon content which is given as 4 in, for a 1 per cent increase. More accurately the figure should be 3.9 in, for a 1 per cent carbon increase. Similarly the more accurate figure for a

silicon increase of 1 per cent is  $\frac{1.5}{-} = 1.2$  in., while the 1 per cent 1.3 = 2.0 phosphorus increase figure remains at  $\frac{1.5}{-} = 2.0$  in. Deriving the 1.0

composition factor from these figures fluidity is proportional to

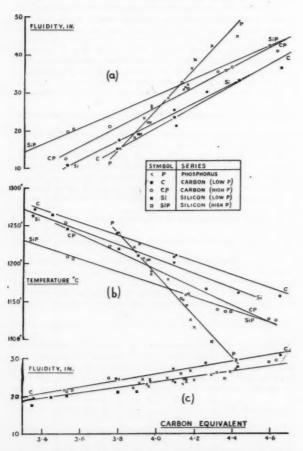


Fig. 1x—Relationship between carbon equivalent value and (a) fluidity at 1350 C (b) solidification temperature (c) fluidity at 150 C above the solidification temperature.

This expression is much nearer to my expression  $C + \frac{Si}{3} + \frac{P}{2}$  than to the authors' own expression  $C + \frac{Si}{4} + \frac{P}{2}$ .

Thus it is suggested that the formula CEF  $\equiv C + \frac{Si}{3} + \frac{P}{2}$  is

the correct one to use for calculating relative fluidity. In their expression derived for calculating fluidity, the authors have included a term bringing in the effect of pouring temperature which assumes that a straight line relationship exists between these two variables. Although the experimental evidence plotted in Fig. 9 justifies this, other evidence does not. If the straight line relationship was correct then, in Fig. 5, at any one carbon content value, the lines drawn at 2400 F and 2800 F would be equidistant from the line drawn at 2600 F. Also, on the authors' conclusions from Fig. 9, the lines shown in Fig. 5 should be parallel. A glance at Fig. 5 indicates that the lines are neither parallel nor equidistant even at temperatures above the 350 F superheat range. It would be expected that the relationship between pouring temperature and fluidity should be illustrated by a curve concave to the temperature axis at all temperatures on a consideration of cooling effects due to increased temperature gradients at higher temperatures. Work carried out at the British Cast Iron Research Association on

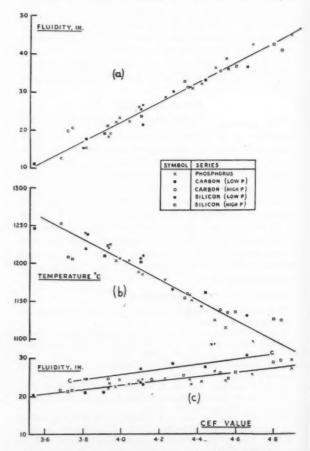


Fig. 2x—Relationship between CEF value and (a) fluidity at 1350 C (b) solidification temperature (c) fluidity at 150 C above the solidification temperature.

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irons of not two, but many compositions, has shown that the graph does take the form of a curve as expected.

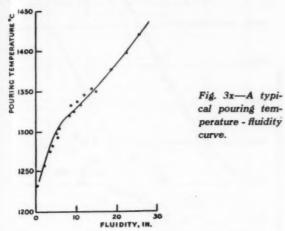


Figure 3X illustrates a typical pouring temperature-fluidity curve with the point of zero fluidity corresponding to the liquidus temperature. The inflection in the curve was noted in all cases where sufficient points were plotted at temperatures approaching the liquidus and very little extrapolation was needed to indicate the solidification temperature as the temperature of zero fluidity. It would be interesting to know whether the authors would have obtained similar results if more tests had been carried out at lower temperatures. It is explained in the paper that test pieces were poured at about 90 F above the liquidus temperature, and then at 100 F intervals up to 2850 F in the case of malleable iron and 2750 F in the case of gray iron. My results have shown that the curved form of the graph continues up to at least 1480 C (approximately 2700 F) for typical gray irons.

With such a variable relationship between pouring temperature and fluidity a very involved formula would be required to calculate fluidity from composition and pouring temperature data, and a chart of the form illustrated by the authors in Fig. 10 has obvious advantages. Figure 10 could be slightly modified to allow for the points expressed above whilst I have expressed the same data in another form of chart illustrated in Fig. 4X.

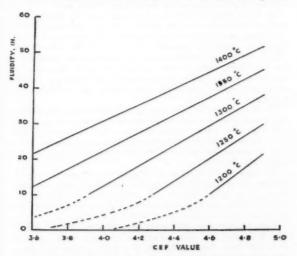


Fig. 4x—Effect of pouring temperature and CEF value on fluidity of molten cast irons.

MESSRS. PORTER AND ROSENTHAL (Written Reply to E. R. Evans): In reply to Mr. Evans, when one alters the carbon, silicon, or phosphorus content of gray cast iron, but pours at constant temperature, he is measuring the effect of two variables on fluidity. The first variable is that of composition changes, the

other is the effect of that composition change on the change in the degree of superheat. It was with the intention of separating these two variables that we poured our spirals at a constant superheat temperature rather than at a constant pouring temperature.

Our experiment was so designed that we obtained the effect of composition on fluidity by calculation, and the effect of temperature was determined separately by experiment. Mr. Evans, on the other hand, determined the effect of these two variables simultaneously, and then by trial and error obtained the factors for composition. The fact that Mr. Evans could not arrive at a good composition factor when he replotted our data in the manner he handled his, points to the weakness in his approach and the obvious advantage of ours; namely, that the factorial calculation clearly determines the effect of such element and its interrelation thus establishing an accurate composition factor without resorting to trial and error methods.

Based on the manner in which our experiment was conducted, we feel Mr. Evans is not justified in his deduction that comparisons made at constant superheat rather than constant temperature subjected our data to a proportionate error of 4½ times.

It was our intention to simplify the data wherever possible. This was the case when we approximated the composition fac-

tor as C + 
$$\frac{\text{Si}}{4}$$
 +  $\frac{P}{2}$ . We agree that Mr. Evans composition Si P

factor of C +  $\frac{1}{3}$  +  $\frac{1}{2}$  is an equally good approximation of our

data. The main bone of contention appears to be whether a factor of ½ or ¼ should be used for silicon. It can be shown in Mr. Evans' Fig. 2X that a value of ¼ for silicon will result in points which for all practical purpose fall as close to the line as those obtained using a value of ½ for silicon. The factor for phosphorus is of greater magnitude and here we agree completely. An illustration of how well a factor of ¼ for silicon checks an actual heat is seen in the following case where an iron having an extremely high silicon content was poured at a temperature of 2800 F. The composition of the iron was: C, 1.91%; Si, 8.22%; P, 0.18%. The fluidity obtained was 47 in. The fluidity calculated from the composition and pouring tem-

perature using the formula, 14.9 (C + 
$$\frac{\text{Si}}{4}$$
 +  $\frac{\text{P}}{2}$ ) + .05 (T) -

155, was 45.2 in.; a very satisfactory check.

We cannot agree with Mr. Evans that the temperature function of fluidity is something other than a straight line at a temperature in excess of 300 F superheat. This function has been carefully determined and we believe that the curves shown on Fig. 9 of our paper are as accurate as is possible for our conditions. Mr. Evans points to Fig. 5 of our paper as evidence that fluidity is not a straight line function of temperature. However, in this case the slope of the curves at 2800 F and 2400 F cannot be considered as absolute since each was established using only three points.

We believe that our determination of the influence of composition and pouring temperature on fluidity is well within the limits of practicability and are gratified to find that independent work by Mr. Evans and the British Cast Iron Research Association is so closely in agreement with the results we have obtained.

RICHARD SCHNEIDEWIND (Written Discussion): The writers should be commended on their success in presenting so well organized a paper on fluidity and showing the quantitative relationship between composition, temperature, and fluidity.

I have had the privilege of examining this paper in the rough manuscript form at the same time as the rough manuscript of E. R. Evans of the British Cast Iron Research Institute. The latter was very recently published in England under the title, "The Fluidity of Molten Cast Iron." The results of both sets of investigators check quite well at 2450 F at the higher carbon equivalent values though there is some deviation at the lower values. This can be explained on the basis of different fluidity spirals and method of reporting the length of the spiral.

The major difference between the results of the two investigators is that Porter and Rosenthal find a sharp drep in fluidity after the eutectic composition is exceeded whereas Evans finds

² University of Michigan, Ann Arbor, Mich.

increasing fluidity up to over 4.8 carbon equivalent (automotive piston ring range).

It is most interesting to find that in the last few years so many investigators have seen fit to modify the expression for carbon equivalent. For fluidity purposes the following have been used:

Porter and Rosenthall 
$$C + \frac{Si}{4} + \frac{P}{2}$$
Evans 
$$C + \frac{Si}{3} + \frac{P}{2}$$
For strength:
$$Schneidewind and McElwee 
$$C + \frac{Si}{3} + \frac{P}{4}$$$$

In view of this fact is it not time to re-evaluate the expression for carbon equivalent to attempt to make it a more accurate Si P

tool or to use the common expression  $C + \frac{1}{3} + \frac{1}{3}$  with reser-

vations as a guide rather than a precise measuring stick?

S. F. CARTER (Written Discussion): Some fluidity work we have done makes us appreciate the volume of valuable data contributed by the authors. The wide chemical variations in cast iron make it a more difficult metal to understand from the fluidity standpoint.

The mold we have evolved to suit our more empirical purposes is similar in principal but different in several details. Our mold is a core sand assembly in contrast to green sand. Our mold also has a spiral of similar dimensions originating from the center and poured through an orifice core. The cross sectional area of our channel is 11 per cent greater. Our test was designed primarily for simplicity and compactness for shop use. It does not have an overflow basin which should make the authors test less sensitive to pouring technique. We do not expect an accuracy any better than 5 per cent on individual

⁸ American Cast Iron Pipe Co., Birmingham, Ala.

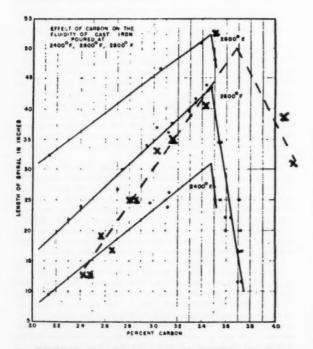


Fig. B—Effect of pouring temperature on fluidity of malleable and gray cast iron (Porter and Rosenthal). Carter's data superimposed. Induction furnace heats are marked X; analysis C, 2.98; Si, 2.15; P, 0.47; Mn, 0.64; S, 0.081. Cupola heats are marked with triangles; analysis C, 3.30; Si, 2.65; P, 0.40; Mn, 0.70; S, 0.085. Oil sand molds. Optical temp., 0.4E.

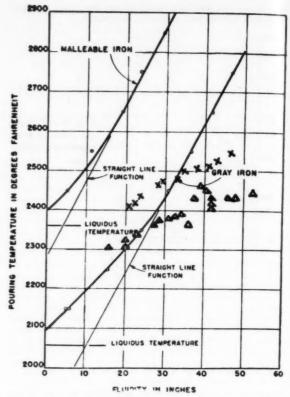


Fig. A—Curves showing effect of carbon contents on fluidity of cast irons at various pouring temperatures (Porter and Rosenthal). Carter's data superimposed. X indicates no inoculation whereas a triangle indicates inoculation with FeSi. Analysis: Si, 1.50; Mn, 0.50; S, 0.060; P, 0.53; C, varied. Poured 2460-2500 F Opt. Induction heat, oil sand mold.

tests, which is not nearly as good as the 2 per cent of the authors. For better accuracy we average several tests.

We have found several very practical uses for fluidity tests. On one operation we use the fluidity test as a routine shop control, where small bolts are cast in permanent molds on a revolving wheel. Tests are poured each hour and the fluidity controlled within an optimum range by making scrap additions to the forehearth ladle. Drastic changes in metal chemistry are also reflected immediately in the fluidity test, making it pessible to make cupola adjustments long before the analysis is reported.

The fluidity test has also been used to determine proper pouring temperatures for unfamiliar compositions of cast iron, steel, and high alloy types. Experiences on a tube of similar dimensions from a familiar chemistry have been translated to an optical temperature of an unfamiliar chemistry that will give equivalent fluidity.

We have found the factors affecting fluidity to be very important in centrifugal casting and indicated with reasonable accuracy by such a fluidity test. Our experience confirms that of the authors that degree of superheat above the liquidus seems to be the most important single factor in fluidity.

Some of our data have been superimposed on the authors Fig. 5 showing the effect of carbon additions we made to an induction heat. See Fig. A. Our data falls in line with that of the authors surprisingly well in view of the differences in mold construction and the lower silicon of our iron.

Similarly some of our data on the effect of temperature has been superimposed on the authors Fig. 9. See Fig. B. The triangles are shop tests of a cupola iron and the X's from an induction heat. On our tests fluidity did not appear as sensitive to temperature. However, a large proportion of our tests are poured less than 350 F above the liquidus since most of our castings are poured 200 to 400 degrees above the liquidus. Con-

sequently our data should probably be compared with the lower end of the gray iron curve. Even then our data seems to give less slope. This might be due to differences in mold construction. I would appreciate the authors opinion on this.

I would like to congratulate the authors on a valuable contribution to one of the sciences very fundamental to the casting

processes.

MESSRS. PORTER AND ROSENTHAL (Written Reply to Messrs. Carter and Schneidewind): We wish to thank Mr. Carter and

Professor Schneidewind for their kind remarks.

We have no definite explanation for the difference in slope between our carbon content vs fluidity curves and that supplied by Mr. Carter. However, it is quite possible that the effect is due to differences in test mold design. The fact that Mr. Carter's carbon content for maximum fluidity is higher than ours can be attributed to the lower silicon and phosphorus content of his base composition.

Mr. Carter has asked our opinion on the difference in slope between our curves and those he has submitted showing the effect of pouring temperature on fluidity. We have found that when an optical pyrometer is used to measure temperature, readings at low metal temperatures will tend to be on the high side because of an oxide film formation which gives a high emissivity factor, whereas at high temperatures, fumes from the iron will tend to give low readings. If this were the case for Mr. Carter's measurements, the differences in slope between his curves and those given by us in Fig. 9 could be explained.

Professor Schneidewind notes that our data differ somewhat from those supplied by Mr. Evans in regard to the carbon equivalent value where maximum fluidity is obtained. In some work conducted since the paper was prepared, we have found that, like Mr. Evans, we have also obtained maximum fluidity at carbon equivalents in excess of the normally accepted value of 4.3 for the eutectic composition. Our data show a decrease in fluidity only after a carbon equivalent value of 4.56 is exceeded.

Our use of C + Si/4 + P/2 as the best representation of the effect of these elements on fluidity was dictated only by the evidence we obtained. We are not suggesting it as a carbon equivalent expression or to complicate the picture. However, it should be recognized that fluidity measurements are governed by what happens to the metal in the liquid state whereas the carbon equivalent is commonly used in connection with solid metal properties. It should not be unreasonable for silicon or phosphorus to have a different effect in the two different situations.

# MELTING IRON IN A BASIC-LINED WATER-COOLED CUPOLA

By

Wally Levi*

This paper deals with the operation of a basic-lined, water-cooled, production cupola of rather special design. Inasmuch as the literature contains very little information regarding basic cupola practice, some development work was necessary before desired results could be obtained with regularity. After several months of experimental work which involved some modifications in cupola design, a procedure was developed which has been quite satisfactory. It must be borne in mind that the practices established were developed in connection with one particular cupola and for this reason cannot necessarily be looked upon as princip'es to be followed to insure successful operation of all basic cupolas.

#### Carbon Control in Cupola

In many foundries today, one of the main problems from the metallurgical standpoint is the production of irons with sufficiently high carbon to satisfy customers' specifications and/or requirements of the castings being produced. The carbon content of the iron in acid cupola practice can, of course, be controlled within reasonably narrow limits by adjusting the carbon content of the ingoing charge. Generally speaking, high carbon irons are produced from mixtures containing high percentages of pig iron and low carbon irons from those containing high percentages of steel. However, in times of peak production sufficient quantities of pig iron are not always available and as a result cupola operators find themselves without the principal high-carbon raw material generally used to keep the carbon content of their irons up to the desired level.

It has been previously reported that somewhat higher than "normal" carbons can be obtained in acid practice through the use of ordinary by-product coke and/or beehive coke in combination with special fuels such as pitch coke, carbon electrodes, graphite electrodes, and briquetted carbon raisers of various types. However, with the materials just mentioned, the increase in carbon above normal will rarely exceed 0.25 per cent and while this is an ap-

preciable amount it is rather insignificant in comparison with what can be accomplished with basic cupola melting. Furthermore, some of the special fuels referred to are rather expensive and the supply is not always adequate.

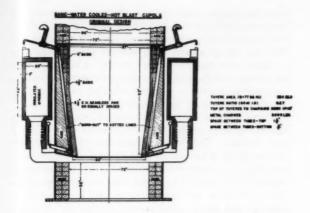
#### Basic-Lined Cupola Installation

The advent of the process for the production of gray cast iron with graphite in nodular or spheroidal form in the as-cast condition is one of the principal reasons for having installed a basic cupola at one of the company's plants. It is well known that irons best suited to the production of nodular iron castings are those containing low sulphur, low phosphorus, and reasonably high total carbon. With acid cupola practice, nodular iron castings produced by the author's company have been made from charges containing from 50 to 70 per cent of premium priced special pig irons, while with the basic cupola the charges have contained 70 per cent of steel and a maximum of 15 per cent pig iron.

Furthermore, the basic irons contain less sulphur with no desulphurizing treatment after melting than the acid irons after sulphur reduction with fused soda ash. Obviously the reduced cost of the metal charge used and the reduction in the quantity of expensive magnesium alloys required to form spheroidal graphite, place nodular iron castings made from iron melted in the basic cupola in a more favorable competitive position than when melting in acid curpolar

Another reason for having installed a basic-lined cupola is the need for high carbon iron used in the manufacture of super deLavaud pipe which constitutes the bulk of the tonnage produced at one of the writer's plants. For this product an iron with a minimum of 3.6 per cent total carbon is desired. With acid practice the cupola charges used contain from 30 to 60 per cent pig iron and the balance cast scrap, while in the basic cupola the charges contain a minimum of 70 per cent steel and no pig iron whatever. Shortly after the basic cupola was put into operation it became evident that the iron produced could well be used as a substitute for purchased pig iron and used to advantage in the acid cupolas operated by

^{*} Chief Metallurgist, Lynchburg Foundry Co., Radford, Va.



All horizontal dimensions shown inside the shell are diameters of circles. Water enters cooling tubes at bottom and discharges at top. Note conical-shaped melting zone, insulated windbox, expansion joints between windbox and tuyeres, and acid lining between shell and cooling tubes.

Fig. 1-Original Cupola Design, Vertical Section.

the company with which the author is associated. The sulphur content is lower than that in most of the merchant iron received and the carbon content is at a uniformly high level. The silicon content can be adjusted within a reasonably broad range and results indicate that it can be controlled within reasonably narrow limits. For example, it was desired to produce a carload of pig iron with approximately 1 per cent silicon. While it was being made, ten samples were taken for analysis and results showed a minimum of 0.95 per cent silicon, and a maximum of 1.16 per cent silicon, with an average for all ten samples 1.09 per cent silicon. A pigging machine having a capacity of 20 tons per hour has been installed and is used in connection with the production of from 1200 to 1400 tons of pig iron per month.

#### Basic Cupola Design

The basic cupola in question was designed in the author's company Engineering Department. It is not a made-over or revamped cupola, but one which is entirely new and stands in the place of one of the four cupolas which have been in service for many years. The shell is 96 in. i.d. from the mandrel plate up to, and including the charging door opening, while the upper portion of the stack is 72 in. i.d.

In the original design, Fig. 1, the water-cooled melting zone section was formed by 60 extra heavy 2½-in. hot rolled seamless steel tubes. They were arranged in the form of a truncated cone having a vertical height of about 7 ft, 9 in. The diameter at the bottom was 5 ft, 6 in., and the diameter at the top was 7 ft, 5 in. Spacing between tubes at the upper ends was about 1¾ in. and at the lower ends about ¾ in. The small diameter of the conical surface was located just above the tuyere openings, the purpose of this design being to provide an "overhang" and thus prevent molten slag or iron from entering the tuyeres during its descent from the melting zone to the cupola well.

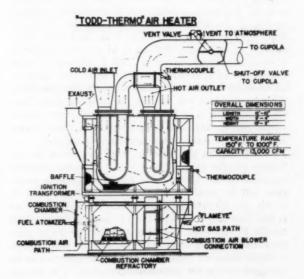
Refractories were used to form the bottom and

sides of the tuyere openings through the lining and an alloy steel plate formed the top. The portion of the alloy steel cover plate toward the inside of the cupola burned off after a few heats and this made maintenance in the tuyere zone rather difficult. It is felt that the tube-type cooling system has many advantages over those built in the form of sectional tanks. Steel piping or tubing is generally more readily available than the special steel castings required to form the tank sections. In case of a build-up of pressure within the cooling system, the tubes can withstand much higher stresses than can the tanks, and consequently the tube system is considerably safer.

Water circulation through tubes is more positive than through sectional coolers principally because there are no "dead" corners where heat transfer might be inadequate. Consequently the tube-type system is the more efficient cooling unit, If necessary, individual cooling tubes can be easily removed for repair or replacement.

Ordinary molding sand covered with about 2 in. of magnesite or a dolomitic type refractory is used in the cupola bottom. The well is lined with burned magnesite brick and the conical surface formed by the cooling tubes was covered by one course of basic brick 4½ in. thick. Daily repairs to the well and melting zone are made with a dolomitic refractory material blown into place with an air gun. Screen analysis and chemical composition of this material are given in Table 1.

Based on information available at the time our cupola was installed it seemed more than likely that the amount of silicon lost during melting with basic practice would be relatively high. (A loss of 10 per cent of the silicon in the charge is usually considered



Diagrammatic sketch of heater shows combustion chamber, heat exchanger tubes, and arrangement of the by-pass feature. Arrows indicate direction of flow of combustion gases as well air used for melting.

Fig. 2-Cupola Air Blast Heater.

TABLE 1-PROPERTIES OF REFRACTORY USED IN DAILY REPAIR OF BASIC CUPOLA

Typical Screen	Analysis*	Typical Chemical	Analysis*
+6	0.0	Ig. Loss	0.79
-6 +8	1.4	SiO ₂	5.38
-8 + 10	15.5	Fe ₂ O ₃	5.14
-10 + 12	10.3	Al ₂ O ₂	1.27
-12 + 14	10.2	$P_2O_5$	0.01
-14 +16	9.7	Cr ₂ O ₈	0.61
-16 + 20	5.7	CaO	39.72
-20 + 28	8.7	MgO	46.50
-28 + 35	3.8	Na ₂ O	0.52
-35 + 48	1.9		
-48 + 100	4.7		
-100+200	8.4		
200	19.6		

^{*} Presented by permission of Basic Refractories, Inc.

normal in acid practice.) Losses up to 20 per cent were reported as more or less "normal" in the basic cupola and T. W. Curry* reported in some of the well-run basic cupolas which he saw in operation in England in the summer of 1950, that the silicon loss was calculated to be 50 points (0.50 per cent) regardless of the silicon content of the ingoing charge. This, of course, represents quite a high percentage loss, particularly if the ingoing silicon content is in the neighborhood of 2 per cent or less.

#### Air Blast Temperature

Research has pointed out that increasing cupola air blast temperature reduces silicon losses during melting in normal practice, and it was felt that these findings would hold true to some extent, at least, in basic practice. In view of the fact that silicon is a relatively expensive commodity, a cupola air blast heater was installed in conjunction with the basic cupola. A general arrangement of the heater is shown in Fig. 2. Its dimensions are 12 ft, 10 in. in length, 9 ft, 4 in. in width, and 19 ft, 5 in. in height. It is capable of heating 13,000 cu ft of air per minute to a maximum temperature of 1000 F, through a heating surface of approximately 4100 sq ft, and based on its capacity it is felt that this is a very compact unit. This heater does not use the combustible gases from the cupola but is of the externally-fired type using a single oil burner, and this feature in itself makes temperature control circuits relatively simple.

Oil atomizers of different sizes are used in the burner to take care of the various load requirements imposed on the heater. For example, a single atomizer will be used to heat between 450 and 600 lb of air per minute (approximately 5,800 to 7,800 cu ft per minute) to temperatures ranging from 500 F to 900 F. To heat this same quantity of air to temperatures ranging from 900 F to 1000 F will require a larger atomizer. When necessary, atomizers can be changed during the course of operation. The change can be made in a few minutes and the resulting drop in temperature is negligible. The blast temperature is held within narrow limits by means of automatic control equipment of the modulating type. This re-

sults in much closer control than can generally be obtained with the "on-and-off" type control. All control equipment, pressure gages, starting switches and the like are mounted on a control panel located in a clean, well-lighted control room. Thus, the man in charge making routine visual inspection of the instruments can tell at a glance whether or not the desired conditions are being maintained.

From a "cold" start the air heater is brought up to operating temperature at a maximum rate of 20 F per minute. This rate is controlled manually until the air blast is within 50 F of operating temperature and then the control is turned over to the automatic equipment. After the heater has been in operation for a period of an hour or more and it is desired to change the blast temperature, it can be done by simply resetting the automatic controls. During the "warm-up" period, air going through the heat exchanger is "spilled" to the atmosphere by opening a specially designed, motorized by-pass valve. As soon as the blast has reached the desired operating temperature, the by-pass valve is closed and melting starts, thus providing hot air at a predetermined temperature from the very beginning of the heat.

If for any reason it is necessary to stop melting temporarily during the course of a heat, the cupola blower is not shut down but the by-pass valve is opened, just as it was while the heat exchanger was being brought up to operating temperature. Melting is resumed by simply closing the by-pass valve and with this procedure the cupola air blast temperature remains exactly the same as it was prior to the shutdown. It is felt that the "by-pass" feature is a distinct advantage over hot blast systems operated in conjunction with cupola blowers which must be shut down in case it is necessary to stop the melting

operation.

In connection with the heater there are a number of built-in safety devices designed to protect both the heater and the cupola. For example, if the water pressure in the cooling tubes inside the cupola falls below a certain minimum, the cupola blower is automatically shut off, and when the blower stops or the air flow falls below a certain minimum level, the burner in the combustion chamber goes out. If the burner flame goes out, the oil flow to the burner stops for a length of time sufficient to purge the combustion chamber of any explosive air-oil mixture, and if the combustion air blower fails, the flow of oil to the burner is cut off. In case of failure of any of the thermocouples in the system, there is a signal light to indicate that this condition exists.

#### Temperature Recording Control Instrument

A recording control instrument makes a continuous record of the cupola air blast temperature. A section of the control chart when operating the heater at 950 F is shown in Fig. 3. When heating about 550 lb of air per minute to a temperature of 850 F, the rate of consumption of No. 2 fuel oil is approximately 60 gallons per hour while the same amount of air heated to 950 F requires about 75 gallons per hour.

Before putting the basic cupola into operation, estimates were made to determine whether or not the

^{*} Metallurgist, Lynchburg Foundry Co., Lynchburg, Va.

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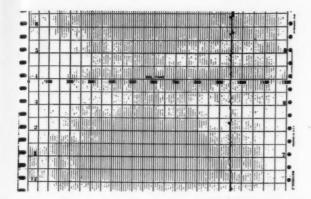
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Figures across the chart (0-1200) represent temperature in degrees F. Control instrument is maintaining blast temperature at 950 F, with occasional deviations of only  $\pm$  10 degrees.

Fig. 3-Section of Air Blast Heater Control Chart.

heat extracted by the water used in the cooling tubes would equal or exceed the amount of heat added by the hot blast. It seemed likely that when using 400 to 450 lb of air per minute heated to 850 F that the heat in-put would exceed the amount taken out by the cooling water. Based on these findings it was decided to use a coke ratio of 6.7 to 1, which is the same as used successfully in the cold blast, acid cupolas with mixtures containing more steel and a lower carbon in the charge than some of the first mixtures used in the basic cupola.

#### Refractory Life

The first few heats run were fairly satisfactory but it is now felt that this is because they were of relatively short duration. The cupola is of the frontslagging type and it soon became evident in order to operate a full 12-hr shift that one of the chief problems was to find suitable refractories for the tap hole as well as for the trough and slag dam. With the strongly basic slag formed and high temperatures encountered (upwards of 2900 F at times) attack on the brickwork was extremely severe. For the tap hole, the first trials were with various sizes and shapes of water coolers, but these were abandoned after a few heats because it was felt that they were unsafe. Next, several different kinds of basic brick were tried, including stabilized dolomite brick, and a number of types of basic ramming refractories. None of these materials were satisfactory under the operating conditions involved.

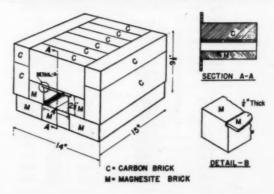
The difficulty encountered lay in the fact that the refractory material forming the top of the tap hole was severely attacked by the slag and within a relatively short time was so badly "eaten-away" that it was impossible to prevent air leakage through the breast. After several heats it was decided to build the tap hole with top and sides of carbon brick and the bottom of magnesite brick. With this construction, there was evidence of attack on the lower portion of the carbon brick forming the sides of the hole but after 4 hr of operation the carbon brick top was found to be practically free of attack. The condition just described indicates that slag and iron flow

through the tap hole in separate layers with the iron toward the bottom as might be expected. The carbon brick sides were probably being slowly dissolved by molten iron flowing against them and this along with some erosive action accounts for the attack noted above. The magnesite brick forming the bottom of the tap hole were, of course, in contact with the iron only and these were found to be in good condition at the end of the heat.

These observations indicate that magnesite brick are more satisfactory in contact with iron than carbon brick, and that the carbon brick are much more resistant to slag attack than magnesite. Consequently, a tap hole was built with magnesite brick bottom and sides, and carbon top as shown in Fig. 4. The brick forming the sides are notched with a masonry saw to permit insertion of a wafer (approximately 1/4-in. thick) of magnesite brick extending the full length of the tap hole. See Fig. 4, Detail B. The purpose of this is to protect the carbon top from attack by the iron until slag starts flowing at the beginning of the heat which may be from 10 to 30 min after tapping out. With this arrangement the upper portion of the magnesite brick sides is subject to slag attack. However, the sides are "backed-up" with carbon as a precaution against failure which might occur at such time as they are competely consumed by the This "dual-refractory" construction has been found to be satisfactory. The magnesite brick referred to are of the hard burned type and should contain as little iron oxide as is possible.

In heats lasting from 24 to 30 hr, continuous service, the same grade of magnesite brick has been found suitable for lining the cupola trough from the tap hole out to, and including the slag dam. After changing the refractory practice as just described, it became possible to gradually extend the duration of the heats from an average of about 2 to 4 hr to approximately 6 to 8 hr. As the length of the heats increased difficulties arising from changing conditions within the cupola became serious. At the beginning of the heats, composition of the iron was for the

#### TAP HOLE CONSTRUCTION - BASIC LINED FRONT SLAGGING CUPOLA



Contruction consists of magnesite brick bottom and sides, with carbon brick top. Sides are backed with carbon brick and top is faced with 1/4-in. thick wafer of magnesite as shown in Detail B.

Fig. 4-Dual Refractory Tap Hole for Front Slagging.

most part much more satisfactory from the standpoint of low sulphur and high total carbon than it was toward the end.

It was reasoned that the favorable results obtained were attributable partially to the fact that the coke bed height of 56 in. above the tuyeres was probably greater than actually necessary especially in view of the fact hot blast was being used. (As air blast temperature increases, the melting zone moves progressively closer to the tuyere level.) The high coke bed should help maintain reducing conditions at least until such time as it "burned down" to a height corresponding to the blast pressure. It was also evi-

dent, as the heat progressed, that the thickness of the lining over the cooling tubes in the melting zone became gradually thinner. Figure 1 shows lining contour at the end of the heat indicated as "Burn-Out to Dotted Lines." It was felt that some of the air entering the tuyeres would pass upward along the relatively cool surface of the thin lining, inasmuch as this represents the path of least resistance,

The lining over the tubes above the melting zone becomes thicker as the distance upward from the tuyeres increases. Thus, with more insulation over the tubes, the air in its upward travel becomes hotter and hotter until finally a "secondary melting

Table 2—Heat No. 3251, With 25% Steel

		Cupola Cha	rge				Flux Char	rge
	Materials		lb %			Dolomite		100 lt
	Steel Scra	р	500 25			Limeston	ne	150 lt
			700 35			Spar		50 lt
			500 30					
	Silvery (8	% Si) 2	200 10			Total .		300 lb
	Totals	20	000100			Amalouia Cu	mala flor of	
Si	S	TO	T F	F-0	C-0		pola Slag, %	Al ₂ O ₈
31	3	T.C.	Temp. F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃
1.16	0.096	3.30	2773	1.65	43.26	33.06	7.89	11.54
0.95	0.082	3.32	2773					
0.43	0.095	2.98	2773	2.32	45.08	33.06	8.55	7.74
0.69	0.122	2.98	2741					
0.49	0.130	2.57	2725	7.53	37.50	34.86	6.72	7.96
0.98	0.089	3.03	2725					
0.79	0.109	2.68	2757	3.00	41.20	36.70	6.52	8.33
0.84	0.108	2.99	2773					
1.12	0.102	2.98	2773	3.84	41.04	34.52	8.22	7.33
		-						
Aver. 0.83	0.104	2.98	2757	3.67	41.61	34.44	7.58	8.58
Silicon I	n Charge, %		2.16		Carbon	In Charge,	6	2.63
Aver. Sil	icon Loss, %		61.5	-	Aver. C	arbon Pickup	, %	0.35
Air Blast	Temp		.900 F		Coke Ra	tio		6.7

Table 3—Heat No. 53151—Charge With 50% Steel Original Cupola Design

		Cupola (	Charge					Flux Char	rge	
	Materials		lb	%		1	Dolomite .		100 11	)
	Steel Scrap		1000 .				Carbide		50 11	)
						5	Spar		20 11	)
	Cast Scrap		450 .	22.5		5	Soda Ash .		4 11	)
	Silvery (89	% Si)	150 .	7.5						
		@ 2 lb Si					Total		174 11	) _
	Totals		2002 .	100.0						
		*Iron					Analysis	Cupola S	lag, %	
	Si	S	T.C.	Temp, F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃	S
-	1.65	0.020	4.08	2908				~		
	1.69	0.017	4.08	2970	0.46	50.40	28.06	8.72	5.10	1.26
	1.46	0.016	4.26	2954						
	1.44	0.015	4.04	2908	0.46	53.48	26.76	9.20	5.58	1.23
	1.56	0.018	4.01	2923						
	1.44	0.015	4.20	2954	0.53	49.70	27.24	8.36	5.88	1.53
	1.72	0.017	4.04	2954						
	1.95	0.013	4.04	2954	0.59	50.04	28.30	8.66	5.96	1.36
	1.01	0.016	4.09	2941	0.51	50.91	27.59	8.74	5.63	1.35
	Aver. 1.61				0.51					
		n Charge, %					on In Char			
		icon Loss, %					Carbon P			
	Air Blas	t Temp					Ratio			5:1
			*Note: I	ron samples fro	m cupola	spout ever	y 30 minut	es.		

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zone some 5 to 7 ft above the tuyere openings is developed. That such a secondary melting zone actually existed was definitely established by the presence of pieces of fused iron and steel scrap adhering to the cupola lining at a considerable distance above the tuyeres. Needless to say the condition just described is highly undesirable and results in considerable "damage" to the metal, because at least a part of the iron melted in the upper melting zone drips for some distance through air containing free oxygen. The iron oxide formed eventually finds its way into the slag and thus reduces its desulphurizing power. As a result, the sulphur content of the iron will be high and the carbon will be low because under some conditions, at least, carbon decreases as sulphur increases.

#### Some Difficulties Encountered

Table 2 gives results obtained in connection with heat No. 3251, with 25 per cent steel, using the original cupola design. At first, the unsatisfactory results obtained were attributed largely to cupola design, though as will be seen later, good results were obtained with this design when larger quantities of coke were used. That poor melting conditions existed is clearly indicated by the high iron oxide content of the slag. Furthermore, the ratio of CaO (41.61 per cent) to SiO₂ (34.44 per cent) in the slag is considerably lower than anticipated and it is felt that this is due chiefly to the excessively high silicon loss. As is well known "silicon loss" represents that portion of the silicon in the charge which is oxidized to SiO₂ during the melting operation. The SiO₂ thus formed is absorbed by the slag and drastically reduces the basicity when large quantities are involved.

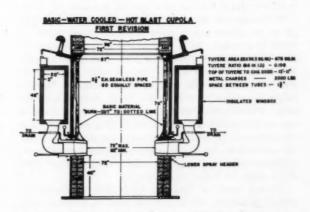
In order to counteract some of the difficulties attributed to cupola design it was decided to attempt a few heats using 21/2 per cent of calcium carbide and coke ratio, exclusive of bed coke, of 5.0 to 1. For each 2000-lb metal charge, 300 lb of coke plus 50 lb of 31/2-in. x 2-in. lump calcium carbide were used. In addition to this an extra 1000 lb of coke was used every tenth charge. It was anticipated that the large quantity of coke plus the carbide would establish a strongly reducing condition favorable to both carbon absorption and sulphur removal. The air rate was 400 to 450 lb per minute at a temperature of 850 F. Results obtained in connection with Heat No. 53151 using a mixture containing 50 per cent steel are given in Table 3, and are more satisfactory than those previously obtained.

After a number of heats, it became evident that it was necessary to use large quantities of coke and a very expensive flux charge, in order to produce the low sulphur, high carbon irons desired. In connection with the original cupola design, development of leaks in the cooling tubes was quite a common occurrence. Most of the leaks occurred near the lower ends of the tubes where molten iron and slag dripped onto them during the course of the heat. This situation was not dangerous from the standpoint of explosions as proven by the fact that in many cases the leaks were not discovered until the bottom was dropped at the end of the day's operation. However,

fairly bad leaks affected metal temperatures, slag fluidity, and chemical composition and in one instance when a serious leak occurred the cupola was completely bridged over. This condition further indicated that the cone-shaped, water-cooled melting zone was undesirable. It was therefore, decided to rearrange the cooling tubes so that they are now parallel with the cupola shell. At the same time, 12 water-cooled copper tuyeres projecting beyond the surface of the cupola lining were installed as shown in Fig. 5.

A feature of these tuyeres is that the distance they project can be adjusted over a 5-in. range. Thus, the circle bounded by them can be as small as 62 in. in diameter, or as large as 72 in. in diameter. Since installation of the cooling tubes parallel with the cupola shell there have been no indications of leaks whatever, over a period of many months of operation. Originally there were some leaks in the watercooled copper tuyeres but it was found that this was caused by a temporarily inadequate water supply. With the revised cupola design it was reasoned that the tendency to establish a secondary melting zone several feet above the tuyeres would be minimized or eliminated entirely. Some of the air entering the tuyeres could, of course, not be prevented from passing out of the cupola upward along the water-cooled wall. However, before reaching the wall it would have to travel upward an appreciable distance due to the fact that it passes through the tuyeres with a fairly high velocity.

It is not conceivable that air entering any particular tuyere would completely reverse its direction of flow at the tuyere outlet and pass directly back to the water-cooled wall at the tuyere level. Furthermore, any portion of the air blast having a tendency to flow outward toward the water-cooled lining would have to travel in a horizontal direction through a thickness of incandescent coke equal to the distance from the tuyere outlet to the surface of the lining over the cooling tubes. Assuming there is 1 in. of refractory over the cooling tubes after conditions within the cupola have reached equilibrium, this dis-



Melting zone is cylindrical with cooling tubes parallel with shell. Note water-cooled copper tuyeres (12 in all) projecting beyond inside of lining, and spray headers to provide cooling water on shell between and below tuyeres.

Fig. 5-Revised Cupola Design, Vertical Section.

Table 4—Heat No. 73151, Charge with 70% Steel

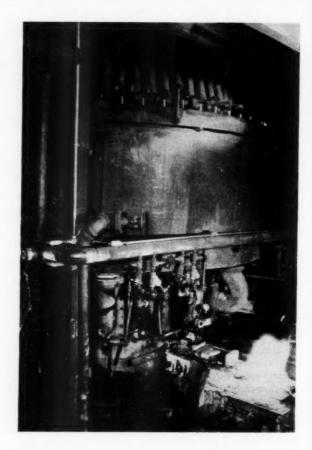
	Cupola	Charge				F	lux Charg	e	
Materials		lb	0%			Dolomite .			b
	р					Carbide		50 lt	)
Cast Scrap	p	500	25			Spar		20 11	b -
	% Si)					Soda Ash			
	@ 2 lb Si								_
			-			Total		174 lb	)
Totals		2002	100						
	*Iron					Analysis	S Cupola S	lag, %	
Si	S	T.C.	Temp, F	FeO	CaO	SiO2	MgO	Al ₂ O ₃	S
1.12	0.013	3.98		0.70	52.68	29.52	8.92	4.30	1.20
0.95	0.010	4.01		0.80					
1.12	0.015	3.92	2923	0.70					
1.13	0.012	3.97		0.76					
1.06	0.014	4.05	2954	0.76	54.68	26.88	7.40	4.46	1.27
1.05	0.010	3.94		0.78					
1.12	0.012	3.92	2923	0.67					
1.03	0.016	3.94	2893	0.74					
1.14	0.018	3.94		0.74	56.24	26.60	6.80	4.76	1.56
1.16	0.015	4.02		0.80					
Aver. 1.09	0.014	3.97	2923	0.75	54.53	27.67	7.71	4.51	1.34
	in charge, %		1.20		Carbo	on in char			.1.17
	licon loss, %					Carbon pi			
	t temp					ratio			

tance would be between 6 and 11 in. depending on how far the tuyeres project beyond the lining. Satisfactory results have been obtained with the tuyeres so arranged as to form a 66-in. diameter circle. Under these conditions they project 9 in. beyond the lining in the melting zone and there is no evidence whatever of the existence of the secondary melting zone previously referred to. Several views (both internal and external) of the revised cupola are shown in Figs. 6 to 10 inclusive.

After making the design changes mentioned above, the percentage of steel was increased and results (representing five consecutive hours of operation) obtained in connection with Heat No. 73151 using a 70 per cent steel charge are given in Table 4. The above metal was poured into the pigging machine, previously mentioned, and shipped to one of the author's plants for use as a special purpose pig iron.

To bring the cupola bed up to the desired height of 56 in. above the tuyeres requires about 5,000 to 5,500 lb of coke. As in acid practice, it was expected that the sulphur content of the first iron tapped would be relatively high. This is due partially to the fact that the first charges absorb sulphur from a larger quantity of coke than those melted later in heat. Furthermore, at the start of the heat there is apt to be insufficient slag volume at a relatively low temperature. At low temperatures the slag lacks fluidity and when this is the case, reactions favorable to sulphur reduction proceed more slowly than with fluid slags at high temperatures.

During some of the first heats 200 lb limestone, 200 lb dolomite, 100 lb carbide, and 16 lb fused soda ash were placed *ON* top of the coke bed. Under these conditions the first iron tapped contained from 0.118 to 0.195 per cent sulphur. The materials were gradually changed until the flux charge *ON* the bed consisted of 700 lb dolomite, 100 lb carbide, 45 lb spar,



General view showing 3 of the 12 tuyere connections. Vents and water overflow pipes from cooling tubes may be seen at the top of windbox. Cupola spout is in right foreground.

Fig. 6-Revised Cupola Design.



This shows "goose-neck" castings between windbox and watercooled copper tuyeres. Note cooling water flowing over shell between and below tuyeres. Front-slagging spout and slag overflow are in right foreground.

Fig. 7-Revised Cupola Design.



View inside of cupola after daily repair with dolomitic refractory blown into place with an air gun. Note projecting watercooled copper tuyeres. Some of the cooling tubes can be seen at left.

Fig. 9-Revised Cupola Design.

and 16 lb fused soda ash. With these materials the first iron tapped contained between 0.071 and 0.093 per cent sulphur. The next combination consisted of 200 lb of carbide mixed IN with the portion of the bed coke extending upward from the tuyeres about 37 in., along with 700 lb of dolomite and 45 lb of spar ON the bed. Apparently increasing the amount of carbide and placing it down in the coke bed was quite effective because under these conditions, the first iron contained from 0.039 to 0.054 per cent sulphur.

In an attempt to reduce the sulphur content of the first iron still further the same quantity of flux materials as described in the previous test was used. However, the carbide was concentrated in the portion of the bed coke extending upward from the tuyeres about 18 in., but this did not bring about further reduction in the sulphur content of the first iron tapped.

The type or quality of coke used will in all prob-



Close-up of part of water header ring and connection to lower end of one of the cooling tubes extending through the shell. This is shown just to left of center. Note water spray on shell and "A" shaped angle to keep water out of trough.

Fig. 8-Revised Cupola Design.



Inside of cupola after 65 hr continuous service. Cooling tubes and projecting water-cooled copper tuyeres can be plainly seen.

Fig. 10-Revised Cupola Design.

ability have a definite bearing on carbon control. Results presented elsewhere in this paper report the effect of changing the quantity of coke per charge, but to date only one type has been used to melt most of the iron produced in the basic cupola in question and for this reason it will be impossible to report on the effects of other types at this time. The coke which has been used is a good quality beehive type with average analysis as shown in Table 5.

TABLE 5-COKE ANALYSIS

Fixed Carbon, %93.5	CaO, % 0.32
Volatile, % 0.9	MgO, % 0.094
Sulphur, % 0.54	Al ₂ O ₈ 1.52
Ash, % 4.76	Fe ₂ O ₃ 0.50
SiO ₂ , % 1.93	P ₂ O ₅ 0.008

In attempting to control the carbon content of the iron, it was felt that lower carbons might be obtained by reducing the slag volume. This will reduce the basicity of the slag by virtue of a reduction in the percentages of CaO and MgO present, while the amount of SiO₂ absorbed from the coke ash remains practically constant. Minor variations in oxidation of silicon during melting will have little effect on slag basicity while any marked increase in the amount of silicon oxidized will, of course, reduce the basicity still further. In order to determine the effect of slag volume, the quantity of flux materials used during the course of Heat 10351 was changed as shown in Table 6. All other details of the operation remained the same.

Two separate and distinct 5-hr periods within the heat just mentioned were selected for purposes of comparison. The first period includes results obtained from the fourth to eighth hours inclusive, while the second period includes results from the sixteenth to twentieth hours inclusive. Results obtained during the first three hours of operation have been deliberately omitted because it is felt that "normal" or "equilibrium" conditions may not exist within the cupola at the start. Immediately after the eighth hour of operation, the flux charge was changed to the quantities shown in Table 6 under column headed "9th–20th Hrs. Incl." Thus, the 7-hour interval between periods has provided ample time for the elimination of any "carry-over" effect of the flux charge

TABLE 6-FLUX PER ONE-TON METAL CHARGE

Heat No. 10351							
Materials	4th-8th	Hrs.	Incl.	9th-20th	Hrs. Incl.		
Dolomite		1b			75 lb		
Carbide	10	lb .			None		
Spar	15	lb .			10 lb		
*CaO Charged .							
*MgO Charged							
	s based on resi						
used which indi							
(CaO) and 209							
mercial calcium							
the balance lime							
0.9 lb of lime							

looked upon as containing 90% CaO.

used at the beginning.

Results of silicon, sulphur, total carbon, and slag analyses, along with temperatures of iron at the cupola spout are given in Table 7 and Table 8, respectively. A comparison of average values in these tables indicates a silicon loss of 11.4 per cent at the beginning of the heat and a PICK-UP of 12.2 per cent at the end. (Silicon loss, or PICK-UP, is calculated by subtracting the average silicon in the iron at the cupola spout from the silicon in the charge and dividing this difference by the silicon in the charge. The silicon pick-up noted is probably the result of reduction to metallic silicon of some of the SiO₂ in the slag by carbon, and the silicon thus produced is dissolved in the molten iron. The reaction could be written:

$$SiO_2 + 2C = Si + 2CO$$

This reaction would normally be expected to proceed more rapidly as temperature increases and basicity of the slag decreases. Data recorded indicate that the conditions existing at the end of the heat were more favorable to reduction of SiO2 than in the early hours because the average temperature toward the end was 2889 F with a slag basicity of 1.4 as compared to an average temperature of 2794 F and a slag basicity of 1.6 at the beginning. Slag basicity was calculated by dividing the sum of the percentages of CaO + MgO by the sum of the percentages of SiO₂ + Al₂O₃. It is of interest to note that the iron at the start of the heat contained an average of 0.032 per cent sulphur, while the slag contained an average of 1.46 per cent sulphur. The iron melted at the end of the heat contained an average of 0.055 per cent sulphur and the slag contained an average of 1.85 per cent sulphur.

In spite of the higher percentage of sulphur in the slag at the end of the heat the sulphur content of the iron melted during this period was greater than at the start. However, it must be remembered that the quantity of flux used at the end was less

TABLE 7-BASIC CUPOLA HEAT No. 10351 4th-8th Hours of Operation, Incl.

	Cupola C	Charge				F	lux Charge			
Steel Se	rap		1400 lb		Dolomite 150 lb					
Cast Sc	ар		600 lb		Calcium Carbide 10 lb					
	Silvery—15% Si					spar				
, *	*Iron					Analysis	Cupola S	lag, %		
Si	S	T.C.	Temp, F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃	S	
0.82	0.033	4.13	2792							
0.93	0.031	3.88	2827	0.37	43.60	27.70	15.12	7.88	1.45	
1.15	0.032	4.08	2775							
0.95	0.039	3.98	2810	0.35						
1.22	0.035	4.07	2775							
0.98	0.020	4.25	2775	0.35						
0.96	0.027	4.14	2792							
1.13	0.028	4.05	2810	0.44						
1.01	0.046	3.89	2792							
0.95	0.029	4.08	2792	0.73	42.00	27.24	11.43	7.44	1.47	
Aver. 1.01	0.032	4.06	2794	0.45	42.80	27.47	13.28	7.66	1.46	
Silicon in charge, %       1.14         Aver. Silicon loss, %       11.4         Air Blast Temp       950 F					Aver.	Carbon p	ick-up, %			
			ron samples fro	om cupola						

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TABLE 8—BASIC CUPOLA HEAT No. 10351 16th-20th Hours of Operation, Incl.

	Cupola Charge           Steel Scrap					Flux Charge					
	Briquets	@ 2 lb Si •Iron		1 pc			Analysi	s Cupola	Slag. %		
	Si	S	T.C.	Temp, F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃	S	
	1.31	0.068	3.94	2845	0.44	37.04	29.84	19.02	10.89	1.68	
	1.34	0.056	4.06								
	1.21	0.059	4.10		0.51						
	1.29	0.050	4.26	2893							
	1.22	0.048	4.20		0.37						
	1.35	0.041	4.25								
	1.21	0.040	4.25		0.37	37.20	27.60	20.40	11.42	2.01	
	1.31	0.056	4.25	2878							
	1.32	0.069	3.91		0.40						
	1.17	0.053	4.18	2938							
					0.40			10.01			
Aver.		0.055	4.14	2889	0.42	37.12	28.72	19.71	11.16	1.85	
	Silicon in Charge, %								******		
									*******		
	Air Bla	st Temp								5.3:1	
			*Note: I	ron samples fro	om cupola	spout ever	y 30 minu	tes.			

than half of that used at the beginning and consequently there was an appreciable reduction in slag volume. While there may be a higher concentration of sulphur in the smaller volume, it is readily understandable that it has a smaller total capacity for sulphur simply because there is less slag present. This indicates that sulphur reduction is a function of slag volume as well as slag basicity.

The average total carbon in the first hours of the heat was 4.06 per cent as compared with 4.14 per cent in the last hours. This difference is not particularly great as the deviation from the average of these two figures is only plus or minus 0.04 per cent. However, the higher carbon level was obtained at the higher melting temperatures and this is what would normally be expected. It has been reported that an increase in sulphur content of the iron will be accompanied by a decrease in total carbon. However, this was not the case during the heat in question and based on results obtained it appears, within certain limits at least, that the carbon level may be dependent on other factors such as the quantity of FeO in the slag and the maintenance of strongly reducing conditions. That such conditions were maintained is indicated by the fact that the slag at the start of the heat contained an average of only 0.45 per cent FeO, while that at the end contained an average of only 0.42 per cent FeO.

The composition of the slag in the last hours of the heat was such that the attack on the basic refractories in the cupola trough and slag dam was so severe that the heat had to be discontinued about 3 hr ahead of schedule. Tests to determine the effects of still further reduction in the quantity of flux materials used will have to be left as the subject of a future investigation.

Attention is again called to the fact that the temperature of the iron at the end of the heat (average 2889 F) was 95 F higher than that of the iron at the start (average 2794 F). This increase in tempera-

ture might possibly be explained by a heat balance based on the following:

- *To calcine 1 lb of CaCO₃ or MgCO₃ requires 788 Btu.
- (2) *To form and superheat 1 lb of slag to 2800 F requires 900 Btu.
- (3) *Specific heat of cast iron is 0.23 Btu per lb of per°F.
- (4) The dolomite used contains 30 per cent CaO and 20 per cent MgO which is equivalent to 95 per cent total carbonates.
- (5) Dolomite used, 4th-8th hours inclusive, was 150 lb per ton of iron.
- (6) Slag formed, 4th-8th hours inclusive, was calculated to be 131.6 lb per ton of iron.
- (7) Dolomite used, 16th–20th hours inclusive, was 75 lb per ton of iron.
- (8) Slag formed, 16th-20th hours inclusive, was calculated to be 70.6 lb per ton of iron.

 These values taken from Handbook of Cupola Operation, 1st Edition, Pages 308, 314, and 353 respectively.

HEAT REQUIRED TO CALCINE STONE AND FORM SLAG

112.290	Bru
,	
118,440	Btu
230,730	Btu
56.145	Btu
63,540	But
119,685	Btu
111.045	Btu
460	Btu
	230,730 56,145 63,540 119,685 111,045

At 100% efficiency the extra heat available per ton of iron-16th-20th hours, inclusive—would raise the temperature of the iron  $111,045 \div 460$  or  $241^{\circ}F$ . The actual increase in temperature of 95 F represents an efficiency of 39.4 per cent which is reasonable for a hot blast cupola.

There is some evidence that carbon control might be achieved through variations in the cupola air blast temperature, and as might be expected the carbon level will be directly proportional to blast temperature. On one occasion, an interruption in the operation of the blast heater occured between 6:00 A.M and 9:00 A.M. During this period the blast temperature fell from 950 F to a low of 170 F at 7:20 A.M. and then returned to 950 F. Samples of iron for chemical analysis were taken at 30-min intervals and results of analysis of the iron and slag as well as tapping temperatures are recorded in Table 9.

TABLE 9-EFFECT OF AIR BLAST TEMPERATURE Heat No. 10551

		*Iron			Analysis Cupola Slag, %					
Time	%Si	%S	%T.C.	Гетр, F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃	
5:30	1.14	0.024	4.10	2845						
6:00	0.91	0.044	3.89	2827	0.50	42.20	28.60	16.60	8.70	
6:30	0.79	0.036	4.03	2792						
7:00	1.08	0.042	3.83	2792	0.64					
7:30	1.19	0.057	3.52	2757						
8:00	0.84	0.054	3.74	2845	1.30					
8:30	0.60	0.067	3.63	2878						
9:00	0.83	0.043	4.10	2878	1.06	41.80	28.00	14.60	9.55	
9:30	0.85	0.038	4.36	2862						
10:00	1.29	0.032	4.32	2878	0.80					

Results indicate rather poor uniformity of chemical composition from one sample to the next, but it must be remembered that the samples were taken from the stream of iron flowing in the cupola trough where no mixing was possible. In spite of this there is a definite trend toward lower silicons, total carbons, and iron temperatures along with an increase in both the sulphur content of the iron and the FeO content of the slag, as the cupola air blast approached the minimum of 170 F. Once the blast temperature

returned to 950 F. The trends just mentioned were reversed and the operation returned to normal. In-asmuch as only a small amount of data is available covering this phase of carbon control, further comment will be withheld at this time.

Another approach to carbon control would be through adjustment of the amount of coke used per charge. A comparison of results obtained during Heat No. 112051 with those obtained during Heat No. 103151 are given in Table 10 and Table 11 respectively, and indicate the effect of reducing the amount of coke, all other things remaining the same. In both cases the mixture contained 70 per cent steel scrap, sufficient silvery (15 per cent grade) to bring the silicon in the charge to 1.17 per cent the balance being cast scrap. The flux charge consisted of 150 lb dolomite, 10 lb calcium carbide (3½-in. x 2-in. lump), and 15 lb spar per ton, and the air blast temperature was 950 F.

During Heat No. 112051 the coke per 2,000 lb metal charge was 350 lb, while 275 lb of coke per 2000 lb metal charge were used in connection with Heat No. 103151. The most significant figures in the tabulations of results are those representing the amount of FeO in the slag. Reducing the amount of coke has created a condition more favorable to oxidation and this is clearly indicated by a comparison of the percentages of FeO in the slags. As a result of the condition just mentioned the silicon loss has increased from 11.9 per cent using the lower coke ratio to 48.8 per cent with the higher coke ratio. The high silicon loss lowers basicity of the slag, reduces its ability to remove sulphur and consequently produces iron with higher sulphur and lower carbon. The actual difference in silicon content of the iron at the cupola spout is 0.34 per cent or 6.8 lb of silicon per ton of iron. Replacing this quantity of silicon in the form of 75 per cent ferrosilicon will cost about \$1.42 per ton of iron which is considerably more than the value of the additional 75 lb of coke used.

In cupola melting the principal sources of sulphur found in the iron are from the metallic components of the charge plus that in the coke. Assuming that the coke contains 0.5 per cent S, that all the sulphur

TABLE 10-HEAT No. 112051 WITH 70% STEEL 350 lb coke per ton

		*Iron				Analysis Cupola Slag, %				
	Si	S	T.C.	Temp, F	FeO	CaO	SiO ₂	MgO	Al ₂ O ₃	
	1.11	0.038	4.07	2722	0.44					
	0.93	0.036	4.18	2757						
	0.95	0.033	4.13	2792	0.36					
(	0.93	0.025	4.37	2775						
	1.08	0.019	4.31	2775	0.30	42.35	29.34	15.10	8.92	
(	0.98	0.030	4.16	2775						
	1.25	0.031	4.20	2792	0.30					
	0.94	0.023	4.34	2775						
	1.20	0.024	4.36	2740	0.26	42.35	28.68	14.38	8.42	
-	0.79	0.026	4.33	2792						
-	_								-	
Aver. 1	1.02	0.029	4.25	2770	0.33	42.35	29.01	14.74	8.67	
		n Charge, % . licon Loss, % .				Aver. Ca	in Charge, % arbon Pick-up			

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in the coke between charges is absorbed by the iron, none is absorbed from the bed coke and that there is no melting loss (or gain), it could be reasoned that each 100 lb of coke per ton would result in an increase of 0.025 per cent S. Based on these assumptions, it is evident that the sulphur in the charge (S in metal plus S in coke) can be lowered by reducing the amount of coke between charges. It could be reasoned further that with lower sulphur in the charge, a reduction in the amount of sulphur and an increase in carbon in the iron at the cupola spout might be achieved. However, attempts to control carbon in this manner may account for unsatisfactory operation as indicated by results in Table 11.

For long heats melting charges of a low carbon content (generally high percentages of steel), enough coke must be used to maintain the highly reducing conditions necessary for good sulphur reduction and high carbon pick-up, as previously mentioned. In addition to being an excellent fuel, incandescent coke is a powerful reducing agent, and this property should not be overlooked, especially in connection with the operation of a basic cupola. Furthermore, attention is called to the fact that the carbon pick-up during the melting of charges containing from 70 to 100 per cent steel is between 2.7 and 3.7 per cent. This is equivalent to 54 to 74 lb of carbon per ton which in turn represents from 60 to 80 lb of coke

containing 90 per cent carbon. The coke which is used for carburization of the charge, is, of course, not available for melting and superheating the iron, but must be taken into account when coke ratios are being considered.

There have been no difficulties encountered when melting charges containing 100 per cent steel scrap provided that a sufficient amount of coke is used. Results obtained in connection with such a heat are given in Table 12. In this connection a silicon loss of 46.7 per cent, is reported but quantitatively this represents only 2.8 lb of silicon per ton of iron. From this standpoint the loss is not particularly serious as 2.8 lb of silicon will form only 6 lb of SiO₂. This will have relatively little effect on the composition of the quantity of slag involved. It is important, however, to see to it that the number of pounds of silicon burned to SiO₂ be held to the lowest practical minimum.

In view of the cost of labor and materials to make repairs it has been our objective to operate the basic cupola continuously for as many hours as possible. After modifying the cupola design and using the breast construction shown in Fig. 4 it became possible to operate continuously for 24-hr periods as previously stated. Examination of the lining after these heats indicated relatively severe attack on the refractories between the tuyeres and in the well at the slag line.

TABLE 11-HEAT NO. 103151 WITH 70% STEEL 275 lb Coke Per Ton

	•Iron					Analysis Cup	ola Slag, %	
Si	S	T.C.	Temp, F	FeO	CaO	SiO ₉	MgO	Al _a O _a
0.73	0.052	3.80	2810					
0.73	0.054	3.86	2810	1.26				
0.75	0.055	3.60	2845					*
0.70	0.050	3.71	2810	0.84	39.20	30.38	12.25	6.03
0.62	0.048	3.79	2827					
0.80	0.048	3.63	2810	0.90				
0.51	0.057	3.46	2810					
0.77	0.058	3.48	2810	1.28				
0.39	0.058	3.46	2775					
0.79	0.047	3.87	2862	0.76	39.60	29.56	13.40	7.34
ver. 0.68	0.053	3.67	2817	1.01	39.40	29.97	12.83	6.68
	n Charge, % .		1.17		Carbon	in Charge, %		1.24
	licon Loss, %					rbon Pick-up		
	70		on samples from	cupola eno			10	

TABLE 12-HEAT No. 102451 WITH 100% STEEL

			Flux Cha	arge Per T	on				
			Dolomite		160 lb				
			Carbide		13 lb				
			Spar		13 lb				
	*Iron					Analysis	Cupola S	lag, %	
Si	S	T.C.	Temp, F	FeO	CaO	SiO2	MgO	Al ₂ O ₃	S
0.17	0.037	3.98	2862	0.70					1.14
0.19	0.035	4.03	2827		*				
0.16	0.032	4.10	2827	0.60	44.80	24.20	12.24	10.30	1.24
0.13	0.028	4.03	2792						
ver. 0.16	0.033	4.04	2827	0.65	44.80	24.20	12.24	10.30	1.19
Silicon	in Charge, %		0.3		Carbo	on in charg	ge, %	0	.30
	Silicon Loss, %				Aver.	Carbon pi	ick-up, %	3	.74
	ast Temp							5.	
			on samples from	3- ton mi	ver every 3	0 minutes.			

To prevent failures in these areas and to make it possible to further extend the duration of the heats, two annular spray pipes, 3/4 in. in diameter, were fitted to the outside of the cupola shell. See Fig. 5. The upper spray pipe provides cooling between the tuyeres and the lower one provides a continuous curtain of water on the shell from the lower side of the tuyeres downward to a level just below the top of the sand bottom. These sprays were, found to be quite effective and made it possible to keep the cupola in continuous service for periods of 65 to 85 hr.

During one 84-hr run the total "down-time" was about 9 hr occasioned by the necessity of replacing some of the brick in the cupola trough. Before the next heat, a monolithic carbon lining backed with magnesite brick was rammed in the trough, thus making it possible to operate 121 hr including a total "down-time" of 10 hr. The plastic carbon material was used in the next heat to form the entire breast and trough lining, rammed as a single monolithic unit. With this arrangement the cupola was operated continuously for 135 hr, with a total of only 2 hr "down-time." Preliminary tests indicate that the carbon plastic can be used to advantage for lining the portion of the well extending from the top of the sand bottom upward for a distance of 12 to 16 in. In this manner it has been possible to keep the cupola in continuous service for periods of 168 hr.

#### Banking the Cupola

When necessary to suspend operations, it has been found possible to "bank" the cupola for extended periods of time. Immediately after the last metal charge is put into the cupola, sufficient coke is added so that the coke bed is about 30 in. high and thoroughly ignited when all of the iron has been melted. Care must be taken to melt all of the flux remaining in the cupola and remove all of the slag formed. If this is not done there is danger of forming a bridge at the level of the water-cooled tuyeres which is one of the coldest zones in the cupola during shut-down periods. The last traces of molten slag and iron are removed by opening the slag spout drain. The tap hole is then botted up at once, just as is done at the start of the heat. Prior to resumption of operations, the coke bed is built up to a height of 56 in. above the tuyeres, and after it has become thoroughly ignited charging and melting start just as if it were the beginning of another heat.

The consumption of refractory materials used in connection with a heat during which 1281 tons of iron were melted is as follows:

Dolomitic patching material, 6000 lb ..... or 4.7 lb per ton Basic brick 1925 lb plus 300 lb mortar or ..... 1.7 lb per ton Carbon ramming mix, 1500 lb, or ..... 1.2 lb per ton

All good practices followed in connection with the operation of acid cupolas must be adhered to when operating a basic cupola, and careful supervision is required at all times. The iron yard must be kept in first class order from the standpoint of housekeeping, and only materials reasonably free from adhering sand should be charged. In wet weather it is quite easy to charge large quantities of SiO₂ into the cupola in the form of mud adhering to the metal. Other

raw materials should be stored and handled so as not to become contaminated with foreign materials which frequently contain high percentages of silica. For example, forking coke from the ground or shoveling limestone from bins with "dirt" floors is very poor practice. In the basic cupola, silica is the least wanted material and at the same time one which is generally found in great abundance on the iron yard. It can easily be charged in unknown quantities and/or from unsuspected sources. Eventually it finds its way into the slag and may completely upset the balance of slag chemistry.

#### Summary

- 1. The basic-lined, water-cooled cupola with hot blast has been used successfully for producing low-sulphur, high-carbon irons from mixtures containing up to 100 per cent steel. These irons are well suited to the production of ductile iron, can be used as a substitute for purchased pig iron, and can be poured directly into castings where high carbon irons are known to be desirable.
- 2. A tube-type water cooling system replaces most of the refractories in the melting zone, thus preventing erosion and making it possible to operate continuously for long periods of time. Best results have been obtained with the tubes parallel with the cupola shell rather than arranged in the form of truncated cone with the small diameter at the bottom. The water-cooled copper tuyeres protruding into the cupola beyond the surface of the lining in the melting zone, have definitely contributed to the success of the operation.
- 3. An externally-fired air heater having a capacity of 13,000 cfm at a maximum temperature of 1000 F is used to preheat the cupola blast. Hot blast minimizes silicon losses and increases carbon pick-up during melting. The "by-pass" feature built into the heater provides hot air at a predetermined temperature from the very start of the heat.
- 4. At first, considerable difficulty was experienced due to failure of the refractories in the tap hole used for front slagging. However, the use of a combination of magnesite and carbon brick made it possible to operate continuously for periods of 24 hr.
- 5. By using the proper amounts of fluxing materials IN and ON the coke bed it is possible to tap iron at the start of the heat with a maximum of 0.06 per cent sulphur.
- 6. Reducing the air blast temperature is accompanied by a trend toward lower silicons, total carbons, and iron temperatures along with an increase in both the sulphur in the iron and the FeO in the slag.

7. When melting charges of low carbon content, carburization of the iron may consume as much as 80 lb of coke per ton. For this reason it has not been found possible, in long heats, to use lean coke ratios

when a high carbon pick-up is desired.

8. External water cooling of the shell between and below the tuyeres has made it possible to operate continuously for periods of 65 to 85 hr. Along with this cooling the use of a rammed monolithic carbon refractory for the breast and trough lining has ex-

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tended the duration of the heats to 168 hr.

9. By following a very simple procedure, the cupola has been held-over or "banked" for as long as 24 hr and the operation resumed at the end of that

period with no difficulties whatever.

10. Careful supervision is required at all times and the importance of good housekeeping on the iron yard, in order to exclude as much "raw" SiO₂ from the cupola charges as possible, cannot be over emphasized.

Acknowledgments

The author wishes to thank E. S. Renshaw of Ford Motor Company, Dagenham, England, and G. Vennerholm of Ford Motor Company, Dearborn, Michigan for valuable advice given during numerous discussions concerning the design and operation of basic cupolas; Max Kuniansky and T. W. Curry of Lynchburg Foundry Company for their suggestions regarding arrangement of the material in the manuscript; J. L. Graham, Special Experimental Technician, Radford Laboratory for obtaining photographs and reviewing the manuscript; and E. K. Riddle, College Trainee, for providing the drawings.

#### DISCUSSION

Chairman: J. D. SHELEY, The Black Clawson Co., Hamilton, Ohio.

Co-Chairman: R. A. CLARK, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

Recorder: W. T. BOURKE, American Brake Shoe Co., Mahwah,

N. J.

A. L. DE SY (Written Discussion): 1 The author of this paper should be congratulated for his valuable contribution to the practical knowledge of basic and water-cooled cupola melting. A great deal of the interest of this paper results from detailed and accurate description of the difficulties encountered in the early stage of development.

In reading the paper and comparing the results obtained with this very big cupola to those obtained in our laboratory with a very small (10-in.) cupola similar to the unit in operation at Battelle Memorial Institute, it is remarkable to note

that the results are very similar.

The reducing conditions thus appears to be nearly the same: compensation for the hot blast applied in the big cupola is obtained with a larger amount of carbide in the flux charge of

the small cupola.

One detail however does not appear very clear. In comparing the results given in Tables 4 and 7 it seems difficult to find the reason for the great difference in the average sulphur: 0.014 (Table 4) and 0.032 (Table 7) specially as the average FeO slag content is respectively 0.75 and 0.45.

The author's comments to this remark would be highly appre-

ciated.

MR. LEVI (Reply to Mr. De Sy): Mr. De Sy's question is a good one and one which I am not sure that I can answer satisfactorily. However, I would like to call attention to the fact that 50 lb of carbide were used in the cupola charge given in Table 4 while only 10 lb of carbide were used in the charge reported in Table 7. The basicity of the slag and the temperature of the iron at the cupola spout as reported in Table 4 are both higher than the corresponding values reported in Table 7. Furthermore, it is well known that slag fluidity increases with temperature and becomes more active at the higher temperatures. What has just been said may be a possible explanation for the difference in the sulphur content of the two irons in question, in spite of the relative percentages of FeO in the two slags involved.

S. F. Carter (Written Discussion): The author and his company are to be congratulated on their courage to try such a combination of promising cupola developments; basic slag, water

cooling, continuous operation, and a highly preheated blast.

This company has utilized the chemical potentialities of a basic slag for a number of applications; higher carbon gray iron, iron more suitable for nodular treatment, and profitable utiliza-

tion of locally available steel and cast scrap.

I would like to repeat the author's warning that this cupola combines an unusual combination of conditions. The high temperature blast is very important and must be included in descriptions of this cupola. We have not operated at 850 or 950 F blast but found a 450 blast very powerful on basic operation. Temperature jumped from 2750 F to 2900 F, carbon jumped from 3.40 to 4.00 per cent, sulphur dropped from 0.060 to 0.030 per cent. To bring the chemistry in line for gray iron it was necessary to raise the steel and cut the coke and flux substantially. We too have found that preheated blast reduces oxidation, which enables a basic cupola to be more basic and chemically more powerful in every respect. Experiences at this high blast temperature cannot be translated to cold blast operation without liberal allowances for the effect of blast temperature.

I have four brief questions. The bee-hive coke is unusual in many respects. Has the author tried a more conventional byproduct coke in the basic cupola? If so, what about comparative chemistry? How does the bee-hive coke compare in an acid

cupola?

The steel scrap used in this cupola has been observed to be generally very light. Has any significant variation in chemistry been associated with physical variations in steel scrap, or does the hot blast compensate for the usual effects?

Has the author had a chance to vary the projection of the tuyeres, as planned, to determine any effect on chemistry and

performance?

The first heat, No. 3251 in Table 2, seems rather ineffective, from the basic standpoint, and not entirely accounted for by the coke ratio. With only 25 per cent steel and 15 per cent flux we would expect carbons higher than 2.98 per cent and sulphurs lower than 0.104 per cent with cold blast. The original design does not show such a drastic trend on the next heat (in Table 3), and similar reductions in coke in Tables 10 and 11 did not produce such a drastic effect. I wonder if this heat might not have experienced some irregularity not noted. The high slag FeO, the abrupt variations in slag FeO, and the ups and downs in metal chemistry seem to suggest some irregular conditions within the cupola that might have produced excessive oxidation.

MR. Levi (Reply to Mr. Carter): The average chemical analysis of the coke referred to by Mr. Carter is given in Table 5. At the time the paper was presented we had no data using by-product coke with the revised cupola design. However, on July 14, 1952 the cupola was used to melt 100 per cent steel with by-product coke and the results obtained compare favorably with those obtained using bee-hive. The charge used consisted of 1500 lb steel scrap, 100 lb limestone, 30 lb carbide, 30 lb spar, and 260 lb of by-product coke.

When using less basic slags than those employed in melting 100 per cent steel charges it has been found necessary to add some carbide and spar to the charges when using by-product coke in order to obtain results comparable to those obtained when using bee-hive coke. Materials used in two typical charges are tabulated below:

Heat No.	61752		72952
Cast Scrap	 .2200 lb	-	.2200 lb
Steel Scrap			
Silvery (15% Si)	 . 75		. 40
Limestone			
Dolomite	 . 65		. 65
Carbine	 None		. 15
Spar	 None		. 10
BEE-HIVE COKE	 . 350		. None
BY-PRODUCT COKE	 None		. 350

Chemical composition of the metal produced using the above mixtures was within the following ranges:

Type Coke	% Si	- % S	% T.C.
BEE-HIVE	1.35/1.41	.066/.095	3.54/3.65
BV-PRODUCT -	1.43/1.57	.082 / .093	3.57/3.65

Our experience with Bee-Hive coke in acid practice indicates that it produces irons with higher total carbon than when using by-product coke in the same proportions.

 ¹ Professor, Laboratorium Voor Metallurgie, Gent, Belgium.
 ² Melting Supt., American Cast Iron Pipe Co., Birmingham, Ala.

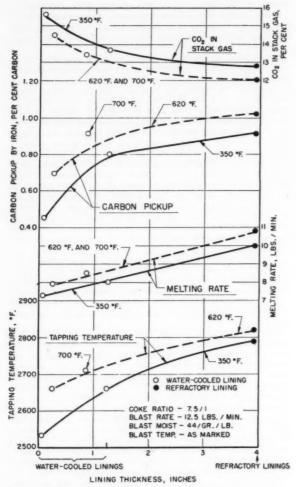


Fig. 1x—Effects of Lining Thickness upon the Operation of a 10-in. cupola.

As indicated by Mr. Carter most of the steel scrap melted in our basic cupola has been of very thin section. On one occasion when melting 100 per cent steel we used a combination of railroad spikes and structural shapes. With the heavier steel scrap no significant variation in chemistry was noted, though, there may be a difference in the percentages of tramp elements present.

To date, we have not had an opportunity to vary the projection of the water-cooled tuyeres. However, there are indications that the projection can be reduced and we feel further that the tuyere size can also be reduced. We are planning to make changes just mentioned but cannot of course report on the effects at this time.

Mr. Carter has asked if there might not have been some irregularities in connection with the results reported in Table 2, Heat No. 3251. Results obtained in connection with this heat are typical of what was experienced in connection with many others run under the same conditions, namely, the original cupola design and an insufficiently strongly reducing condition. We too, feel with as much as 2.63 per cent carbon in the charge that the iron at the cupola spout should contain more than 2.98 per cent T.C. even with acid practice and cold blast. This is mentioned by implication in the last line of the first paragraph on page 743 which states, "Based on these findings—in the basic cupola."

While it may not be Mr. Carter's intention to do so, I do not feel that results in Table 2 can be compared with those in Table 3. My reason for this is because of the difference in coke ratios, temperature of the iron at the cupola spout, and the use of

carbide in one instance and not in the other. I feel that the less drastic results reported in Tables 10 and 11 are attributable almost entirely to the benefits of the revised cupola design.

D. E. Krause (Written Discussion): ^a Although a considerable amount of information has been published on basic-lined cupola operation, the authors seldom mentioned some of the unsatisfactory results obtained during the early experiences with such an operation. The author of the present paper is to be complimented on bringing attention to some of the difficulties encountered in getting a cupola of new design into successful operation and how they were overcome.

One factor which has been properly emphasized is the necessity of obtaining a low FeO in the slag and high tapping temperatures if low sulphur contents and high carbon contents are desired. One of the disappointing features of the operation is the need for a rather rich coke ratio in spite of a blast temperature of 950 F. The extensive use of water cooling in the combustion zone is undoubtedly responsible for part of the need for extra coke. The installation of water-cooled tuyeres which protrude beyond the wall in the combustion zone seems to be a step in the right direction. It has been observed in a number of internally water-cooled cupolas that a secondary combustion zone is often established above the water coolers. Any condition which would make it difficult for air to pass through the combustion zone along the water-cooled surface should minimize this difficulty. The protruding tuyeres as installed in the revised design should therefore be helpful from this standpoint as pointed out by the author.

The advantage in operating a cupola for a period of five days without dropping the bottom from the standpoint of refractory costs is worthy of consideration of any foundry which operates

cupolas 16 hours daily.

It is felt that a word of caution should be added to the statement that the low-sulphur, high-carbon iron produced may be used as a substitute for pig iron. This might only be true if great care is exercised in selection of the scrap steel used in the charge. If the scrap steel carries any tramp elements, they will most likely be present in the iron tapped from the cupola in spite of the carbon being high and sulphur being low. The iron produced from a high steel mix in the manner described may have a higher nitrogen content than pig iron obtained from a blast furnace.



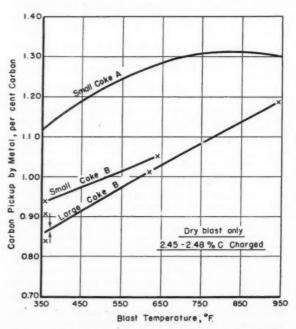


Fig. 2x—Effects of Blast Temperature on Carbon Pickup by Metal for Three Test Cokes in 10-in. Cupola at Coke Ratio of 7.5/1 and Dry Blast Rate of 12.5 lb/min.

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In connection with an investigation of cupola slags carried out by the Gray Iron Research Institute, Inc. in a 10-in. cupola, some work was done with a water-cooled lining. The detrimental influence of water cooling on the temperature of the operation is shown in Fig. 1x.

The points at the right hand side of Fig. 1x are for a conventional lining without the water cooling. Water coolers were installed with a 1½-in. fireclay brick against the face of the cooler. The effect of gradually melting away the brick is indicated by the points on the left side of the chart. Although increasing the blast temperature increases the tapping temperature, it does not eliminate the effect of water cooling. It must be borne in mind that the effect of water cooling the combustion zone has a much more pronounced effect in a 10-in. cupola than in a 72-in. cupola.

It might be of interest to show the influence of blast temperature on carbon pickup and silicon losses obtained with a 10-in. cupola since this effect in the 72-in. cupola was briefly referred to in Table 9. Figure 2x shows the effect of blast temperature on carbon pickup in a 10-in. acid-lined cupola for two grades of coke. A similar effect would be obtained for a basic-lined cupola although the curves would be higher on the chart because of the inherently higher carbon pickups obtained with basic slags.

The influence of blast temperature on silicon loss in an acidlined 10-in. cupola is shown in Fig. 3x. Data for Fig. 3x were obtained from the same heats as the data for Fig. 2x. Attention is called to a silicon pickup with the higher blast temperatures. The silicon pickup is also associated with the higher tapping temperatures, low silicoa losses, and under proper conditions, high carbon pickup.

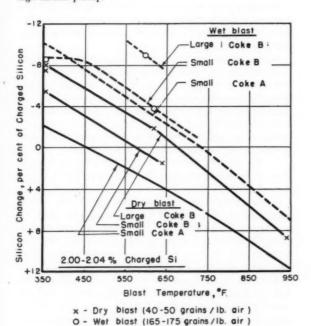


Fig. 3x—Effects of Blast Temperature and Blast Moisture upon Silicon Change in the Metal Obtained with Three Test Cokes in a 10-in. Cupola at Coke Ratio of 7.5/1 and Blast Rate of 12.5 lb/min.

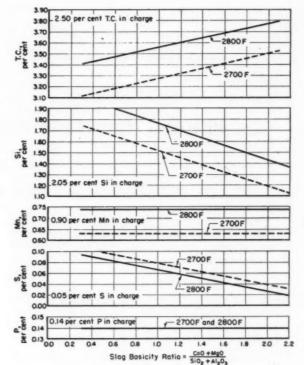


Fig. 4x—Effects of Slag Basicity and Tapping Temperature on Composition of Iron for Cupola Charges containing 35 per cent Steel (Excludes Carbon Linings and Calcium Carbide).

Although the data in Fig. 3x are for an acid-lined cupola, increasing the basicity of the slags will shift the curves upwards to higher silicon losses. This effect of a change in slag basicity is shown in Fig. 4x. Figure 4x also shows the influence of slag basicity on the sulphur, carbon, phosphorus, and manganese contents of the iron at two tapping temperatures. The curves were drawn from data computed from a considerable number of 10-in. cupola heats with constant charge composition. The marked effect of operating temperature on some of the factors is evident from inspection of Fig. 4x.

MR. LEVI (Reply to Mr. Krause): While Mr. Krause does not ask any direct questions in his discussion, his comments are generally in line with our thoughts on the items he has mentioned. His warning that low-sulphur, high carbon irons produced in the basic cupola and later used as a substitute for pig iron, may contain tramp elements is a situation of which we are well aware. However, at the present writing we have had very satisfactory results in one of our foundries where 100 per cent of the pig iron in the charge has been replaced by the low-sulphur, high-carbon iron from the basic cupola. The cupola charge in question contains 20 per cent pig iron and 30 per cent steel scrap. It is my opinion that the data which Mr. Krause has presented in the form of charts and graphs are a valuable contribution to the literature on cupola operations—both acid

### PRICING CASTINGS USING STANDARD COSTS

By

Jeff A. Westover*

The discussion which follows deals specifically with a method which can be turned over to the head of the Standards Dept. or anyone faced with the problem of pricing castings, to follow as a pattern for doing the costing work in a plant. We are attempting to give a comprehensive picture of a method that can be used which will take a job from one extreme to the other to provide a cost that is accurate, as it takes into consideration all of the different factors.

Pricing Castings. One of the major problems of many foundry managements is determining how much to charge for a casting. Pricing of castings is too often based on past experience, on what competition quoted, on an outright guess, or on some figure which suited management for some reason known only to management.

Quoting too high a price may lose business; quoting too low a price, operations may show a loss; both are equally bad and reflect not only on the company making the castings, but the castings industry as a whole. A system that will consistently price a variety of castings in relative position one with another, is not only desirable, but will prove to be a good pricing system, mutually beneficial to invested capital in the business, the company's employees, and the customers served.

In presenting this discussion of pricing castings and the use of various forms in establishing individual casting costs, we refer to the procedure as Standard Costing.

What Are Standard Costs? Stated simply, Standard Costs are pre-determined costs which are established for materials, labor and expenses; their objective being economical production at a profit.

The use of a Standard Cost plan is not one that is beyond the reach of the average and very small company. It does not mean that a company must operate with an incentive or measured work plan, even though this is highly desirable as a means of more accurate control of costs. The costing of jobs will follow the same procedure, regardless of the

plan being followed, in compensating workers for their production, be it day or hourly work; piece work or incentive work. The latter two plans of payment for production do, of course, afford a control of the labor cost, whereas day or hourly work may vary considerably, dependent on the accuracy of the estimate of time required to do the job and the performance of the operator assigned to the job.

Helpful, but not entirely essential to this plan of costing, is the necessity for the company to maintain an accounting system that will allot and record various items of cost,

In a manufacturing business as complex as a foundry, involving as many operations and functions as it does, it is necessary that a division be made of the foundry into Departments and Cost Centers. Usually the two words, Department and Cost Center, are used interchangeably and considered to mean the same. It should, however, be helpful in our discussion to clarify our use of the two terms.

A Department means a major division of the process of production, that is, Molding, Coremaking, Cleaning, etc.

A Cost Center is a distinct spot within the department against which costs are accumulated. That is, in the Molding Department we have as Cost Centers, Floor Molding, Squeezer Molding, etc. In the Core Department we have Bench Coremaking, Core Blowing, Sand Mixing, etc.

It is not our purpose to go into details of accounting for Standard Costs, as this subject was well covered by Fred Ruffolo in A.F.S. Transactions, vol. 57, pp. 415-432 (1949) in a paper entitled "Approach to Standard Costs in the Foundry."

Since we look to accounting for certain information only, we are, in effect, separating the process of job estimating and costing from accounting. In order that the costing and pricing department or individual can be consistent in the practice of estimating or pricing a casting, it will be necessary to use set forms. Set forms act as a check list of elements of cost and when completely filled out leave little chance of error. Information entered on these forms may come from several sources, dependent on the methods of operations and availability of information, i.e., time

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keeping, supervisor estimating, Standards Dept. from Direct Time Study or from Standard Data.

The proof of the accuracy of the estimating and costing work will be reflected in the Profit and Loss statement prepared by the accounting department.

#### **Essentials of Suitable Pricing Methods**

The essentials of a suitable pricing method consist of (1), a definite form which acts as a check list and upon which is recorded each and every operation going into the making of a casting; (2), provisions for easy revision of conversion figures to reflect changes in cost of labor, materials, etc., for each cost center; (3), maintaining a consistent relationship between selling price and actual cost, which in turn makes possible reasonable comparison of the cost of one casting to that of another.

In costing castings it is not necessary that a shop have Time Standards set from Standard Data or Direct Time Study for any or all of their operations. However, Time Standards set for incentive purposes or measured work plans, will produce greater accuracy and a more consistent selling price than if experience estimates for determining the time that will probably be taken or allowed for doing each of the different operations in producing a casting are used. Therefore, the method to be described is of as much use to a shop which operates entirely on a Day Work pay basis as one which covers operations with Time Measured Standards. It has proved to be practical in the very small foundry of only two molders, as well as in the large shop.

The number of employees in the shop, the variation in the type of work, equipment, and the methods used, are all taken into consideration and the variations caused by each of them is reflected in the ultimate selling price answer that is arrived at in pricing the job.

There are four principal forms to be used in making the Cost Analysis—

- 1. The Estimate-Cost and Pricing Form
- 2. The Cost Center-Conversion Figure Analysis
  Form
- 3. The Schedule of Depreciation to Cost Center Form
- 4. The Description-Overhead Items of Cost Form

Additional forms would be used to establish the time required for doing a job, providing the company was using elemental time standards for measured day work or for incentive pay purposes. These forms would be the job analysis breakdown or specification for each operation and time values entered thereon would come from Standard Data.

When this latter form is not used, then the time for doing any part of the job must be arrived at through direct time study, estimate or by use of job time clocks. Just as Standard Data can serve to establish time for doing a job, so, too, can tables of values for materials be very helpful in working up cost figures.

The extent of breakdown of costs depends greatly on the desires of Management for detail. However, in the end you come up with one figure—the selling

FORM #197	DATE-BY	EST	ACT	0
		MEAS. /CSTG.	CSTG.	COOLOME
EACH CASTING NET V	WEIGHT	lb.		118
	IELD %@	\$	\$	N 2
COREMAKING	@ \$ .0589	m,		1
MATERIAL	.0080	lb.		1
CORE FINISHING	. 0589	m,		
MOLDING	. 0657	m.		
MATERIAL M	old Indirect			
	old Indirect		-	ı
SHAKEOUT M	old Indirect			I
SAND CONDITION M	old Indirect			
SAW	. 0753	m,		
STAND GRIND - SNAG	. 0599	m.		1
STAND GRIND - TRIM	. 0599	m.		
CHIP AND HAND GRIN		m.		
BELT SANDER	. 0620	m.		
DISC SANDER	. 0692	m.		
SANDBLAST	. 0686	m.		
HEAT TREATING	. 1500	lb.		
CHEM. & PHYSICAL T	ESTS . 4000	lb.		FALL
SUBTOTAL PE	R CASTING		5	I. NO
JOB RISK			-	1.
ADMIN , SELLING & P	ROFIT			
SET-UP				
_1	PRICE PER	CASTING	\$	
	,	V		

Fig. 1

price of the casting.

To illustrate the various points covered above, let us take a comparatively simple casting which is being made in a small non-ferrous foundry and run through an example of the procedure which is used in pricing it. This casting is made of No. 356 heat treated aluminum. Figure 3 is a photograph which shows the casting, a core, and the equipment from which they were made.

Since the all important thing in connection with this casting is how much we should charge for it, let us first look at Fig. 1—the Estimate Cost and Pricing Forms. This form acts as a check list upon which is recorded each and every operation going into the making of a casting.

Figure 2 is the Estimate Cost and Pricing Form filled in with the example job we are discussing, a small aluminum casting made in a squeezer mold, six castings per mold and a core for each casting, with a minimum of work in the Cleaning Dept. This casting is heat treated and made to a chemical and physical specification requiring a certified analysis.

To fill in the Estimate Form, Fig. 1 or 2, we should begin with the identification of the job being figured, by filling in the "Pattern No." and "Customer"

PART NAME Housin	ng Bracket			702340
FORM #197	DATE-BY	EST_ //-//-	510	ACT V
	0	MEA /CS	rG.	CSTG.
EACH CASTING NE			lb.	
METAL <i>35</i> 6 @\$.3725			2	\$ .203
COREMAKING	@ \$ .0589			.058
MATERIAL	.0080		lb.	.002
CORE FINISHING	. 0589	-	m.	_
MOLDING	. 0657	.70	m.	.046
MATERIAL	Mold Indirect			
POURING	Mold Indirect			
SHAKEOUT	Mold Indirect			
SAND CONDITION	Mold Indirect			
SAW	. 0753	.07	m,	.005
STAND GRIND - SNA	AG . 0599	.57	m.	.034
STAND GRIND - TRI			m.	.007
CHIP AND HAND GR		.36	m.	.023
BELT SANDER	. 0620		m.	1000
DISC SANDER	. 0692		m.	
SANDBLAST	. 0686	.50	m.	.034
HEAT TREATING	. 1500	.45	lb.	.068
CHEM. & PHYSICAL		.45	lb.	.180
SUBTOTAL I	PER CASTING			\$ .753
IOB RISK /07 + 257		175/	? [	
ADMIN, SELLING			-	1.318
SET-UB / 100 + 1.00 =>	\$ 2.00 / 400 cstys			.005
	PRICE PER	CASTI	NG	\$1.323
				1.32

Fig. 2

spaces. Now, beginning with the top of the card and working down, the first item is "Part Name." It is suggested that the pattern number be recorded here again and also a word descriptive of the casting.

"Date-By" provides space for recording the initials of the person figuring the job cost and the date on

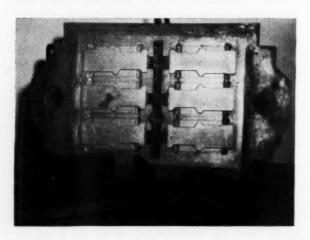


Fig. 3

which it is done. This information of "Date" and "by" will become of real importance as a job is refigured and checked from time to time.

"Estimate____Actual____"—one or the other of these two should always be checked, so that whenever a question arises as to the basis on which a price was established, it is possible to ascertain the answer definitely and quickly. "Estimate" would be checked when working from the blueprint or a pattern that has not been run, so that heading and gating is still subject to experiment, etc. "Actual" would be checked only after experience has been gained by actually running the job and the work involved is definitely known.

Now—to continue with the important thing we are after—the price of the casting. The first item of cost to be determined is the metal and melting cost of the casting. To find this it is necessary to know four things—(1), the net casting weight; (2), the kind of metal to be used in making the casting; (3), the ingot or material price of the metal and (4), the percentage of yield from gross weight of metal poured per mold to the net casting weight.

As advised, the casting being used as the example is made of No. 356 aluminum alloy and each casting weighs 0.45 lb net. The ingot price from the smelter is \$0.3725 per lb and the yield, which is the net casting weight per mold, divided by the gross metal weight per mold, is 50 per cent. Referring to our metal cost table, Table 1, we find that on the basis of a 50 per cent yield and at a market price of \$0.3725 per lb, we have a conversion cost of \$0.4512 per lb of net casting. Therefore, for the 0.45 lb of metal per casting, at \$0.4512 per lb, the cost per casting is \$0.203 for metal, which is entered in the first space of the "Casting Cost" column.

TABLE 1-METAL COST TABLE

Per lb Process	Yield of Net	M	farket Pr	ice of In	got Meta	1
Cost	Castings	.3700	.3725	.3750	.3775	.3800
.0220	85	.4331	.4359	.4387	.4415	.4443
.0234	80	.4345	.4373	.4401	.4429	.4457
.0249	75	.4360	.4388	.4416	.4444	.4472
.0267	70	.4378	.4406	.4434	.4462	.4490
.0288	65	.4399	.4427	.4455	.4483	.4511
.0312	60	.4423	.4451	.4479	.4507	.4535
.0340	55	.4451	.4479	.4507	.4535	.4563
.0374	50	.4484	.4512	.4540	.4568	.4596
.0415	45	.4526	.4554	.4582	.4610	.4638
.0468	40	.4579	.4607	.4635	.4663	.4691
.0524	35	.4635	.4663	.4691	.4719	.4747
.0623	30	.4734	.4762	.4790	.4818	.4846

The next consideration is the Casting Cost for Coremaking. A specification is set up for the core; then by utilizing Standard Data, the Standard expressed in Measured Minutes is determined, which is the time required and allowed for the making of the core. This time is entered in the "Measure per Casting" column and is in minutes and hundredths of minutes, being in this particular case, 0.99 of a minute.

It should possibly be noted here that this "Measure per Casting" and the following operations are alTS

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most entirely expressed in Measured Minutes. For those who have them available, Standards from Standard Data expressed in Measured Minutes would be the figure utilized here. However, for those without such a measure, it is possible for Supervision or Management to make their considered estimate of the time it will take to do the operation.

Because it is possible to use either a Standard or a figure of time based upon someone's experience, this method is, as previously stated, usable by the shop which has and the shop which does not have Standards or time study as one of its Management tools. The degree of accuracy and consistency of the estimates is varied by which method is used for determining the time for doing each operation.

To determine the Coremaking cost per casting, multiply the minutes of coremaking time by the conversion figure for this operation. The result is

the Coremaking cost per casting.

The next consideration is the core materials used, and since the core weight is 1/4 lb per casting, and the conversion cost per pound of material is \$0.0080, we have a cost per casting of \$0.002. This particular core requires no core finishing such as paste, filing, filling or assembling two or more cores together. Accordingly, there is no cost per casting for Core Finishing.

B FLASK TYPE TIGHT POP-OFF MACHINE TYPE JOLT SO . HD 50 PATTERN MATERIAL Mela! PATTERN TYPE Plate SPECIFICATION SISTEMEN AND COMPANY OF CREAT AND CRE FLASK SIZE DRAG VOLUME 768
COMMENTED VOLUME /536 COPE DEPTH # PATTERN DRAW COPE 1/2 DRAG 3/4 FLASK AREA- PATTERN DRAW FROM DRAG (I)
DRAG VALUES (n) (b) (2) (3) FACE CORP (II) (II) (1) SHOVEL FACE COPE MODLE DRAG VOLUME VALUE COPE VOLUME VALUE- DRAW FROM PATTERN (1) (2) (3) COMBINED VOLUME VALUE CARRY OUT - ONE MAN 178 TWO MAN DRAG SEPARATELY COPE SEPARATELY JOLTS - COPE DRAG 6 SPECIAL RAM-UP COPE DRAG SPECIAL DRAG SQUEEZE FINGER PEIN COPE LENGTH DRAG LENGTH SHOVEL PEIN 56 161 DRAG LENGTH TUCK RAM DEPTH CSTGS per MOLD DRAG LENGTH DEPTH SPRUES NO / TYPE GAR DIA LGTH TYPE LGTH RISERS NO TYPE DIA LGTH TYPE DIA LGTH RAP FLASK MTT NO 707340 CORES- IST PICK UP & Class 3 2-ND 3-RD 4- TH S-TH Trim Mold 12.5 BACK OF CARD- SUBTOTAL UNADJUSTED M M / MOLD SHILL ADJUSTMENT FACTOR TOTAL MM / MOLD

Fig. 4

The next operation is Molding. Figure 4 shows the form of a specification card for Squeezer Molding, filled in with the Standard Data allowances required by the example job, which has a Standard of 4.20 Measured Minutes per mold. However, because there are six castings per mold, The Standard per casting is 0.70 Measured Minutes. Seventy hundredths Measured Minutes per casting times the conversion figure of \$0.0657, equals \$0.046 cost per casting.

In this particular setup which was used for illustration, it was found more practical to include molding material costs—the pouring, shakeout and sand conditioning—as indirect costs of the molding and melting. Therefore, they are included in the conversion factor of \$0.0657 for molding and metal cost conversion factor, and are not considered as separate figures at this time. At a future time it is planned to increase the accuracy of this estimating setup by changing these figures from molding and melting indirects to their own respective directs, which will then require that a measure per casting for each direct operation cost center must also be determined.

In the Cleaning Department the first operation is the removal of the casting from the gating system by the use of the band saw. This work carries a Standard per casting of 0.07 Measured Minutes. Multiplying this by the Conversion figure of \$0.0753 gives a cost per casting of \$0.005.

On Stand Grinding the Standard is 0.57 Measured Minutes times the Conversion Factor of \$0.0599, giving a cost per casting of \$0.034.

Chip and Hand Grind has a Standard of 0.36 Measured Minutes times the Conversion Factor of \$0.0641, giving a net cost per casting of \$0.023.

Sand Blasting has a Standard of 0.50 Measured Minutes, times the Conversion Factor of \$0.0686, or a cost of \$0.034 per casting.

Heat Treating, the next operation figures in the estimate, is done by an outside firm who charge \$.15 per lb. Accordingly, the \$.15 per lb figure was placed in the Conversion Cost column and the "Measure per Casting" column shows the 0.45 lb weight per casting, giving a cost per casting of \$0.068.

The cost for the certified, physical and chemical analysis is the next item. The conversion cost for this was determined to be \$.40 per lb. The measure per casting is again 0.45 lb, therefore \$.180 per casting is the cost. The cost of the certified analysis includes not only the work which is done by the laboratory, but also the test bar molding costs, the metal used in making the test bars, and the complete handling of them, so that as near as possible, the cost indicated here is the total cost incurred. Because of the requirement of test bars and the certification of physical and chemical properties of the heat, this figure will, of course, bring out very forcibly the advantage of having the greatest possible weight per heat, so that this cost can be spread over the greatest amount of casting weight possible.

This brings us to the "Sub-Total per Casting" or \$0.753, the total of the individual casting cost items. To this Sub-Total per Casting" it is necessary to add a consideration for the job risk involved in making the casting. In this instance the foundry process risk

is low, so that 10 per cent is added for casting scrap loss, due to the design of the job and the problems involved in producing it in the foundry. However, because this job required certified chemical and physical tests, an additional 25 per cent was added to the job risk to cover the risk in not conforming to requirements.

Administrative, Selling and Profit percentage is the next item and it has been determined 140 per cent is required. The total percentages—covering Job Risk, Administrative, Selling and Profit—amount to 175 per cent. Multiplying 175 per cent by the Sub-Total per Casting figure of \$0.753 equals \$1.318, which is the selling price, exclusive of one item. That item is the adjustment for setup and teardown of the

equipment for making the job.

Setup is considered separately and last, so that the quantity which is ordered at any one time can be taken into consideration. Today we might have an order for 100 pieces—tomorrow it might be an order for 1000. To take care of getting the corebox and pattern equipment out of storage, cleaned and rigged for production, a setup charge is necessary. This should be worked on a definite schedule of prices, divided by the particular quantity which is required at any one time. The setup cost, added to the previous cost per casting, gives the total net price per casting for the specific quantity desired.

In following the pricing method to this point, it is noted that we have a definite form which provides a check list of each of the operations the casting must go through to reach the shipping room door. Figures of Measure have been used which can be determined either from Standards set from Standard Data, as illustrated, or we could just as well have used the estimated time determined from someone's experience of the time it would actually take to do the different operations in a particular shop.

As we have figured out the selling price of the casting used as an example, we have at each operation made use of a figure on the Estimate Form, Fig. 1, to be multiplied by the measure per casting for de-

COMPANY X Foundary Company
DEPARTMENT: Middle CONVERSION FIGURE
BUILD UP SHEET

DEPT, PERSONNEL - DIRECT 4 88 120

COST		COST PER
	DIRECT LABOR COST	\$ .02.83
	(a) Insurance cost against payroll @ 2% 7	.0007
	(b) Tax cost against payroll @ 4%	,0011
2.	INDIRECT LABOR COST	.0/00
	(a) Insurance cost against payroll @ 2%2	. 0003
-	(b) Tax cost against payroll @ 42	.0004
3,	SUPERVISION (share of)	_
	(a) Insurance cost against payroll @	
	(b) Tax cost against payroll @	
4.	SUPPLIES AND EXPENSE (Minor Items)	.000/
5.	POWER (share of) Accompressed light and Conveyer hand Sychon	.0047
6.	FUEL FOR	
7.	MAINTENANCE - Labor	.0002
	Materials	.0001
8.	DEPRECIATION ON SET-UP AND EQUIPMENT	.0038
9.	SHARE OF OVERHEAD COSTS	.0/60
10.		
11.		
12,		
	CONVERSION COST PER MINISTE	. 1659

Fig. 5

termining the \$ Casting Cost, which is the "Conversion Figure."

The Conversion Figure is the amount of dollars and cents it is necessary to charge for each unit of the measure used in measuring the cost at the Cost Center, whether this be minutes of time, or pounds,

or whatever medium may be used.

Let us now follow the making up of the Conversion Figure. Figure 5 is the "Conversion Figure Build-up Sheet." The purpose of this sheet is a check list to be used in determining the conversion cost per minute for the specific Cost Center by filling in one of these sheets for each operation Cost Center. The first consideration, of course, is the department personnel and as an example, we have as direct personnel, four molders at a base rate of \$1.70 per hour. The indirect labor personnel consists of two men at \$1.20 per hour. In this particular shop there are no costs covering Supervision, for this department, as the operations are so set up that the General Superintendent is the Supervisor for the Cost Center and his salary is included in the general shop overhead. Therefore, his salary will be included as a share of the overhead cost.

The second essential of an estimating system is ease of revising conversion figures to reflect changes in cost of labor, materials, etc., for each Cost Center. In addition, it must be easy to build up the original conversion figure for a Cost Center.

Proceeding into the details of this method we will go through the process of determing one of the cost conversion figures for the Molding Department, as an example, which we have found is \$0.0657 per minute of direct labor time.

#### Conversion Figure Build-up Sheet

Working down the Conversion Figure Build-up Sheet, on the left hand side under the heading "Cost Items," the first item is "Direct Labor Cost." On the right hand side is a column with the heading "Cost per Minute." For purposes of illustration we will use an uncorrected, or 100 per cent direct minutes of production for each minute of direct measured work. However, under a setup for normal application, we would make a correction to take care of the molder being unable to work the full hour on what would be classified as direct productive work. This extra cost per direct measured minute also takes care of miscellaneous delays which would be encountered under normal conditions. It would be the actual average cost against the direct labor measure and if the foundry operates on a day-work basis and is in the habit of estimating the time, they also can use the 100 per cent of work for each minute estimated. To find the conversion figure, we would take the molder's rate of \$1.70 per hour and divide it by 60 min per hour, which will give a cost per minute of \$0.0283.

Item "a" under Direct Labor Cost is the Insurance cost, which must be paid by the Company against the Payroll dollar. This has been determined from the insurance policies as being 21/2 per cent of the direct labor cost. Accordingly, \$0.0283 times 0.025 equals \$0.0007 per direct minute.

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Item "b" under Direct Labor Cost covers Taxes against every payroll dollar which will be paid out, such as Social Security, Unemployment Insurance, etc. As an example, we will use the figure 4 per cent, multiplied by the Direct Labor Cost, which equals \$0.0011.

Cost Item No. 2, Indirect Labor Cost. This is based on a figure of \$1.20 per hour, divided by two, because for each Direct Labor Hour we have a half hour of Indirect Labor that must be provided for. Therefore, \$0.60 per hour divided by 60 min per hour, gives an Indirect Labor Cost of \$0.0100 per minute. Provision must also be made for insurance cost and tax cost against payroll for Indirect Labor. These are the "a" and "b" items under Indirect Labor Cost No. 2, i.e., \$0.0003 and \$0.0004 respectively.

Cost Item No. 3, "Supervision (share of)." As previously mentioned, the Supervisor of this cost section is included in the general overhead figure and therefore we will not show any cost per minute at this particular time. However, if the Cost Center had a supervisor working either full or part time, whose hours could be charged directly to this Cost Center, we would then determine a cost per direct minute for this.

To determine a supervision cost per direct minute, our first step would be to establish the cost of supervision for the Cost Center per hour, day, week or month. Then for the same period of time we would establish the supervision cost, for which we would work out the number of direct minutes of work which would be done in this Cost Center. A word of caution here should be noted and that is to be very conservative in the expected number of direct minutes of work which will be accomplished in the period, allowing for absenteeism, breakdowns, etc., which can cut down the total amount of work done. The number of direct minutes worked out are then divided into the supervision cost and the resulting quotient is the Cost per Minute for supervision, which will be filled in on the Conversion Figure Build-up Sheet for the Cost Center.

The same procedure is followed in breaking down other items of cost to a Cost per Minute basis.

Cost Item No. 4, "Supplies and Expense." It may be considered minor items but are necessary to keep the Cost Center in operation by building up a fund to pay for the items. For example, on the Estimating Card for the Molding Department, we have left out material cost. This is because the cost of adding new sand and the binding materials for it constitute a very minor figure from the percentage standpoint to the total of the direct payroll dollar expended.

However, if you remember in the case of the Core Department, we did have a separate cost figure for material. This was for the core sand used for each particular job, based upon the weight of the cores per casting. In the case of the Core Department, the separate material cost will include the cost of the sand, the cereal binders, the core oils and the cost of mixing a batch of sand, including, if you desire it to be broken up as such, even the labor expended for that, instead of including this in the indirect portion of your labor cost.

A general rule which can be followed in determining whether an item should be kept as a separate item or not is what will happen to the general picture when going to extremes, if the item is either covered up as a per cent of cost of the conversion figure or if it is kept as a separate figure. As soon as the figure becomes, in an extreme case, 5 per cent or more of the total casting cost, it should then be given serious consideration as being a separate cost figure, to be considered on each and every job, such as our core material cost is (Cost Item No. 4). This is because of the fact that a large core, having only 1 hr of labor connected with it may be involved, but because it is a block type may weigh and consume 1000 lb of material. On the other hand, it may be a core which is extremely intricate and the box designed with numerous loose pieces to be tucked, drawn, etc., and because of the reinforcing, venting, etc., may weigh only 10 lb. Therefore, material should be given only small consideration in the overall cost of the second core, while it is a major consideration in the first core.

Another point to be used in determining whether an item is to be kept as a separate cost figure or included as a per cent of the Direct Labor Cost is the practicability of providing a reasonably accurate measure of the item. That is why the core situation is different from the molding, in that it is usually relatively easy to determine the weight of the core sand used in producing a particular core, but molding

COMPANY: X Faunday C.

DATE: Seriambo HV9

DEPRECIATION

BUILD UP SHEET

SCHEDULE OF DEPRECIATION TO COST CENTERS

JAN, ITEM DESCRIPTION NEW PERCE VALUE DEPRI

QUAN.	ITEM DESCRIPTION	VALUE	PERCENT	YEARLY DEPREC
1	Mulding Unit	\$.5000.00	10	\$ 500.00
4	Maldink Machines	800.00	2.0	/60.00
40	Steal Flacks	400.00	20	20.40
22	Matal Powering Jackets 12 x 12	11000	-	40.00
10	12 × 14			
10	12.116			
10	12.X18			
10	14 x 16	2 64.00	20	42.44
151	Wood Prusing Shekets	320.00	100	324.00
3	SHAP Flacks ( Wood) 12.812	20.00	20	18.40
2	12.X/V	64.00	24	12.80
3 .	12.816	/02.00	20	20.00
5	12.118	182.00	20	36.40
4	/2.X20	152.00	20	30.40
1	12.X36	46.00	20	9.20
1	1414	32.00	20	6.40
3	14 × 16	102.00	20	20.40
1	14218	36.40	20	720
1	14 X 20	40.00	20	2.40
1	14 8 24	44.00	20	2.50
2	14 1 28	24.00	20	19.20
2	16 % 16	74.00	2.0	14.80
5	Riddles	10.00	20	2-00
27	Mald Weights	84.00	20	16.20
ret	Bellow Boards	150.00	100	150.00
1	Rayas Sand Catted	175.00	20	35.00
1	Rie Compressor	700.00	10	90.00
-				

TOTAL YEARLY DEPRECIATION - 3 1/6/32.00

[Divided by) TOTAL DIRECT MINUTES - 42.000

DEPRECIATION COST PER DIRECT MINUTE - 5.00216

250 days/yr. x 420 minutes/day x 4 operators 424 000 DIRECT MINUTES PER YEAR

sand used is subject to a good deal of variation due to flask depths, etc., and it is not too easy to obtain an accurate measure of the amount used to make a

particular casting.

Cost Item No. 5, "Power" and the share of it which will be used by this particular Cost Center. One of the most expensive items in any shop is, of course, the air cost and this can be included in the item of power cost. Light and any other power that is used in this particular section must also be provided for. For purposes of illustration we will use one-half cent per minute of direct labor time as the cost of "Power."

Cost Item No. 6 is "Fuel," used for a particular cost section. In the case of the Molding Cost Center we are using no fuel for baking or other purposes; therefore, we will leave it out.

Cost Item No. 7, "Maintenance-Labor and Material." On Labor Maintenance we should, if possible, go back to determine what has been expended for the continued operation of this Cost Center. Materials are kept as a separate item but are on the same basis. In the case of this particular Cost Center, it was found that depreciation covered the Maintenance Labor and Material costs.

Cost Item No. 8 is "Depreciation on Set-up and Equipment." To determine what this figure should be we should refer to the next form, Fig. 6, for the form that could be followed in determining the depreciation on the setup and equipment for any Cost Center.

Using the "Depreciation Build-up Sheet" we will enter the Quantity, the Description of each item of equipment in the Cost Center; the New Value; the Per cent of Depreciation and the Yearly Depreciation that should be provided for from a cost stand-

point.

It might be well at this point to enlarge upon what we would term the "New Value." The New Value is given as a heading because of the opinion that for the purpose of determining the cost conversion figure, the amount that would be paid for replacement of the item if purchased new would be the figure to use. We are not interested in what this piece of equipment may have actually cost at the time of purchase, because of a fortunate buy or the methods used in obtaining it, as would be considered from an accounting standpoint. There is a difference in the point of view as to whether we are doing costing work or doing accounting work from the standpoint of taxes.

From a costing point of view we are only interested in providing for the building up of a fund to replace this piece of equipment at the time when it is worn out through use or as it becomes obsolete due to age. Applying the percent of depreciation which we believe it requires for replacement against the New Value of the equipment, we find the yearly depreciation in dollars which should be provided for, and this is totaled in the Yearly Depreciation column of all the equipment in this particular Cost

To determine the depreciation cost per direct minute, take the number of working days per year-250-

times the number of direct minutes of work per day that can be expected from each of the direct operations-times the number of direct operators in this Cost Section. In the case under discussion it will give us 420,000 direct minutes per year. For purposes of illustration it has been assumed that this department Cost Center will work a full day with a full crew for each working day of the year. However, if you will refer to your own experience, it is very likely that you will find that a safe figure to be used for the total number of direct minutes of work per year would be more nearly 50 per cent of the figure we have determined. As an opinion on this, it is reason able to assume that 50 or 55 per cent of the theoretic al direct labor minutes will be the actual production over a period of years.

At the moment we are going through the problem of not being able to provide enough production capacity on the whole to provide for the policing action going on in the world. However, we should go back into our records previous to World War II and take into consideration the good times, as well as the bad. Very likely you will find that your opinion will also be that 50 per cent of the Direct Labor minutes that can be figured theoretically, will be more nearly the actual Direct Labor minutes on a year-in and year-out basis. Therefore, this would be the figure that would be used and it will provide a higher depreciation cost per direct minute than the figure of 420,000 direct labor minutes per year as

COMPANT: X Founday Company
DEPARTMENT: MI OVERHEAD COST
COST CENTERS MI SUIL D UP SHEET

TEM NO.	DESCRIPTION OVERHEAD ITSMS OF COST	MONTHL
NO.	DESCRIPTION - OVERHEAD ITEMS OF COST RENT (or like charge)	\$ /75.00
2.	TELEPHONE & TELEGRAPH	
3.	INVESTMENT RETURN - 6 % on 25 M:12 Months	20.00
4.	INSURANCE - Not against Payroll	125.00
5.	TAXES - Not against Payroll	20.00
6.	DEPRECIATION - General Plant Items (not in cost centers)	20.00
7.	HEATING 300.00//2	7.90
8.	EXPENDABLE TOOLS AND EQUIPMENT	2.5.00
9,	EMPLOYEE INSURANCE PAID FOR BY COMPANY	10.00
		-
11.	VACATION PAY	-
	PAID HOLIDAYS	-
12.	CHRISTMAS PRESENTS	-
	OTHER MISCELLANEOUS ITEMS	10.00
14.	PLANT MAINTENANCE - Parts	25.00
	- Labor	175.00
	(a) 2/4/2	4.38
	(b) 4%	7.00
	PERSONNEL - OVERHEAD CLASSIFICATION	-
15,	Plant Superintendent	400.00
_	(a) Insurance cost against payroll @ 2%2	10.00
	(b) Tax cost against payroll @ 47	16.00
16,		-
_	(a)	
_	(5)	
17,		
_	(a)	
	(b)	
18.		
	(a)	
	(b)	-
19.		
	(d)	
	(b)	
20.	7.	
	(a)	
	(b)	
21.		
	(a)	
*	(b)	
	TOTAL MONTHLY OVERHEAD COST	\$ 1050.28

250 days/yr. x420 minutes/day x/0 direct operators ÷ (12 months) x.77 (2622)

DIRECT MINUTES PER MONTH

OVERHEAD COST PER DIRECT MINUTE \$ .0/600

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shown in the example. Using the figures of the example, we have a depreciation cost of \$0.00384 per Direct Minute.

Referring back to the Conversion Figure Build-Up Sheet, Fig. 5, Item 8—"Depreciation on Set-up and Equipment," this was transferred directly from the "Depreciation Cost per Direct Minute" form, Fig. 6.

Cost Item No. 9, "Share of Overhead Costs." This is built up by using a form similar to Fig. 7, "Overhead Cost Build-up" Sheet. Here again the item numbers are on the left hand side and the monthly cost on the right hand side, with the description of the overhead items of cost.

## Overhead Cost Build-up Sheet

Item No. 1—"Rent"—(or like charge). This covers charges for the property that is occupied by all operations. In the particular example under discussion, this item has been included in the general overhead items of cost. It may be argued that this is an item which should be divided up by the floor space occupied by each Cost Center and put into the conversion figure built up for each Cost Center. However, for a small company this could be debated without serious consequences to the final figure—whether it was included as part of the overhead item of cost or put directly to the Cost Center.

Item No. 2—"Telephone and Telegraph" cost per month. These two items of cost can usually be ascertained quickly and accurately from the accounting record of Operation's Expense Statement for the different months. If not, a quick check can be made of the bills themselves. This is an item of cost that bears close watching, as it can run away from what has been set up as a standard cost by an excess of long distance calls.

Item No. 3—"Investment Return." This covers the amount of return which should be expected on the overall investment of the plant, equipment, working capital, etc., which goes into making and operating the business. It is possible, on first thought, that this should be covered in a percentage figure that would be added to the final cost. However, when we stop to think about it, this is an item of cost which is a must to be recovered and so we should include it as part of our overhead of operations.

Item No. 4—"Insurance—not Against Payroll." There are certain items of insurance that cannot be allocated against the Payroll Dollar and they must be covered in our costs. This insurance coverage would be comparable to fire and theft insurance, bonding and all other types carried by the Management of a Company as protection against risks which are not figured by the insurance companies against the payroll dollars that are expended. This is an item which must be watched, so that as insurance rates change, the cost of this item can be adjusted in our overhead build-up.

Item No. 5—"Taxes—Not Against Payroll." Taxes, such as property taxes, licenses, etc., must be covered and in this particular build-up of costs, have been considered as being definitely overhead in nature. Taxes are an item that must be carefully watched these days as to their rates, so that an adequate sum

will be built up to cover them when due.

Item No. 6—"Depreciation." This is on general plant items not included in the Cost Center Depreciation Build-ups. To determine the amount of money that must be provided to cover the depreciation of all those items which have not been included in the other depreciations, we must follow the same procedure that we used for the individual Cost Center, making use of the form shown in Fig. 7 for the determination of this overhead depreciation.

Item No. 7—"Heating." This is the yearly heating bill for the Company and here again the question may be raised as to whether this should be divided to the individual Cost Centers by space occupied or whether it can be left as an overhead item of cost as shown. Depending upon the size of the operation and the total amount expended for this item per year, would be the basis for a decision as to whether it should be reduced back to a figure charged against the individual Cost Center. However, in the case of a small plant such as the company from which this example set-up has been taken, it has been found that it is as practical to charge it against the general overhead as it is to break it down to the individual Cost Center.

Item No. 8—"Expendable Tools and Equipment." This covers those tools and equipment which are general in nature and not easily allocated to a particular Cost Center in any way. This is an item which, if your accounting record of operating expenses is so broken up as to set it up as an individual item of cost, then the average cost for it can be taken from the accounting record.

Item No. 9—"Employees Insurance Paid for by Company." This may become an item for the individual Cost Center if it can be charged directly against the Center, either as a percentage of the payroll dollar or as a set amount of cost per minute, so that it can be put in the conversion figure build-up sheet, Fig. 5. Here again a decision must be reached based upon the total amount that this figure may run and if it is sizeable, it should definitely be worked into the Cost Center figure directly; whereas if it is a small company and the amount is small, it can more conveniently be built into the overhead figure and distributed in that manner, as was done in this example build-up.

Items No. 10 and 11—"Vacation Pay and Paid Holidays." The distribution of cost for these two items depends upon how they are considered in the long run—whether they can be charged directly against the Cost Center as a percentage of the labor dollar or whether they are to take that figure which must be charged as part of the general operations overhead. In the particular example under discussion they were not included. If Vacation and Paid Holidays are something you cannot tie directly to the Payroll as a percentage figure, then if you were paying them you would put them in as an overhead figure.

Item No. 12—"Christmas Presents." The company's policy from the standpoint of sales and/or personnel will decide the inclusion of a figure of this type as part of the overhead cost of doing business. This particular company, as shown in the example, did

not consider this as an item which should be included at all in their cost of doing business, so that in Fig. 7 no amount was included in the overhead cost build-up.

Item No. 13—"Miscellaneous Items." This will take care of any other items of cost of an overhead nature which have not been specifically mentioned previously.

Item No. 14—"Plant Maintenance." As reported earlier in this discussion we did not show a maintenance figure under Cost Item No. 7 on the Conversion Figure Build-up Sheet covering the Molding Department. However, we have included it in our overhead in this case. This Plant Maintenance figure is made up of two parts—(1) to cover the cost of part materials used in maintaining the plant and (2), to cover our labor cost for maintaining the plant. To the Direct Labor figure will, of course, be added the provision for our insurance cost and our tax cost against this payroll.

Item No. 15—"Personnel." This is the personnel who will carry the overhead classification, such as plant manager, clerical help, inspectors and shipping labor. Against each of these will be charged the insurance cost and the tax cost against the payroll, as was done in the build-up of each of the individual Cost Centers.

Adding together the individual items of cost we have a total monthly overhead cost. To break this down to an overhead cost per Direct Minute it is necessary first to determine the make-up of the Direct crew setup for the entire plant. In this case there are four Molders, two Coremakers, one Direct Melting and three Direct Cleaning Department operators, or a total of ten operators; divided by twelve months and adjusted for the expected volume of work, which is 75 per cent of capacity, we get a total of 65,625 direct minutes per month. This figure divided into our total monthly overhead cost gives an overhead cost per Direct Minute of \$0.01600, which is the total that is entered in Cost Item No. 9 on each of the "Conversion Figure Build-up Sheets," Fig. 5 for each Cost Center. The total of the cost per minute for each of the cost items on Fig. 5 gives a conversion cost per minute which must be charged to cover the cost of doing business in this particular Cost Center.

In determining the conversion cost per minute for any other Cost Center, exactly the same procedure and considerations are gone into.

In using this kind of a cost build-up, it is apparent that it is possible that any time any item of cost should change, to set it into the build-up sheet of the conversion figure and see what effect it will have upon the overall cost. This means that by using this method of pricing castings you can immediately take advantage of any major change of your items of cost and either increase or decrease the selling price per casting to reflect your own costs of doing business.

It is advocated that a selling price per casting be determined in all cases and then only if a customerinsists that this be turned into a price per pound, that a price per pound be furnished. It has been found in costing a great variety of different jobs that the fluctuation in price per pound of castings is al-

most unbelievable in its scope because of the difference in the amount of labor that goes into castings of identical weight. This variation in price is due to the amount of work required in some castings, such as core work, special molding work, excessive cleaning time, etc., so that the cost of one job from the per-pound standpoint may be low, while in other jobs it may be extremely high.

After using a method such as this, anyone will soon become thoroughly convinced of this as a policy to be followed, as have others who have and are already making use of such a method. It would seem that it is a good policy to make a profit and recover your costs on each job and not hope that your work will be in such proportions that even though some jobs must carry the cost of others, the proportions will be such that there will be a resulting profit, because enough jobs with a fairly high selling price per piece will actually have a low cost and therefore provide the cushion and profit for the job which has a high cost actually per piece, but is sold at a price per pound that is not adequate to take care of the costs incurred in producing it.

Using this system we will consistently price a variety of castings in relative position, one with another, which will be mutually beneficial to invested capital in the business, the company's employees and the customer served.

#### DISCUSSION

Chairman: R. L. Lee, Grede Foundries, Inc., Milwaukee.

Co-Chairman: M. E. Annich, American Brake Shoe Co., Mahwah, N. I.

Recorder: M. E. ANNICH.

P. L. Arnold (Written Discussion): This paper is a very valuable addition to existing literature on the perplexing question which has always confronted the foundries. As a matter of fact, it is the most comprehensive and thorough treatment that this writer has had the pleasure of reading.

All of the author's work definitely gives the impression that he is relying on timestudies to establish all of his work factors. In one particular instance, it is even suspected that he resorted to micro-study in spite of the fact that at the outset he states that it is desirable, but not essential, that the cost standards be based on timestudy data. It is believed that he is doing an injustice to his work, because unless records are kept with painstaking accuracy over extended periods of time, the data are not apt to be reliable and it will take much longer and cost more to obtain the cost data on basis of experience from records than it will to establish them in a reasonably short period of time by means of timestudy. Timestudy has the further advantage that it will indicate the exact performance level at which the various jobs are being performed from time to time and variation in performance level is apt to give erratic results in an experience record.

The author's statement in the upper righthand column on page 757, "the proof of the accuracy of the estimating and cost work will be reflected in a Profit and Loss statement prepared by the Accounting Department" may be deliberate from his point of view. The control rests with the Accounting Department and is not an estimating and costing function, but it is suggested that if a foundry goes to the trouble of installing this or any similar system, that it should be utilized for control purposes also, so that the shop management will at all times have a ready reference to indicate whether the system is effective and accurate and/or the performance of the shop personnel is up to a reasonable expectancy. The author's approach to the problem by dividing the various operations into individual cost centers and charging depreciation and maintenance cost to the individual centers and classifying overheads in the same manner, is excellent. In the past, many shops have made the mistake of

¹ Resident Mgr., United States Pipe & Foundry Co., Chattanooga, Tenn.

figuring all of these items on an overall basis, which may be perfectly satisfactory as long as the entire shop is fairly uniformly loaded or the castings being produced are fairly representative of average conditions, but highly unsatisfactory if only part of the shop is running or if for some particular reason

difficult and intricate castings are preponderant.

Referring to Fig. 2, it is assumed from the text that a "job risk" covers scrap losses incidental to this particular type of casting and not an average figure for the shop. Is it not possible that the administrative, selling and profit items would become more valuable if the profit was shown up separate, because the administrative and selling expenses are very likely to remain reasonably constant; whereas, if the estimated profit was directly comparable to the actual profit, management would have a very effective tool with which to adjust the estimating procedure from time to time, rather than waiting for the Profit and Loss statement at the end of the year to show that they had priced themselves out of business.

It is assumed that the process cost shown in the extreme lefthand volumn of Table 1 contains, in addition to the melting

charges, all incidental handling and freight charges.

It is recommended that all readers of this paper also acquaint themselves with the contents of "Approach to Standard Cost in the Foundry" by Fred Ruffolo, referred to by the author earlier in his paper. After thoroughly digesting both of the papers, it would be realized that an excellent control on a day-to-day basis could be made available with very little extra effort. This writer believes that once it is fully realized what can be accomplished in the line of cost reduction through effective control and improvement in methods by studying the variance between actual and standard cost resulting from improper utilization of labor, ineffective methods, etc., everyone will agree that the extra cost of competent timestudy men, either recruited from the outside or, still better, developed from experienced shop personnel, will be a good investment.

MR. WESTOVER (Reply to Mr. Arnold): The author greatly appreciates the comments of Mr. Arnold. In his comments, Mr. Arnold has porposed certain lines of thought that should have

an answer.

1. Mr. Arnold stated, "All of the author's work definitely gives the impression that he is personally relying on timestudies to establish all of his work factors."

In answer to this statement, while the original conception of this costing formula and method were developed from a time-study and standards background, it has been the experience since then in installing a number of these set-ups, that some of the most successful of them are in shops that are entirely on a day work basis of pay; also the installation time of the set-up was less than in those shops where it was necessary to correlate time study performance record, etc.

2. Mr. Arnold also stated, "Referring to Fig. 2, it is assumed from the text that 'job risk' covers scrap losses incidental to this particular type of casting and not an average figure for the

shop."

In reply to this I say that this is a risk factor that is for the individual job. See the last paragraph on page 759, "To this Sub-Total per Casting" it is necessary to add a consideration for the job risk involved in making the casting.

3. Mr. Arnold stated, "Is it not possible that the administrative, selling and profit items would become more valuable if the profit was shown up separately, because the administrative and selling expenses are very likely to remain reasonably constant; whereas, if the estimated profit was directly comparable to the actual profit, management would have a very effective tool with which to adjust the estimating procedure from time to time, rather than waiting for the Profit and Loss statement at the end of the year to show that they had priced themselves out of business."

In answer to this I say that because the administrative and selling expenses do usually remain reasonably constant, they were included with the Profit figure to cut down clerical detail; however, the more breakdown and detail you have in cost figures, the better you are able to analyze and know where corrections should be made.

4. Mr. Arnold stated, "It is assumed that the process cost shown in the extreme lefthand column of Table 1 contains, in addition to the melting charges, all incidental handling and freight charges."

To this I say that the assumption is correct.

J. F. KANZLARICH: ^a On a long run job where an hour's coremaking time is sufficient to keep a molder in cores for several days, how do you handle subsequent set-ups?"

Mr. Westover: In costing the job, the extra coremaking setups should have been ascertained and included in the selling price.

C. A. REAM: ³ Is it accepted practice to include a % on Investment in Cost Data?

Mr. Westover: Yes, because in one way or another you do have to provide funds for investment return.

MR. REAM: Does actual practice procedure figure depreciation on replatement value rather than cost reflected on books?

Mr. Westover: It should be figured on replacement value for costing purposes, particularly in a time of inflation, as at the present. From the standpoint of tax account, however, the Cost value must be used.

M. M. MRVICHIN: 4 In revaluing assets to reflect current replacement costs, is it necessary to obtain the approval of the Internal Revenue Dept.?

Mr. Westover: It is not necessary to obtain approval where the figures are to be used for costing purposes only.

PAUL B. HARNER: 5 Please elaborate on A—The set-up charge,

and B-The short order charge (if any).

MR. WESTOVER: The set-up charge takes care of the short order or the production order equally well. A foundry has certain costs in getting a pattern and corebox out of the storage and ready to run. Then when they have been run there is the cost of cleaning them and putting them back into the storage. This cost should include the writing of the order as well as the equipment handling. Then whatever the set-up cost of running the job is, divided by the number of castings to be run, gives the company a cost which must be included in the price per casting, if the cost of doing this is to be recovered.

Methods Engineer, Peoria Malleable Castings Co., Peoria, Ill.
 Treas., Riverside Foundry & Galv. Co., Kalamazoo, Michigan.
 Project Engineer, Petibone-Mulliken Corp., Chicago, Illinois.
 Treas., Union Mfg. Co. Inc., Boyertown, Pa.

# HOW IRON AND STEEL MELT IN A CUPOLA

By

H. W. Lownie, Jr.*, D. E. Krause, ** and C. T. Greenidge†

#### ABSTRACT

While a small cupola was in normal operation with tapping temperatures of about 2880 F and the stack full of charges, the blast was shut off and the charges quenched in the stack with a full stream of water. The charges were then carefully dug out and examined visually, metallographically, and chemically to obtain information on the behavior of cast iron, pig iron, and steel while being melted in a cupola. Information was also obtained on the behavior of coke during its descent in the cupola stack.

From examination of the quenched charges it was possible to construct a curve representing the distribution of temperature in the cupola during the heat. The order of melting was found to be (1) pig iron, (2) cast iron, and (3) steel. All three of these metallic components of the charge tended to decarburize and oxidize on the surface until they began to melt. High-carbon layers on pieces of steel were found to be remnants of molten cast iron or pig iron that had washed over the solid steel. Cast iron and pig iron melted "from the inside out" leaving a shell that was high in sulphur but low in carbon.

Cupola operators often speculate about how iron and steel behave while they are being melted in a cupola. Some of the pertinent questions are: How much does steel carburize in a cupola before it begins to melt? How much higher does steel melt in a cupola than pig iron or cast iron? How fast does the metal charge pick up heat as it descends in the cupola stack? What is the actual bed height during operation?

To gain some information on this subject a small cupola was placed into normal operation. After operating conditions and tapping temperatures became steady, the blast was shut off while the stack was full of charges. The charges in the stack were immediately quenched with water. When cool, the charges were carefully dug out of the stack, starting from the charging door, and the locations of the various components of each charge were carefully recorded. Samples of iron and steel from various parts of the stack were examined visually, metallographically, and chemically to learn as much as possible about their manner of melting.

The method of quenching cupola charges in the

stack to obtain information on the melting of the iron and steel was not completely original. A similar investigation was reported by Rambush and Taylor in 1945. The new results give additional data which supplement those of Rambush and Taylor.

#### Description of the Cupola and Cupola Operation

The test was made in a 10-in. diameter cupola under closely controlled conditions. A photograph of the cupola and its control panel is shown in Fig. 1. Its construction and operation were described in detail in a paper by Krause and Lownie.² The cupola stack is about 10 ft high, and the shell is 18 in. in diameter. Proportions and tuyere ratios are quite similar to those of commercial cupolas.

The original height of the coke bed was 42 in. above the lower tuyeres. Each charge consisted of 30 lb of metal and 4 lb of commercial oven foundry coke sized



Fig. 1—Ten-inch cupola set up for operation. Control panel is shown in background. Water-cooling as shown in this photograph was not employed in the heat used to study the manner in which metals melt in a cupola.

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to  $21/2 \times 15/8$  in. Therefore, the operating metal/coke ratio was 7.5/1. The composition of the metal components of the charge was as follows:

	Weight,	C	ompos	ition, p	per cen	t
Material	lb	T.C.	Si	Mn	P	S
Steel harrow teeth	10.5	0.58	0.09	0.92	0.07	0.06
Malleable pig iron	9.6	3.76	2.76	0.73	0.19	0.04
Cast iron	9.9	3.19	3.30	0.98	0.18	0.04
	30.0	2.46	2.00	0.88	0.14	0.05

The steel harrow teeth were about ½ in. square and 3½ in. long, and were rerolled from rail steel. The malleable pig iron was in thin pigs about ½ in. thick, and was broken to fist size. The cast iron had been previously melted in an electric induction furnace and cast into special shapes, such as crosses and "doughnuts," so that its composition was uniform and known, and so that the cast iron could be recognized by shape until it began to melt.

Flux charges with each coke split consisted of 1.25 lb of limestone and 0.25 lb of silica pebbles, and gave a typical acid cupola slag at the slag hole. The slag contained about 52 per cent silica, 25 per cent lime, and 8 per cent alumina. The cupola was lined with a siliceous ramming mix.

The cupola blast rate was 12.5 lb of air a min, a rate equivalent to about 6100 cfm in a 60-in. cupola. The blast temperature was controlled to about 950 F and the moisture content of the blast was controlled to about 45 grains/lb of dry air.

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The cupola was tapped intermittently by tapping about 30 lb of iron at intervals of about 21/2 min. Slagging was also intermittent and was from the rear slag hole which was left open.

The cupola operated smoothly and reached a tapping temperature of 2850 F on the fourth tap. On the twelfth tap the tapping temperature was 2900 F. On the twentieth tap the tapping temperature was about 2800 F, and the operation of the cupola indicated that a stable operating condition had been reached. The CO₂ content of the stack gas was about 11 per cent, and was holding steady. The stack was full of charges and the melting rate was about 750 lb/hr. The wind-box pressure was constant at 10 oz/sq in. The composition of the iron at the spout was almost of eutectic composition, and had the following analysis for Tap 20:

	Comp	osition, pe	r cent		
T.C.	Si	Mn	P	S	
3.60	2.19	0.70	0.14	0.09	

Immediately after Tap 20 was made, the blast was shut off and quenching of the charges was started.

#### Quenching Charges in the Stack

Immediately after the blast was turned off the tuyeres were opened and a full stream of water from a 5/8-in. hose was squirted into the cupola through the charging door. Quenching was rapid. After a few minutes water started to drip through the bottom. Quenching was continued until the cupola and its contents were cold. Although large amounts of steam were evolved during quenching, the operation was quite uneventful.

The rapidity of the quench was revealed by later metallographic examinations which showed that much of the steel and çast iron in the lower sections of the cupola had a martensitic matrix.

#### Removing Charges from Cupola Stack

Charges in the stack were removed by digging them out of the stack one at a time, starting at the charging door. There were about 14 charges in the stack when the heat was stopped. The top three charges were easily separated, because the layers of coke and metal were fairly distinct. Below this, metal and coke were so thoroughly mixed that separate charges could not be distinguished. The charges were dug out in layers and were placed in separate containers and identified according to the distance of the layer above the lower tuyeres.

Because of the small diameter of the cupola, and the difficulties involved in reaching as much as 10 ft down the stack to remove material, it was not possible to segregate individual components of the layers according to whether they were in the center or along the wall of the cupola.

The quenching of the hot cupola with water had surprisingly little effect on the lining. The upper-stack sections which were lined with silicon carbide backed up by an insulating ramming mix did not crack. The rammed siliceous lining in the melting zone, around the tuyeres, and in the well did not crack or deteriorate.

## Coke Condition in Frozen Charges

Reduction of Size: The coke reclaimed from the cupola showed there had been surprisingly little reduction in size as it descended in the cupola to the melting zone.

The first coke that showed any reduction in average size was located about 30 in. above the lower tuyeres. Above 30 in. from the lower tuyeres, there was very little ash on the coke. The amount of ash on the coke increased, and the average size of the coke decreased slightly, as the lower tuyeres were approached. However, even at the level of the lower tuyeres, some lumps of coke were almost as large as when charged.

Coke in the Well: The coke in the well was larger than that immediately above it in the tuyere zone. This shows that there was little movement of coke from the tuyere zone into the well.

Results of other work done in the 10-in. cupola on the behavior of coke in the well of the cupola are worth mentioning here. The well coke in one heat included 19 pieces of coke sawed into 2-in. cubes. These cubes weighed a total of 2294 grams when charged. The cupola was burned in by the usual practice, including a long lightup by means of a gas burner firing through the breast onto the coke in the well. A 1-hr heat was made at an average tapping temperature of 2770 F. The charges contained 33 per cent steel and averaged 2.55 per cent total carbon and 2.10 per cent silicon. The iron tapped from the cupola averaged 3.45 per cent total carbon and 1.95 per cent silicon.

When the bottom was dropped, 17 cubes, or pieces

of cubes, were recovered from the original 19 coke pieces. After drying, the recovered cubes weighed 1113 grams, 49 per cent of their original weight. Four of the recovered cubes weighed over 80 grams, as compared to an original weight of about 121 grams per cube. Photographs of the cubes before and after this heat are shown in Fig. 2.

It is surprising that so little of the cubes was lost during the heat. Some of the cubes were broken, and no pieces of coke were recovered from the drop unless they could be positively identified as part of one of the original cubes. Therefore, the total loss of well coke was less than half its original weight.

Obviously, some of the well coke must have been consumed by combustion during the lightup, and some must have been lost by breakage. However, even if all of the coke that was not recovered from the drop had been absorbed by the iron in the well, it would

-2 MINUS

Fig. 2—Appearance and weight of coke cubes used in well of 10-in. cupola in 1947. Top—Coke cubes as placed in well before heat—2294 grams. Bottom—Coke cubes recovered from drop after 1-hr heat at 2770 F tapping temperature—1113 grams.

have accounted for a pickup of only 0.40 per cent carbon. The actual pickup was 0.90 per cent carbon. Therefore, most of the carbon pickup occurred before the iron reached the well.

The results with the coke cubes in the well led to the same conclusions as those with the frozen charges. The coke in the well was consumed very slowly.

#### Metal Condition in Frozen Charges

The melting range of each component of the metal charges was calculated from their known compositions. By visual examination of the metal salvaged from the frozen charges, it was possible to construct a curve representing the probable distribution of metal tem-

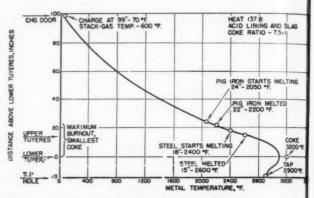


Fig. 3—Probable distribution of temperature of metal charges in heat 137B in 10-in. cupola with 960 F blast.

peratures in the cupola. This curve is shown in Fig. 3. Gas temperatures and coke temperatures were undoubtedly higher than the metal temperatures.

Stack-gas temperatures at the charging door were 600 to 700 F. The metal was charged at room temperature and began to heat immediately. No melting of any component was found higher than 24 in. above the lower tuyeres. At this point, pig iron began to melt. The temperature of the metal, therefore, was close to 2050 F, the approximate solidus temperature for this pig iron. Once melting of the pig iron began, it progressed rapidly. Almost all of the pig iron was apparently melted by the time it has descended 2 in. lower in the stack to 22 in. above the lower tuyeres. Therefore, at 22 in. above the lower tuyeres, the metal temperatures must have been close to 2200 F, the approximate liquidus temperature for this pig iron. The zone between 22 and 24 in. above the tuyeres accounted for the melting of the pig iron, and also was the zone at which appreciable refractory erosion from the lining began.

Based upon the charging rate, it is known that charges descended in the cupola at about 4 in./min. Apparently, the pig iron was completely melted about  $\frac{1}{2}$  min. after melting began.

In the zone where pig iron was melting, the cast iron was beginning to melt, but lagged slightly behind the pig iron. That is, the cast iron descended further in the cupola before it melted. The steel scrap showed no signs of melting until 18 in. above the lower

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tuyeres. By this time all of the pig iron and cast iron remelt were completely melted. At 15 in, above the lower tuyeres all of the steel was melted, and only coke was found in the balance of the cupola. This zone, within 15 in, of the lower tuyeres, contained the smallest coke in the cupola.

According to the usual definition of bed height, the operating bed height in this heat was 15 in. above the lower tuyeres, the lowest level in the cupola at which solid metal was found.

Measurement with an optical pyrometer indicated that the temperature of the bed coke a few inches inward from the lower tuyeres was about 3200 F. From the temperature-distribution curve in Fig. 3, it seems reasonable that the metal could have reached a maximum temperature of about 3000 F as it dropped past the tuyere zone. An assumed temperature drop of 100 F in the well gives the actual tapping temperature of about 2900 F in this heat.

The plotting of these points on the temperaturedistribution curve in Fig. 3 gives a reasonable reconstruction of the temperatures and conditions of melting that existed in the cupola during this heat. Two partially fused aggregates taken from the frozen charges further substantiate the curve in Fig. 3.

Figure 4 shows a partially fused mass removed from the cupola at about 23 in. above the lower tuyeres. The mass shows an almost completely melted piece of pig iron, a partially melted piece of cast iron which was originally a full ring, and a piece of steel that had not started to melt. Because this mass was fairly small, it is reasonable to assume that the temperature was fairly uniform throughout the mass. Therefore, this one mass established that the order of melting was (1) pig iron, (2) cast iron remelt, and (3) steel. This is the order that would be expected on the basis of calculated melting ranges for the three materials.

Figure 5 shows another interesting mass removed from the cupola. The pig iron at the top of the photograph has not yet started to melt, but the steel harrow tooth at the bottom of the photograph was actively melting. An "icicle" frozen to the steel shows that this mass of metal was oriented in the cupola the same as in the photograph, with the unmelted pig iron at the top. Just below the piece of unmelted pig iron is a blob of pig iron that was almost completely melted. Figure 3 shows that the dividing line between the unmelted and melted pieces of pig iron was probably

close to 24 in. above the lower tuyeres. This would put the steel harrow tooth at about 20 in. above the lower tuyeres. Figure 3 shows that the steel should not begin to melt until it falls to a level 2 in. lower than this in the cupola, that is, to 18 in. above the lower tuyeres. However, the manner in which the steel tooth was melting, as shown in Fig. 5, suggests that it was being washed away by high-carbon pig iron that was melted above it. That this is so was verified by analysis of the "'cicicle," which showed a carbon content of 2.94 per cent and a silicon content of 2.28 per cent. This silicon content could have come only from admixture with pig iron or cast iron.

Although the steel tooth in Fig. 5 was obviously in a high-temperature zone and close to its melting point, a metallographic examination of the surface of the steel tooth just above the "icicle" showed that the surface was decarburized rather than carburized.

The melting of steel by "washing" with molten high-carbon iron suggests one reason why low-steel charges melt more readily than high-steel charges. With high percentages of steel in the cupola charges, the chances of obtaining melting by this washing action are greatly reduced, and the steel is probably carried lower in the cupola before it melts.

# Melting Behavior of Pig Iron, Cast Iron, Steel

Pig Iron and Cast Iron showed similar behaviors during melting, but their melting characteristics were quite different from that of steel. Metallographic examination of pig iron and cast iron showed no significant change in structure, even at the surface, until they reached a level about 35 in. above the lower tuyeres, or an estimated temperature of about 1600 F. From this position until melting began, a thick oxide scale was built up and the surface of the iron was decarburized quite deeply.

The combination of decarburization and oxide formation produced a skin of high melting point on the pig iron and cast iron. Therefore, when these materials began to melt, they melted "from the inside out." There were numerous examples found in the salvaged charges of mere skins of pig iron or cast iron. One such example is shown in Fig. 6. The skin has been decarburized from 3.19 per cent to 2.19 per cent carbon, and the sulphur content has increased from 0.04 to 0.07 per cent.

Steel Scrap: It has sometimes been said that before

PARTIALLY MELTED PIGIRON

3.76 %-TC, 2.76%-SI
MELTING RANGE
2050-2200*F

CAST IRON RING
3.19%-TC, 3.30%-SI
MELTING RANGE
2.100-2240*F

COLLAPSED SKIN
LEFT WHEN PIECE OF
CAST IRON MELTED

Fig. 4—Mass of partially melted metal taken from about 23 in. above lower tuyeres where metal temperatures were about 2150 F. Notice that order of melting was (1) pig iron, (2) cast iron, and (3) steel. At the temperature of this mass, pig iron was almost completely melted, cast iron was melting, and steel had not started to melt. Melting ranges shown for each material were calculated from their composition and equilibrium diagrams. Photo reduced one half in reproduction.

steel melts in a cupola it is carburized so that its melting point is lowered. Several pieces of steel that had just begun to melt were examined metallographically for evidences of carburization. In no case was any carburization found before melting began. In fact, every surface of steel that was examined, and had not yet actually begun to melt, was decarburized. However, immediately after melting the carbon content was much higher. This suggests that the steel melted by a process of carbon pickup, with the carbon pickup and melting occurring simultaneously.

#### Photomicrographs of Typical Structures

Photomicrographs of some typical structures found in pieces of metal reclaimed from the cupola stack are shown in Figs. 7 through 10. It was often difficult to interpret the significance of microstructures found in most samples taken from the lower sections of the cupola because the matrices were usually martensitic, reflecting the rapid cooling that must have accompanied the water quench. A group of about 15 specimens taken from the cupola for metallographic examination included most of the microstructures known to ferrous metallurgy. Matrices included ferrite, mar-

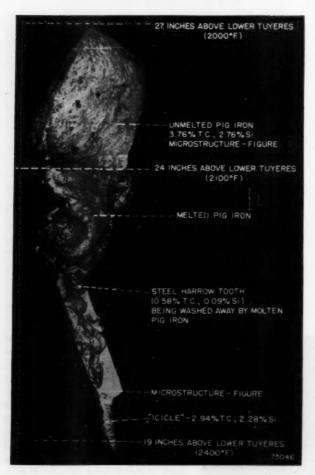


Fig. 5—Semi-molten mass showing melting of steel by washing action of molten gray iron. Photo reduced 40 per cent in reproduction.

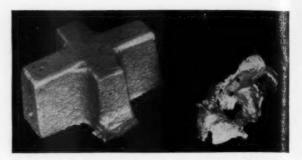


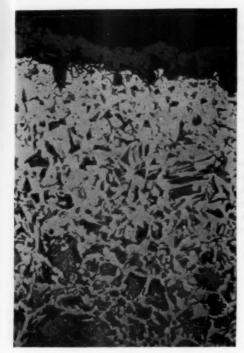
Fig. 6—Samples of cast iron cross removed from different levels in 10-in. cupola. Photo is reduced 40 per cent in reproduction. Left—Cast iron cross; 28 in. above lower tuyeres, metal temperature about 1900 F; 3.19% T.C., 3.30% Si, 0.04% S; pearlitic matrix, some surface decarburization and oxidation. Right—Shell of cast iron cross; 23 in. above lower tuyeres; metal temperature about 2150 F; 2.19% T.C., 3.87% Si, 0.07% S; graphite flakes in martensitic matrix, large amounts of MnS.

tensite, several types of pearlite, and several intermediate products, such as high- and low-temperature bainite. Among the inclusions were massive carbides, steadite, manganese sulphide, nitrides, and flake graphite.

Figure 7 shows the etched microstructures of one steel harrow tooth as it was charged and another tooth just before it began to melt. The decarburization and scale on the tooth as charged came from the process of hot working the steel into the tooth form. This decarburization on the surface persists up until melting actually begins. Also, the scale becomes heavier as the steel descends in the stack. From Figs. 3 and 5, and the rate of descent of the charges, it is estimated that the decarburized steel tooth shown at the right in Fig. 7 would have begun to melt within 15 sec if melting had not been stopped by the water quench. Even this late in the process, no carburization of the steel surface was found. The same situation existed for several other samples of steel examined just before melting would have begun.

Figure 8 shows the structures of two samples of cast iron taken from the hotter zones of the cupola. The photograph on the left shows the surface structure of a piece of pig iron that had started to soften. The appreciable amount of decarburization and scale that make up the high-melting skin are apparent. Thus, both steel and pig iron contained appreciable amounts of decarburized surface right up to the time they began to melt.

On the right in Fig. 8 is shown an unusual structure found in an unidentified mass of partially melted metal. The very small size and compactness of the graphite flakes are of particular interest. At first glance these flakes would appear to come from decomposition of massive carbide. However, none of the constituents charged contained massive carbide. Also, it is extremely unlikely that massive carbide was formed by the water quench, and that there was sufficient heat or time available for such carbides to graphitize. This specimen might have already started



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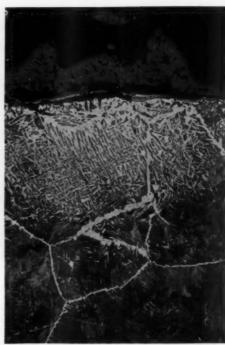
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Fig. 7—Surface structure of steel harrow teeth as charged (left) and just before melting (right). When charged, the 0.58 per cent C steel is decarburized and coated with an oxide scale. Just before melting begins (right) steel is still decarburized and oxidized on the surface. Specimen for photograph on the right was cut from just above the "icicle" shown in Fig. 5.



250X

to melt along the grain boundaries and then graphitized while solidifying after the water quench. To appreciate the fineness of these flakes, the magnification of 500 diameters used for the photographs must be considered. The average length of the flakes is actually about 0.0003 in.

Figure 9 shows the types of concentrations of in-

250X

250X



Fig. 8—Pig iron just before melting (left) showing decarburized layer and
thick oxide scale on surface. Unusual structure
(right) of very fine compact graphite flakes found
in an unidentified mass
that had partially melted
in the cupola. Notice high
magnification.

clusions found in the shell left when cast iron melted. Most of the inclusions were manganese sulphides, but an appreciable number of nitrides were also present. This photograph illustrates that sulphur was picked up on the surface of the iron shortly before it melted. These large amounts of inclusions were not found in any specimens except in the shells. Although this cast iron had a sulphur content of 0.04 per cent as charged, this particular skin contained 0.07 per cent sulphur.

Figure 10 illustrates a possible explanation for reports in the literature that steel melts after its surface has been carburized appreciably to reduce its melting point. This photomicrograph shows a piece of steel with a surface layer consisting of graphite flakes in a martensitic matrix. Chemical analysis of surface scrappings or millings would show the surface to be high in carbon content, and might lead to the conclusion that the surface of the steel had been carburized by

500X

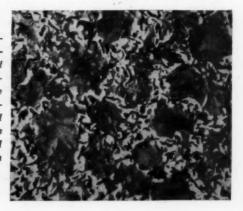




Fig. 9 (left)—Example of concentrations of inclusions found in shell left after cast 'ron melted. Most of the inclusions are manganese sulphide; some are nitrides. This shell is the same as that shown in Figure 6 (unetched).

Fig. 10 (right)—Graphite flakes shown along the surface of a piece of steel are actually in cast iron that has washed over the steel and been welded to it.



100%

500X

the cupola atmosphere. Actually, however, the surface shown in the photomicrograph is a weld zone between steel and molten cast iron which has washed over it. This condition was common on pieces of steel removed from levels below those at which cast iron and pig iron melted. It was not found on steel existing at levels above those of melting pig iron. The same type of surface "carburization" was found on the steel tooth in Fig. 5 in those areas that had obviously been washed with molten pig iron or cast iron. The "washing" theory is further substantiated by the fact that such areas, although numerous, occur discontinuously. If gas carburization were responsible, a mere continuous layer of graphitized metal would be expected.

#### Discussion

In 1945, Rambush and Taylor¹ reported that they charged marked materials into a 42-in. cupola near the end of four regular production heats, purposely quenched the heats with water, and then progressively dug out the charges. Their work was more extensive than that done in the small cupola.

Some of the conclusions and findings of Rambush and Taylor are quoted below and compared with conclusions reached from similar work.

1. "However carefully the charge materials are placed in a cupola in relation to one another, their original relative position is not maintained during the descent of the charge." This was verified in the 10-in. cupola to the extent that coke and metal charges became intermixed shortly below the charging door.

2. "The shape and size of the metallic pieces in the cupola charge are dominant factors in governing their manner of descent." This was not checked in the 10-in. cupola because pieces of metal charged were of about the same size.

3. "The coke, although evenly sized to start with, is progressively reduced in size from top to bottom of the shaft. This remark applies to both bed and charge coke." The coke in Rambush and Taylor's cupola was probably more subject to crushing than that in the 10-in. cupola. This may explain why their results

showed more reduction in coke size than obtained in the 10-in, cupola.

4. Rambush and Taylor feel that the coke in the well begins to float when the well becomes full of metal. This does not happen in the 10-in. cupola.

5. "The reduction in coke size in the bed proper is not uniform across a horizontal plane, the coke in the center being reduced most, while the bed coke against the walls of the cupola remains almost unaffected." This was not checked in the 10-in. cupola because of its small diameter. However, it seems to be a reasonable occurrence for a large cupola, especially where an appreciable slag bridge builds up above the tuyeres, as in the heats made by Rambush and Taylor.

6. "The carbon content of the steel test bars increases very slightly in their downward travel when solid, but on commencement of melting the carbon absorption becomes much more rapid." Results in the 10-in. cupola showed that no surface carburization of steel occurred before actual melting began. During melting, carbon pickup by steel was rapid, especially when molten pig iron washed over the surface of the steel.

7. "Carbon absorption by the melting steel takes place chiefly on the surface, the center of the pieces remaining essentially unaltered in this respect. The highest carbon content found in a solidified drop of steel was 1.2 per cent."

8. "The passage of water and steam through the burden appears to have no deleterious effect on the brickwork of the furnace lining." In the 10-in. cupola, the drastic quench also had surprisingly little effect on the lining.

9. "The steel in the charge picks up sulphur rap-

idly on its surface shortly after its introduction into the cupola, but on descending some of the sulphur seems to be given off again." In the 10-in. cupola, the greatest pickup of sulphur occurred on the surface of cast iron shortly before it melted.

10. In the small cupola, a large percentage of the steel in the charge was melted at temperatures below its normal melting point. This took place because molten cast iron and pig iron dripped down onto the solid steel and dissolved it before its melting tem-

perature was reached.

11. "Some of the carbon picked up by the steel remains graphitic in nature." Results from the 10-in. cupola suggest that the graphitic carbon that seems to be in the surface layer of steel is actually in cast iron or pig iron that has welded to the partially melted steel.

#### Acknowledgment

The authors wish to express their appreciation to the Gray Iron Research Institute, Inc., which sponsored this investigation and gave permission to publish the results.

#### References

1. N. E. Rambush and G. B. Taylor, "A New Method of Investigating the Behavior of Charge Material in an Iron-Foundry Cupola and Some Results Obtained," Foundry Trade Journal, Nov. 8, 1945.

2. D. E. Krause and H. W. Lownie, Jr., "Designs and Operation of a 10-Inch-Diameter Cupola," Transactions, A.F.S., vol. 57, pp. 163-172 (1949).

#### DISCUSSION

Chairman: J. D. Sheley, The Black Clawson Co., Hamilton, Ohio.

Co-Chairman: R. A. CLARK, Electro Metallurgical Div., Union Carbide & Carbon Co., Detroit.

Recorder: W. T. BOURKE, American Brake Shoe Co., Mahwah,

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E. F. FISHER (Written Discussion); ¹ The wide variations in melting and temperatures shows that each piece of coke and metal was not surrounded by an envelope of gas. Indeed it is impossible in an open top cupola to have this condition prevail. Even in such a small diameter as a 10-in, cupola uniform airflow through the charge is impossible. Anyone who has watched the fire in a blacksmith's forge will remember how careful the blacksmith was to control his blast and distribution of coal to get uniform heating of his iron.

In an open top cupola the air takes paths of least resistance through the charge with no uniformity of velocity through the voids. For this reason melting occurs as a progressive washing away of solid metal that has not yet reached the melting point, by molten metal dripping into the well. Any unmelted metal that drops into the well is dissolved by the molten metal. Any coke remaining in the well unburned from the bed is not con-

sumed as the blast does not get into the well.

In an open top cupola with free air blast it is impossible to get uniform combustion and melting. This can be achieved only in a pressurized cupola. Even had all the coke and metal in the 10-in. cupola been in the form of 2-in. cubes there would still have been the same variations in melting and combustion.

L. G. Gudorf, F. H. Backman and F. J. Wurscher (Written Discussion: ² The authors of this paper are to be saluted in recognition of their valuable contribution toward solving a problem that has long awaited thorough investigation. For the purpose of clarification of certain statements made and subsequent evaluation of findings disclosed by the authors this discussion is presented.

 Scrutinizing the data given and referring to performance records of the cupola used, a normal melt is to be anticipated. The quoted melting-rate is but 4.53 per cent short of the rated capacity based on cross-section area.

In the paper reference is made to dimensions "above the lower tuyeres." Is it to be inferred that upper tuyeres had been employed? If so, what was their number and size? Though Fig. 1 indicates but four lower tuyeres a statement to that effect and quotation of their shape and dimension are of interest.

2. The calculation of the metallic charge used records an initial composition against which that of the metal actually received at the cupola spout shows an increase of 1.14 T.C., 0.19 Si

and a decrease of 0.18 Mn.

The increase in Si and the slightly excessive increase in T.C. suggests that this melt proceeded under a neutral or slightly reducing atmosphere. However, the reduction of Mn in connection with the reported 11 per cent CO content in the affluent gases negates such conclusion. Elucidation pertaining to the gain in Si is therefore invited.

3. Referring to quotations in the paper, this test shows an increase of 0.9 per cent T.C. and a decrease of 0.15 per cent Si, indicating the prevalence of a different composition atmosphere than was found in the previously cited test. The authors found that only a small portion of the specially shaped coke in the well-space was consumed during that heat and calculated that if all of the C contained in the coke (90 per cent F.C.) had been absorbed by the metal it would have accounted for but 0.43 per cent C. Thus, 0.47 per cent C were absorbed through carbonizing reactions above the lower tuyeres.

4. Figures 4-10 seem to be apparent evidence in support of the author's evaluations. By pointing to the decarbonized surface, encased in a heavy layer of oxides the authors conclude that the process of liquification of the mass of an individual part of

cast iron proceeds from the interior.

Is it then inferred that this oxide-decarbonized shell, having a melting temperature in excess of that required for the encased metal, bursts at the moment when the liquid pressure of the contained mass exceeds the resistance of that shell? Could it be possible that the reproduced evidence is rather an oddity resulting from adverse reactions during the passage of that piece through a highly oxidizing zone in the vicinity directly above the tuyeres?

Since there exists a difference of 0.45 per cent, admittedly absorbed by the metal in regions above the tuyeres, the question arises concerning the mechanics of C absorption on the basis of contrary evidence. The two adversely reacting factors are in-

compatible.

5. The experiments by Rambush and Taylor produced the hypothesis that a metallic component of the charge enriches itself with C on the surface during the descent in the cupola until a minute increment of the mass-surface reaches a composition of such liquidus temperature which corresponds to the existing temperature at a given level within the furnace, whence a sufficient quantity is reached to form a metal drop. Following its passage a less enriched surface is exposed for continuance of the reaction. The presented investigation negates this theory.

6. The distribution curve of probable temperature, Fig. 3 indicates that the pig iron used began to melt at a temperature of 2050 F at a level of 24 in. above the lower tuyeres and was completely liquified at a level of 22 in. or, a temperature of 2200 F. The quoted pig iron is one of hyper-eutectic composition, having a C.E. value of 4.65, which suggests a higher temperature of liquidus than quoted in the graph even though

the C.E. value is not a criterion for the liquidus.

7. The same curve places the beginning of liquification of the steel (0.58 per cent C) at a level of 18 in. and designates completion at the 15-in. level corresponding to a temperature band of from 2400 F to 2600 F. Reference literature quotes the temperature of liquidus for a 0.58 per cent steel as being from 2700 to 2714 F. Without any C absorption taking place how can this difference in melting temperature be explained? Is it to be understood that within a 3-in. deep temperature zone amounting to 200 F the steel absorbed C spontaneously to a degree that corresponds to the cited temperatures in the curve?

The work by Rambush and Taylor shows that in the retrieved steel bar, lowest down in the cupola, a C increase of 0.32 per cent was experienced, this increase extending within the center of the specimen. How can these two investigation results be

reconciled?

¹ Faraday Engineering Co., Tucson, Ariz. ² Minster Machine Company, Minster, Ohio.

It is possible that because of the continuous bombardment of the steel piece by the droplets from pig or cast iron some free graphite may be found on the surface of the steel because of the rapid quench. The authors attribute a "washing action" to the mode by which steel melts. Just what is meant by that term or its implication is not entirely clear and calls for further elucidation on that subject matter.

8. In the paper the authors state that a 42-in. coke bed height was employed. With the highest temperature zone being confirmed to a band 3 in. high and having its peak at 18 in. above the tuyeres the question arises of what happened to the bed coke above the 24-in. level and what forces drove the steel scrap through a layer of solid coke 14 in. deep.

MR. LOWNIE, JR.: In answer to the question regarding coke bed height, the original height was 42 in. of incandescent coke above the tuyeres before charging was started. In operation the bed burned down to an operating height of 15 in.

With regard to the question on the sugar-tea analogy, we simply wanted to emphasize the fact that metal could be put in

solution below its melting point.

John Grennan: We ran similar experiments which we described in a paper presented at the A.F.S. Convention in Milwaukee in 1924. In the earlier experiment peep holes were cut in the shell. We could visually observe the melting of iron and steel. Our observations agreed with those of Mr. Lownie, although some of our conclusions differed. Melting was seen to be associated with the channeling effect on the gases which varied as the stock shifted. Pig iron and cast melted in a heavy envelope of slag while steel was free of slag because of its higher melting point. The ability of the metal charge to absorb heat from the gases was a governing factor in melting. Steel would melt faster than cast if sufficient surface area were exposed. In the melting stage, 0.8 to 0.9 per cent carbon was absorbed by a low carbon steel.

MR. LOWNIE: Did you find that pig iron melted higher in the cupola than steel?

MR. GRENNAN: Pig iron 4 in. thick melted in the same location as steel of 11/2-in. thickness. Surface volume was an important factor affecting melting speed. Melting was governed primarily by absorption of heat.

MR. LOWNIE: It is agreed that section size is very important. In our experiments the materials were of uniform cross-section. There probably was a slag coating although we had no means of observing this. In general our observations are in agreement with those of Prof. Grennan.

W. C. CORBEAU: 4 Why was 950 F selected as the blast temperature and would the same results be obtained with lower blast temperature?

MR. LOWNIE: 950 F was chosen simply because of general interest in that type of operation. I believe the same conditions would prevail but take place higher in the stack with lower blast temperatures.

MEMBER: What were your wind rate and coke ratio? Would conditions be expected to differ in a large cupola because of non-uniformity? The coke ratio stated in the paper was 7.5 to 1 and blast rate 12.5 lb of air per min.

MR. LOWNIE: Because of uniform conditions in our cupola channeling was not a problem. We expect that the same general behavior would occur in a larger cupola with possibly some difference due to differences in section size.

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 Melt Supt., National Malleable & Steel Castings Co., Cleveland.

# SOLIDIFICATION OF NODULAR IRON IN SAND MOLDS

By

R. P. Dunphy* and W. S. Pellini**

#### Introduction

In order to deduce the casting wall growth characteristics of metals it is necessary to know the microfeatures of solidification and the thermal gradient conditions which are developed in the casting at various times during solidification. The characteristics of solidification of gray iron in sand molds have been previously reported by the authors.3 This investigation showed that dendrite formation throughout the entire cross-section precedes eutectic reaction in gray iron castings. The reaction proceeds in a general fashion being progressive from the sand wall only in the sense that dendrite formation is more advanced nearest the mold well. An interlaced dendrite structure with intermixed liquids is accordingly developed. Eutectic reaction proceeds differently in that it is highly progressive—a narrow band of eutectic reaction travels across the casting section to complete solidification. Variations of superheat, carbon equivalent, mold composition, etc. were shown to affect the rate but not the nature of solidification.

The basic features of the solidification of iron on a micro-scale have been studied by Boyles¹ and Eash.² These investigators definitely established that gray iron solidifies by the prior formation of either the hypo- or hyper-eutectic phase depending on the analysis thus causing either an enrichment or impoverishment in the carbon content of the remaining liquid which gravitates to eutectic composition. Thereafter, solidification proceeds by the formation and growth of graphite flake colonies or by reaction to ledeburite (Fe₃C + austenite).

The nature of the eutectic reaction is determined both by the rate of heat extraction which, if fast, favors the formation of cementite and by the presence of elements which may either favor or preclude the formation of carbide—in any case undercooling is a prerequisite to the formation of carbide structures. Flake structures in commercial Fe-C-Si alloys form

over a narrow eutectic reaction range rather than at a fixed temperature since the classical concept of an invariant eutectic temperature applies only to pure Fe-C alloys. The eutectic ranges for gray iron alloys of various silicon contents have been the subject of considerable investigation and are accordingly relatively well established.

A recent investigation of the microfeatures of nodular iron solidification by the authors⁴ provided essential information prerequisite to the present study of the macro solidification mode. It was concluded that the sequence of events occuring during the solidification of hypo-eutectic cast iron properly treated with magnesium can be outlined as follows:

STAGE 1: The solidification of dendrites of austenite beginning at the liquidus temperature and continuing to the temperature at which the normal eutectic consisting of flake graphite and austenite would occur.

STAGE 2: Suppression of the normal flake graphite eutectic during which interval graphite precipitates from localized regions of supersaturated interdendritic liquid in a spherulitic crystallization mode.

STAGE 3: Solidification of the remaining eutectic liquid as an austenite-iron carbide complex and rapid malleabilization of the structure resulting in growth of the graphite particles. The number of nodules remains invariant during the eutectic solidification period.

Figure 1 abstracted from the report of the previous investigation⁴ shows that the cooling curves of gray and nodular irons are closely similar expect for the undercooling and the more gradual slopes toward the end of solidification exhibited by the nodular iron curve. Figure 2 from the previous report shows micrographic evidence (samples quenched during solidification) of the development and subsequent malleabilization of the eutectic carbide complex.

Information as to the mode of solidification of nodular iron from sand walls would be valuable to the practicing foundryman in his every-day problems with this new material. The present investigation was accordingly aimed at establishing this information and determining the effect of important foundry variables.

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The statements made in this paper represent the opinions of the authors and not necessarily those of the Navy Dept.

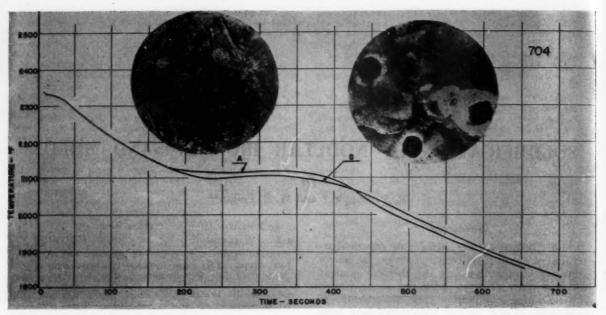


Fig. 1 (Above)—Cooling curves of 2-in. diam sand castings. Mag. 300x.

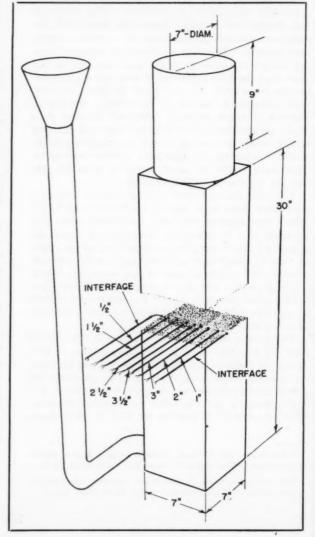
#### **Experimental Procedure**

The irons reported herein (Table 1) were prepared by induction melting charges of Armco iron, graphite, ferrosilicon, and ferromanganese in amounts calculated to produce the desired composition. Magnesium additions were accomplished by tapping onto a ferrosilicon-magnesium alloy containing 11 per cent magnesium. The sand used was a synthetically-bonded sand of A.F.S. Fineness No. 80 containing 3 per cent bentonite, 1½ per cent dextrine, ½ per cent mogul, and 3½ per cent water. The effect of sea coal was established by adding 6 per cent sea coal to this base sand.

The test castings measured 30 in. high and 7 x 7 in. in cross-section. Details of the casting and of the gating and risering methods are shown in Fig. 3. The sand was molded in a 24-in, square flask which provided a depth of  $8\frac{1}{2}$  in. of sand on all four faces of the castings.

The course of solidification of the test castings was followed by observing the temperature changes through the thickness of the casting at its vertical midpoint. A series of Pt, Pt-Rh (13%) thermocouples were cantilevered from the mold wall to the center of the casting according to the arrangement shown in Fig. 3. The couples were enclosed in alundum insulators and sheathed in 6-mm OD, 3-mm ID fused silica protection tubes. The metal interface temperatures were recorded by placing a sheathed thermocouple at the interface so that one-half of the protection tube was in the metal and the other half in the sand. An additional bare couple was placed so that the wires were entirely in the sand and the bare bead at the

Fig. 3 (Right)—Details of casting and location of thermocouples.



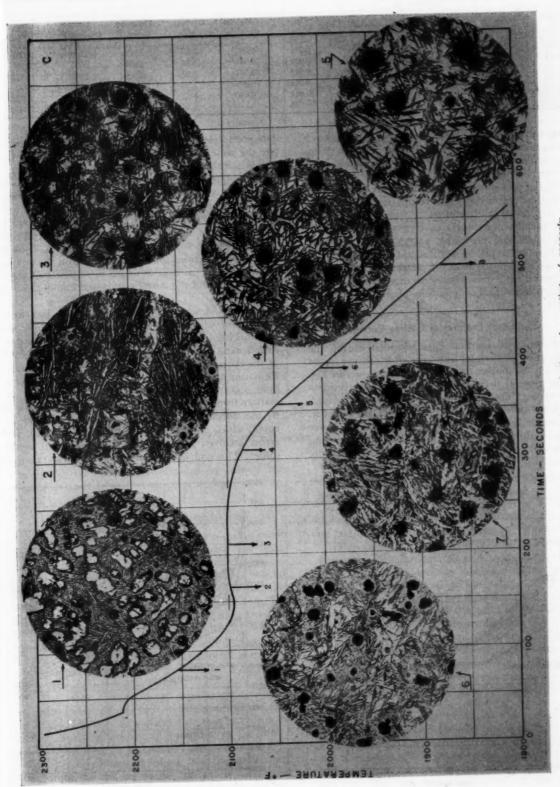


Fig. 2—Cooling curves and quench times for castings of properly treated hypo-eutectic iron (Heat C). Mag. 300x.

TABLE 1

		Analysis, %								
Heat C No.	С	Mn	Si	P	S	Mg	Carbon Equiv.	Liquidus Temp. F	Pour Temp. F	Eutectic Temp. F
C-20	2.95	0.48	2.12	0.015	0.015	0.037	3.66	2247	2500	2105
C-21	3.08	0.50	2.41	0.014	0.012	0.047	3.88	2200	2500	2103
C-22	3.04	0.47	2.33	0.014	0.013	0.047	3.82	2215	2395	2102
C-23	3.04	0.52	1.99	0.016	0.014	0.049	3.70	2236	2500	2092
C-24	2.87	0.49	2.24	0.016	0.011	0.040	3.62	2260	2490	2096
C-13	3.04	0.53	1.39	0.014	0.046		3.50	2286	2500	2109

casting surface. The bead was coated with a thin layer of silica flour mold wash. This couple has been designated as the mold interface couple. The test casting and experimental techniques were exactly the same as used in the previous investigation of gray iron solidification, thus permitting exact comparison of the principal features of solidification.

The metal temperatures were recorded on a multipoint, automatic recording potentiometer, 1500 to 3000 F scale having an accuracy of 0.25 per cent of full scale and sensitivity of 0.1 per cent. The temperature recording sequence was arranged so that readings from individual couples were obtained at approximately 8-sec intervals.

#### Significant Features of Cooling Curves

Representative thermal data recorded during the solidification of nodular and gray iron castings (C-24 and C-13), of similar base analysis, are shown in Fig. 4. The general shapes of the curves are nearly identical—each exhibiting liquidus and eutectic "holds." Significant differences in the two curves are noted in the eutectic hold temperature which is slightly lower for the nodular iron and in the slopes of the cooling curves at the end of solidification which are markedly

less precipitous for the nodular iron. These features are the same as indicated in Fig. 1 for the case of the micro solidification studies.

The sequence of events during the solidification of these irons as indicated by the cooling curves consists of the removal of superheat, the formation of austenite dendrites at the liquidus hold temperature, the growth of these dendrites, and the beginning and ending of eutectic solidification. The timing of these events at the various space positions within the casting is indicated by the time at which these positions reach the liquidus and the eutectic start and end-of-freezing temperatures.

The start of austenite dendrite formation was taken as the time when a significant drop (2 F) below the liquidus hold was observed; the end of austenite dendrite formation as the time an approach to within 2 F of the eutectic hold was reached; the start of eutectic formation as the time a drop of 2 F below the eutectic hold occurred. The choice of a 2 F approach or deviation from the hold temperatures was necessitated by the difficulty of observing a point of merger for curves which approach or deviate asymptotically. The errors involved are small and of no practical significance. The end of eutectic formation was taken as

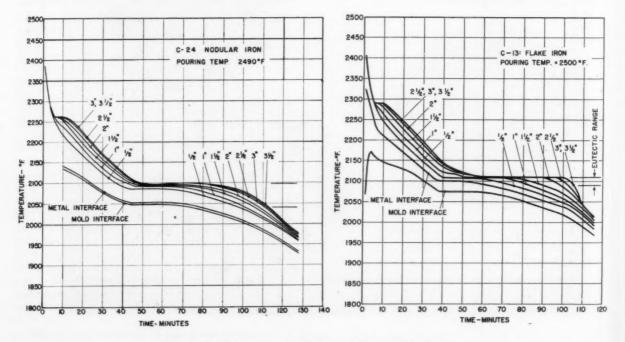


Fig. 4—Cooling curves of nodular and flake graphite irons.

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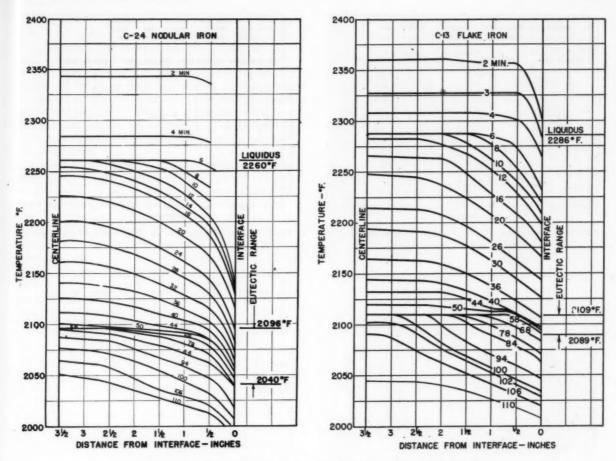


Fig. 5-Thermal gradient curves for nodular and flake graphite irons.

the solidus temperature determined by inverse rate curves obtained from the temperature cycle recorded by the center couple.

#### Comparison of Flake and Nodular Iron

Discussion and visualizations of solidification events are very much simplified by recognizing the waveform progression through the casting of the various cycles of transformation which occur during solidification. The wave like aspects of solidification have been discussed in detail in previous reports.^{3,5,6} The sequence of solidification cycles which have been reported for gray iron is followed also by the nodular iron as indicated by replotting the cooling curve data as thermal gradients which are established across the castings at various times during solidification, Fig. 5. Briefly, the sequence is as follows:

1. The rapid elimination of superheat.

2. The initiation of dendrite formation at the casting wall with the subsequent development of a liquidus "hold" or isothermal condition of the liquid metal in the casting interior. This stage represents the initiation of movement of the "dendrite start wave" from the casting wall.

3. The "dendrite start wave" moves across the section, gradually encroaching on the isothermal region.

4. The "dendrite end wave" denoting completion

of dendrite formation does not move away from the casting wall until fairly extensive dendrite formation has occurred at the center of the casting.

5. The "eutectic start wave" moves outward from the casting wall considerably in advance of the "eutectic end wave" completing its travel to the center prior to the completion of eutectic reaction at the surface. It is only after the eutectic end wave leaves the casting wall that a completely solid skin is developed.

Plotting of the travel of the various solidification waves across the casting, as shown in Fig. 6, permits a detailed comparison of the solidication features of gray and nodular irons. It is observed that the rate of progression of the start of dendrite freeze waves is the same for the two irons. The rate of growth of dendrites during the time the growth is not influenced by eutectic reaction is apparently the same inasmuch as the dendrite end waves coincide at initial stages of their travel.

Differences are noted after this point, however, in that the eutectic end wave in the gray iron casting follows closely behind the eutectic start wave producing a band of active eutectic solidification approximately 1 in. wide, denoting highly progressive solidification. The eutectic solidification of the nodular iron casting occurs in a broad band fashion similar to the dendrite growth stage. The broad band of

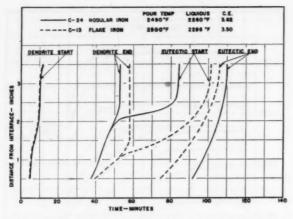


Fig. 6—Comparison of progression of solidification for nodular and flake graphite irons.

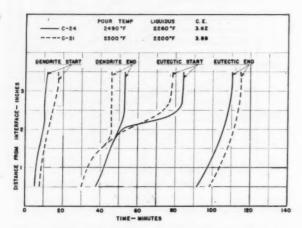


Fig. 8—Effect of carbon equivalent on progression of solidification of nodular iron.

eutectic solidification of the nodular iron results from the faster travel of its eutectic start wave and slower travel of its eutectic end wave as compared to the gray iron. As a final result it is observed that the completion of solidification of the nodular iron casting occurs at somewhat later time than for the gray iron.

The mechanism by which the differences in the rate of travel of the nodular iron eutectic waves as compared to gray are established is not known. It is suspected that the two-step reaction features of nodular iron eutectic solidification involving the prior formation of carbide structures followed by malleabilization to nodular graphite is responsible. It is difficult to theorize how this occurs, however, since a variety of poorly understood factors such as differences in the thermal conductivity of the matrix, liberation of heat of transformation due to malleabilization at post eutectic reaction times, etc., may play a part.

#### Effect of Variables on Solidification

The effect of superheat on the general solidification characteristics of nodular irons, Fig. 7, can be seen by comparison of two castings (C21 and C22) poured with a superheat difference of approximately 120 F.

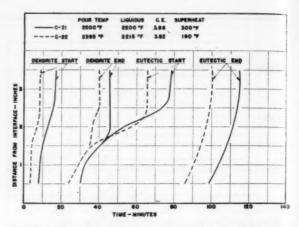


Fig. 7—Effect of superheat on progression of solidification of nodular iron.

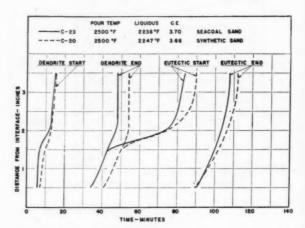


Fig. 9—Effect of sea coal on progression of solidification of nodular iron.

The effect of superheat is a progressive delay in the times at which the various solidification waves travel across the casting section. For example the dendrite "start" and "end" waves in casting C-21 are delayed approximately 5 min, whereas the eutectic "start" and "end" waves are delayed 7 and 13 min respectively. The cumulative effect of these delays is to increase the final solidification time considerably more than would be expected from the single initial delay in the start of solidification. These effects result from the preheating effect on the mold due to the flow of superheat prior to the solidification interval and are exactly analogous to the effects of superheat determined for gray irons³ and steels.⁵

The relative solidification characteristics of nodular irons of different carbon equivalent when poured from the commonly used temperature of 2500 F are shown in Fig. 8. The separate roles of differences in superheat and carbon equivalent are readily recognizable in the two sets of curves. The basic effect of higher carbon equivalent is a shortening of the time period between the start and end of dendrite formation and lengthening of the time period required for

eutectic reaction. Thus, the effect of carbon equivalent is the same as shown previously for flake irons. The somewhat higher superheat condition of the high carbon equivalent iron is reflected in an initial delay in the start of dendrite formation and somewhat longer (4 per cent) time of complete solidification. It is apparent from consideration of final solidification times that the small changes in superheat effected by the common practice of pouring irons of moderately different carbon equivalent at a fixed temperature is relatively small.

The effect of a 6 per cent sea coal addition to the sand on the progression of solidification of nodular irons is shown in Fig. 9. Comparison of castings C-20 and C-23 shows that the timing of the various solidification cycles is decreased by the presence of sea coal in the sand in a manner directly opposite to that of superheat. The cumulative effect is a moderate (4 per cent) decrease in the final solidification time. The increased rate of progression of the solidification waves occasioned by sea coal necessarily indicates an increased rate of heat removal by the mold. This feature of sea-coal sand was also observed in the case of flake irons.

#### Microstructures

Typical microstructures at the centers of representative castings are shown in Fig. 10. In general the graphite nodules were evenly distributed with no significant change in ferrite-pearlite ratio across the section. The tendency toward retention of eutectic carbides in film-like distributon increased slightly toward the center of the section; inclusions were generally associated with the carbide regions. Superheat and mold variables did not have significant effects on the nature or distribution of the microconstituents.

#### Acknowledgment

The authors wish to express their appreciation for the active support of Dr. O. T. Marzke, Superintendent of the Division, and the assistance extended to them by T. W. Snodgrass. The assistance of A. R. Donaldson of the Metallographic section of the Laboratory in preparing the photomicrographs is likewise appreciated.

## **Bibliography**

1. A. Boyles, The Structure of Cast Iron, American Society for Metals, Cleveland (1947).

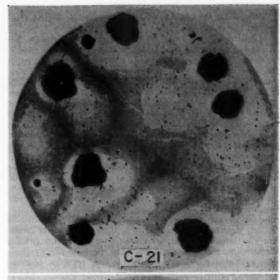
2. J. T. Eash, Transactions, American Foundrymen's Society, vol. 49, p. 887 (1941).

3. R. P. Dunphy and W. S. Pellini, "Solidification of Gray Iron in Sand Molds," Transactions, A.F.S., vol. 59, pp. 425-434 (1951).

4. R. P. Dunphy and W. S. Pellini, "Nodule Genesis and Growth in Magnesium-Treated Hypoeutectic Irons," *Memoire No. 8*, Congres International de Fonderie, Brussels, 1951.

5. H. F. Bishop, F. A. Brandt, and W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," Transactions, A.F.S., vol 59, pp. 435-450 (1951).

 H. F. Bishop and W. S. Pellini, "Solidification of Metals," The Foundry, Feb. 1952.





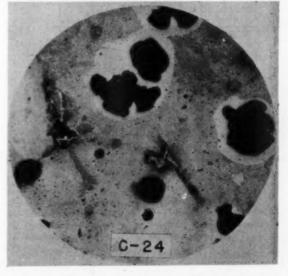


Fig. 10—Typical microstructures of test castings.

Mag. 100x.

#### DISCUSSION

Chairman: H. A. Schwartz, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: E. C. TROY, Foundry Engineer, Palmyra, N. J. Recorder: C. E. Sims, Battelle Memorial Institute, Columbus.

A. L. DE Sy (Written Discussion): 1 We have read this paper with a great deal of interest and think that special attention should be given to the fundamental difference in the shape of the cooling curves (Fig. 4) of nodular and flake graphite irons.

Our results are in perfect agreement with those of R. P. Dunphy and W. S. Pellini as far as the slope of the cooling curves at the end of solidification is concerned; however, concerning the so-called eutectic hold of nodular iron, accurate temperature measurements show a slight drop (approximately 1° C) during what is considered as the hold.

The abrupt termination of the hold in the case of flake graphite irons is very characteristic of a eutectic solidification; on the contrary the very progressive slope at the end of the cooling curves of nodular iron can hardly be considered as corresponding to a real eutectic.

Should we not consider the graphite spherulite growing in (initially supersaturated) austenite by direct diffusion of carbon from the liquid through an austenite layer to the growing spherulite. Such a solidification does not correspond to a eutectic and would give a cooling curve of the type regularly observed with nodular iron.

For further details a recent paper* by A. L. De Sy and J. Vidts should be consulted.

V. PASCHKIS (Written Discussion): 2 In both parts of Fig. 4 a measurable temperature difference between metal and mold interface is shown. The temperature drop may be considered as an expression of thermal resistance across the gap.

In case of nodular iron the temperature difference is approximately 5 F, in case of the flake iron the difference 10 minutes after pouring is 70 F and then decreases slowly to a more or less constant value of 30-40 F. Thus, if the heat flow in both cases is about the same, the thermal resistance in case of the flake iron is 70/5 = 14 to 30/5 = 6 times as high as in the case of nodular iron.

The resistance across the gap consists actually of two parallel resistors: one for radiation and one for conduction. The former is practically independent of the width of the gap, the other is proportional to the width.

Thus a lower resistance (flake iron) indicates a greater width of the air gap, and a greater width, in turn must be attributed to higher shrinkage. Thus, evaluation leads to the conclusion that nodular iron shrinks much less than flake iron, whereas actually nodular iron has a larger shrinkage.

Thus the writer would conclude that the temperature measurements on the faces of the gap are perhaps not entirely reliable. This is not surprising since surface temperatures are most difficult to measure.

It is interesting to note that this difference of shrinkage shows up, even if the influence of an airgap on solidification rates is small in case of casting in sand molds.

This is understandable, because any conceivable value of airgaps resistance is small compared with the resistance of the sand mold.

^{*} A. De Sy et J. Vidts, "Précisions et données complémentaires sur le canisme de formation des sphérolites de graphite," Fonderic Belge, mécanisme de formation des sphérolites de graphite, March 1952. ¹ Prof., Laboratorium Voor Metallurgie, Liege, Belgium. ² Columbia University, New York.

# A SOLIDIFICATION DILATOMETER AND ITS APPLICATION TO GRAY IRON

By

R. P. Dunphy* and W. S. Pellini*

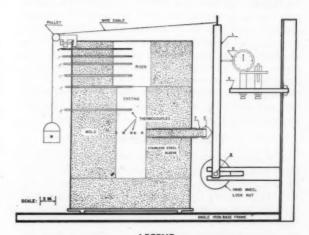
#### Introduction

For many years risering has been recognized as an art based on the hard-won experience of foundrymen. Recently scientific methods have been applied to the problem of risering steel castings.¹ Extensive research and foundry trials were required to reach the present state of development; continued progress is being made with the active support of progressive foundries. A similar scientific approach to the problem of risering gray iron appears to be timely inasmuch as a clearer understanding².³ has been evolved of the specific solidification features of steel and gray iron, thus allowing translation of information from one case to the other.

Basically, the physical conditions which exist during the first stage of hypoeutectic cast iron solidification are the same as for steel, since both are characterized by dendrite formation. The liquid and liquidto-solid volume changes occurring during the dendrite solidification stage require the inflow of feed metal which, if not available, results in the development of shrinkage pockets. Graphite formation during late stages of solidification distinguishes gray irons from steels, however, since the increased volume accompanying the formation of this low density constituent acts to compensate for the liquid-to-solid shrinkage occurring during the dendrite formation stage. The development of risering formulas for this material is consequently difficult, requiring allowance for the complex interbalancing of liquid-to-solid shrinkage and eutectic expansion.

An empirical approach to such a complex problem by correlation of shrinkage, dishing, expansion, etc., of various casting shapes involves an uninviting expenditure of effort with questionable chance of success. A systematic step-by-step approach involving initially the analysis of the features of solidification, metal movement and dimensional changes of simple systems amenable to exact study appears more feasible, economical, and technically sound. This latter course was followed in the present investigation, which represents an initial step in the stated direction with the principal aim of deducing the mechanism of casting wall movement. Recent thermal analysis studies of the features of gray iron solidification provided necessary basic information for these studies.³ There remained the problem of combining established thermal analysis techniques with a physical measurement to relate dimensional changes to the solidification cycles of the casting—this was accomplished by the development of a solidification dilatometer.

The principal features of the solidification dilatometer system are illustrated in Fig. 1. A 4 x 4 x 12-in. casting is molded in 16 in. square cheek flask. The end of a sand-filled fused silica tube (1½-in. op, 3/82-in. wall thickness) is positioned at the mold-metal interface. The opposite end is covered with a conical metal cap, the apex of which rests against a lever arm. A weight (15 oz net) is attached to the top of



# L—Lever Arm D—Dial Indicator S—Shaft Supporting Dial Indicator and Adjusting

Mechanism

C—Tube Cap
B—Lever Fulcrum Bearing
W—Weight

T-Sand Filled Fused

Silica Tube

Fig. 1—Details of solidification dilatometer.

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Note: The opinions expressed in this paper are those of the authors and do not necessarily reflect the views of the Navy Dept.

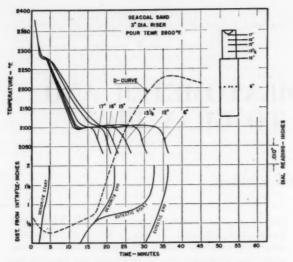


Fig. 2—Thermal and dilatometer data recorded during solidification of 3-in. risered casting in seacoal sand.

the lever by means of a wire cable and serves to maintain intimate casting-tube contact at the mold-metal interface; the lever and fulcrum arrangement produce a load on the casting surface of 2.26 psi. The tube is surrounded with sand near the mold-metal interface to a depth of 3/4-in. At this point a stainless steel tube guide of slightly larger diameter encloses the silica tube along its entire length to the point of its emergence from the flask. A dial gage, calibrated in 0.0001 in., is used to indicate the tube movement, which is magnified  $\times 2$  by the lever connection.

The molds are rammed with an air rammer, particular attention being paid to the attainment of uniform mold hardness. In order to prevent tube breakage during ramming, the sand immediately around the tube end was hand-rammed with a pencil rammer to a circumferential depth of approxiamtely 1½ in., followed by air ramming of the succeeding sand layers. This procedure produced no differences in mold hardness as compared to the remainder of the mold. After ramming of the mold was completed, free movement of the tube was insured by rotation within the ¾-in. sand envelope at the tube end.

#### Thermocouple Locations

The relationship between the metal interface movement and the timing of the solidification cycles of the casting and riser is obtained by cantilevering Pt, Pt-Rh (13 per cent) thermocouples from the mold wall into the mold cavity. The thermocouples sheathed in 3/16-in. op fused silica protection tubes are positioned at various locations in the vertical centerplane of the casting and riser in order to follow the course of solidification of the system in the vertical direction. An additional series of thermocouples is placed in the horizontal plane at the casting midheight (6 in.) to follow the progress of solidification at the point where the dilatometer tube contacts the casting. These thermocouples are located at the mold-metal interface and at 1/2, 1, 11/2 and 2 in. from the mold wall. Temperatures are recorded by means of a multipoint automatic recording potentiometer, 1500-3000 F scale, having an accuracy of 0.25 per cent of full scale and sensitivity of 0.1 per cent.

The general procedural details of all experiments in this investigation involved bracing of the lever arm in a rigid position during pouring, after which the lever is released and allowed to move freely. Dial indicator and temperature readings are taken simultaneously until complete solidification of the casting and riser has occurred.

# Experiments

The gray irons used throughout the investigation were prepared by induction melting charges of armco iron, ferrosilicon, graphite, and ferromanganese in amounts calculated to give the desired composition. All irons were heated to 2650-2700 F, inoculated with 0.50 per cent silicon as ferrosilicon (50 per cent type), and poured at 2600 F. The irons were hypoeutectic in composition, falling in the chemical analysis range of 3.00-3.07 per cent C, 0.38-0.43 per cent Mn, 1.32-1.77 per cent Si, 0.011-0.020 per cent S, and 0.011-0.018 per cent P. All pouring temperatures were determined by immersion thermocouples. The sands were synthetically bonded containing 3.0 per cent bentonite (western or southern), 11/2 per cent cereal, 0.5 per cent dextrin, and 3.5 per cent water. In cases where sea coal was used, 6 per cent of sea coal was added to the western bentonite base mixture. In each test series the riser diameters were varied from 3 to 6 in, in 1-in, increments; the riser height was kept constant at 6 in. The tops of all risers were covered with an exothermic, anti-piping compound, over which was placed a  $\frac{1}{16}$ -in. sheet of asbestos.

The thermal and dilatometer data from the sea coal sand test series are shown in Figs. 2, 3, 4, and 5. Each figure presents 1) the dilatometer data (D curve); 2) the cooling curves for various height positions at the casting centerline indicating the time sequence of complete solidification; 3) thickness-time curves indicating the progression of the dendrite and

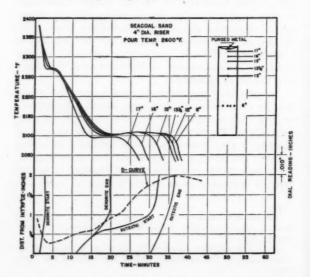


Fig. 3—Thermal and dilatometer data recorded during solidification of 4-in, risered casting in seacoal sand.

eutectic solidification waves across the casting section at the midheight position (6 in.). Construction of the thickness-time curves is accomplished by noting the time at which the various space positions within the casting reach the dendrite and eutectic start-andend-of-freezing temperatures as described in the previous reports.8 By means of this plotting method the solidification process is represented in an easily visualized form. The most significant aspects of the complex solidification mechanism of hypo-eutectic gray iron is the two-stage solidification cycle. The first stage featuring dendrite formation requires feed metal flow to compensate for liquid-to-solid shrinkage, while the second stage featuring eutectic reaction does not require feed metal flow inasmuch as the formation of graphite results in a net volume increase. The two cycles do not follow in immediate sequence at any given casting position—on the contrary, the first cycle essentially goes to completion throughout the casting before the start of the second cycle. This feature is a most important consideration in the analysis of gray iron risering requirements, as will be discussed in sections to follow.

#### **Expansion Registered**

A major feature of the dial indicator records (D-curves) for the western bentonite-sea coal series is the marked casting expansion registered for the casting with the 3-in. riser during the eutectic reaction interval. The D-curve for the next larger riser size (4 in.) exhibits moderate expansion during this interval while the next two riser sizes (5 and 6 in.) show a comparatively negligible expansion.

Detailed examination of D-curves for these castings shows that a forward movement of the tube occurs during the period which precedes the development of a coherent dendrite mass at the casting surface. At this early time the frictional effects of the sand surrounding the tube provides the major resistance to tube movement. The forward movement gradually decreases and ultimately stops as the casting surface attains sufficient coherency to resist the applied load.

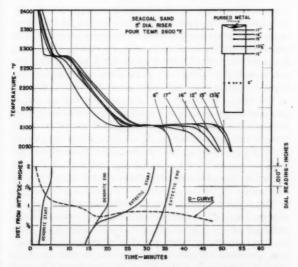


Fig. 4—Thermal and dilatometer data recorded during solidification of 5-in. risered casting in seacoal sand.

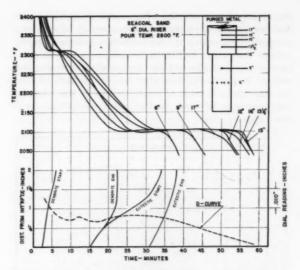


Fig. 5—Thermal and dilatometer data recorded during solidification of 6-in. risered casting in seacoal sand.

Thermal expansion of the tube due to heating may give slight indications of the backward movement during the following interval of completion of dendrite growth. These changes are small, however, and in view of the lack of precise agreement in the curve shapes during these early periods, an exact analysis of the responsible factors is not possible. During the eutectic solidification interval it is noted that the marked casting expansion of the 3-in. riser system occurs at an initially rapid rate, with a gradually decreasing rate being manifested near the end of eutectic solidification. After completion of the travel of the eutectic end wave across the casting, contraction occurs. It is also noted that the onset of marked casting expansion coincides with the time of complete solidification at the top surface of the riser (17-in. position). This occurrence may be considered as "capping," or closing of the casting system, analogous to the steel mill practice of capping of ingots by "freezing-off" the top surface.

#### Riser Size and Expansion

The 4 in. diameter riser system also shows an increase in expansion rate at the time "capping" occurs. The timing of actual capping is somewhat delayed (24 min as compared to 17 min). The rate and extent of expansion are notably decreased as compared with the 3 in. risered casting. Again, contraction occurs at the time of complete solidification of the casting. In the case of the 4-in, riser a small amount of purged metal was observed on top of the riser after shakeout—no purged metal was observed in the case of the 3-in. riser. The castings made with 5 and 6 in. diameter risers show little expansion during the eutectic solidification interval. It is noted that the topmost thermocouple positions for these two castings indicate that the riser surface solidifies at a later time than the casting-capping accordingly did not occur. Large amounts of purged metal were observed on the tops of both risers.

Summary D-curves for the western bentonite-sea coal series of experiments along with the D-curves for

two series of experiments wherein southern and western bentonites were used in the absence of sea coal are presented in Fig. 6. The notations on each curve denote the start of eutectic solidification at the surface of the castings (at the midheight position), the end of eutectic solidification at the center of the castings, and the time of solidification of the topmost thermocouple position in the riser which represents the capping time.

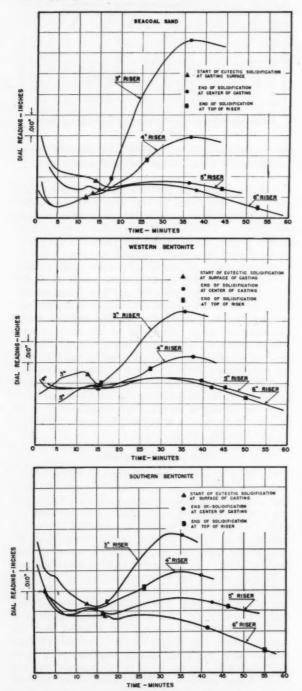


Fig. 6—Summary D-curves for experiments with seacoal, western and southern bentonite-bonded sands.

General examination of these data shows that regardless of the molding medium an expansion of the casting occurs during eutectic solidification in the castings made with 3 and 4 in. diameter risers, and that the amount and rate of development of casting expansion decreases as the riser diameter increases. It is also noted that for the 3 and 4 in. risers in all three sand series the riser freezes, or caps, prior to the time of complete casting solidification, and that the riser capping-casting solidification sequence is reversed for the 5 and 6 in. diameter riser sizes. In each sand series purged metal was observed on tops of all risers except the 3 in. riser size, the amount increased with increase in riser size from the 4 to 6 in. riser diameter.

#### **General Discussion**

The controlling effect of riser size on the nature of wall movement during eutectic reaction clearly indicates the development of a pressure condition within the casting resulting from the volume expansion consequent to eutectic reaction. The pressure which is developed is potentially of a high order, sufficient to

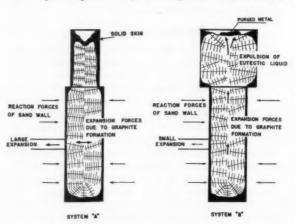


Fig. 7—Schematic representation of effects of graphite formation in "capped" and "uncapped" systems.

deform the sand walls if the casting is capped by solidification of the riser surface in advance of the time of eutectic reaction within the casting proper. In the case in which the riser remains open during the eutectic reaction period, the general level of pressure remains low by virtue of expulsion, or purging, of eutectic liquid through the interdendritic channels of the riser. For cases of riser capping during intermediate stages of eutectic reaction in the casting, partial relief is afforded by purging, and moderate pressures result with consequent moderate casting expansion. It is thus apparent that the anomalous behavior of gray iron in producing oversize castings is related to the pressure level developed during eutectic formation.

These generalizations are fully consistent with the observations of the various riser capping times and the amount of casting expansions and purgings. The known fact that purged metal is of eutectic composition provides further substantiation of this mechanism. A schematic illustration of the process is provided in Fig. 7. For a detailed consideration of the

process it is essential to recognize the two-cycle features of hypoeutectic gray iron solidification described by the thickness-time curves of wall growth; also, that the relative rates of solidification are determined by the specific geometries (volume/surface area ratios) of the casting sections. Systems A and B of Fig. 7 represent the same casting with risers which solidify respectively in advance of and following the solidification time of the casting. In either case the period of dendrite formation is marked by liquid metal flow from the riser to the casting. The process is exactly similar to that which occurs in the case of steel.* A short time prior to full completion of dendrite formation at the center, eutectic reaction starts at the surface of the casting moving in wave-like fashion to the center. The relative geometries of the riser and casting of system A are such that the riser develops a solid skin at the top surface at this time. The entire system is thus capped in a fashion similar to capping a container of liquid, with the mold walls forming the container and the solid skin on the riser providing the cap.

#### Pressure and Volume

At this time all interdendritic spaces are filled with eutectic liquid. As the eutectic wave moves forward expansion occurs due to graphite formation and a pressure rise is developed. Depending on the nature of the resisting forces of the sand wall and the rigidity of the cap, the pressure is relieved by an increase in the volume of the casting. This is recognized as a wall movement by the dilatometer used in this investigation. On completion of solidification, solid contraction occurs due to cooling of the casting-the D-curve shows a contraction following the point of final solidification. The check between the D-curve and the thermal analysis, thickness-time curves is almost exact for the systems which show marked dimensional changes, demonstrating mutual confirmation of the thermal analysis and dilatometer techniques.

The relative geometries of the riser and casting of system B are such that the riser does not solidify at its surface until after the casting is completely solidified. Thus, during the period that the eutectic reaction wave moves across the casting a relief to the expansion pressure is afforded through the interconnected interdendritic passages of the casting and riser, with consequent expulsion of eutectic liquid as purged metal which is noted on the surface of the riser.

It should be expected that the reaction or resisting force of the sand wall will determine in part the magnitude of the observed pressure-induced wall movement. It is observed that the largest expansion is developed in the sea coal-western bentonite sand system. The expansion registered for the sea coal-free southern and western bentonite sands is markedly less. The southern bentonite sand indicated a somewhat higher, but possibly not significant, expansion as compared to the western bentonite. The

characteristics of the sand wall movement should be a function of the intrinsic strength and expansion properties of the various sands under the influence of steep and changing temperature gradients. While hot compressive strengths and expansion-contraction cycles of sands undoubtedly contribute to the overall behavior of the mold, the known properties of sands as determined in simple sand tests present no basis for explaining the apparent greater facility for casting wall movement evidenced by sea coal sand. The need for measurements of the properties and be-



Fig. 8—Section of 4x4x12-in. test casting made with 1½-in. diameter riser showing gross shrinkage cavity. Arrow denotes purge drop.

^{*} If the smaller riser solidified completely at an early stage of dendrite growth of the casting (faster than shown for the 3-in. risers in this investigation), feed metal flow would be cut off with consequent development of shrinkage cavities (Fig. 8).

havior of sands under actual casting conditions involving dynamic temperature changes is indicated.

The relationships of solidification and expansion features observed for the simple casting system used in this investigation provides a basis for understanding the feeding requirements of more complex practical castings. It is evident that unlike the case of steel castings for which the feed metal flow path is always counter to the path of directional solidification, the path of feed metal in the case of hypoeutectic gray iron is actually reversed during the eutectic reaction period. Thus, at late stages of solidification steel castings tend to drain heavy sections to satisfy the demands of lighter sections, leading to dispersed shrinkage in the heavy sections. Gray iron castings, however, behave in an entirely different manner for during this same period light sections actually pump liquid back to heavier sections. Purging, expansion, or even refilling of gross shrinkage voids developed during the dendrite formation period of the heavy section may result from the reversed flow of liquid.

The sufficiency of small bobs for gray iron feeding is explained by the described mechanism. Feeding by the bob is only required during the interval of time prior to eutectic solidification. In fact, the feeding need not be complete for a feeding deficiency will permit room for subsequent eutectic expansion. The relative feeding requirements of irons of various carbon equivalent become apparent when analyzed in terms of the relative fraction of volume which solidifies during the shrinkage-producing dendrite stage as compared to the fraction which solidifies during the expansion-producing eutectic stage. These various factors must be considered presently in a qualitative fashion. The obvious next step in the direction of scientific gray iron risering lies in the combination of the principles of steel risering practice with the effects of graphite formation in gray iron and reducing them to quantitative terms.

# Acknowledgment

The authors are indebted to Dr. O. T. Marzke, Superintendent of the Metallurgy Div., for his active support of the investigation, and to T. W. Snodgrass for his capable assistance in performing the experiments.

#### References

- 1. J. B. Caine, "A Theoretical Approach to the Problem of Dimensioning Risers," A.F.S. TRANSACTIONS, vol. 56, p. 492
- 2. H. F. Bishop, F. A. Brandt, and W. S. Pellini, "Solidification of Steel from Sand and Chill Walls," A.F.S. TRANSACTIONS, vol. 59, p. 435 (1951).

3. R. P. Dunphy and W. S. Pellini, "Solidification of Gra-Iron in Sand Molds," A.F.S. Transactions, vol. 59, p. 425 (1951)

#### DISCUSSION

- Chairman: A. P. GAGNEBIN, International Nickel Co., New
- Co-Chairman: R. SCHNEIDEWIND, University of Michigan, Ann
- Recorder: G. A. TIMMONS, Climax Molybdenum Co., Detroit.
- V. PASCHKIS (Written Discussion): 1 This is certainly a most unusual and excellent paper, showing clearly the influence of several variables on the solidification pattern.
- From the background of the writer's research by means of the analog computer, two comments come to mind.
- 1. The authors state, that, in case of steel, the pouring temperature does not materially influence time and mode of solidifications. However, the author's own Fig. 14 (his Ref. 2) indicate apparently quite strong influence of superheat:
- Increasing the pouring temperature from 2725 to 2900 F results in an increase of solidification time from 42 to 50.5 min (green sand). Similarly his Fig. 7 (l.c.) shows an increase of freezing time from 9.4 to 10.6 min when casting against a chill (2724 F vs 2880 F). This trend is in line with our findings on the Heat and Mass Flow Analyser.
- It would be helpful, if the author would show the influence of the pouring temperature in case of aluminum, bronze, etc. Since the solidification temperatures are lower than in case of steel, superheat may have a different influence. What actually is important is not the absolute degree of superheat, but rather
  - (Pouring temperature) Liquidus temperature)
- (Liquidus temperature) (Initial mold temperature) 2. The authors' reference to the spread between liquidus and solidus and the relationship between this spread and the feeding of a casting opens up interesting possibilities for studies on
- the analyser. Feeding is a phenomenon covered by the fields of liquid flow and of heat flow.
- The author seems to indicate that it is possible to determine for each temperature the amount of liquid and of solid material present in the casting. The cross-sectional area not covered by solid material is available for liquid flow. From a knowledge of the viscosity and the ferrostatic pressure of the feeder, the rate of fluid flow can then be determined.
- Since the temperature distribution in casting and feeder can be computed, such sequence of calculations would make the problem of feeding accessible to rational and accurate prediction. The views of the author on this matter would be of great interest.
- H. J. MEERKAMP VAN EMBDEN: 2 What were the dimensions of the final casting as compared with the dimensions of the pattern? I am particularly interested to know if the castings increased or decreased in total volume.
- MR. DUNPHY: This depended upon the solidification of the riser. Castings were in some cases larger than the pattern and in other cases smaller.
- D. E. Krause: 3 Did you make hardness or density measurements on the castings with different sizes of risers? Did the values change with the size of risers?
  - MR. DUNPHY: There was no difference in hardness or density.

Columbia University, New York.
 Chief Engineer, Phillips, Ltd., Eindhoven, Holland.
 Executive Director, Gray Iron Research Institute, Columbus, Ohio.

# INVESTMENT CASTING OF ALUMINUM

By

H. Rosenthal* and S. Lipson*

#### ABSTRACT

An investigation was made of the mechanical properties of aluminum test bars produced by the investment casting tech-

Properties of two alloys were determined and found to approximate those obtained by sand casting methods. The values obtained for a 5 per cent silicon alloy ranged from 20,000 psi tensile strength and 9 per cent elongation (1-in. gage) for the lowest casting temperature (1175 F) to 15,500 psi and 5 per cent for the highest casting temperature (1615 F). These properties were improved by the utilization of chills and by modification of the melt with sodium just before pouring. An alloy which contained 4 per cent copper in addition to 5 per cent silicon was found to develop 23,000 psi tensile strength and 3.8 per cent elongation when the pouring temperature was controlled at 1210 F. This alloy, however, was found to be relatively insensitive to pouring temperature. In order to use the lower mold temperature required for making aluminum investment castings, it was necessary to develop a wax elimination procedure for melting out the wax in an atmosphere of saturated steam.

#### Introduction

The data reported here represent the second phase of a program to obtain alloy properties of investment castings. The work published earlier1,2 concerns representative copper-base alloys. Investment casting of aluminum-base alloys requires some change in the technique which is successfully used for copper-base alloys.

It is well known that the properties of cast aluminum alloys are quite sensitive to the rate of freezing. This indicates the desirability of using low mold temperatures and minimum pouring temperatures. A method for using retained moisture in the mold to produce a chilling action has been described in a previous paper.3

The drawback in the use of low mold temperatures with the standard dry investment is the tendency for cracks to develop in the mold. These cracks are generally caused by the contraction stresses which accompany the lowering of the mold temperature from the point at which wax residue is eliminated.

#### Methods and Materials

1. Specimens-The standard 0.252-in. diameter, 1-

specimens were sectioned and etched to determine grain size and soundness. Specimens for microstructural examination were obtained by cutting a transverse section from the gage length of an unbroken specimen. Hardness data were obtained on the longitudinal section used for the macrostructure examina-Wax patterns of the tensile specimens were made by injecting into a machined brass die having four

in. gage length tensile specimen was used for these

tests. Tensile and elongation data were obtained from

these cast-to-size specimens and a few unbroken speci-

mens were set aside for density determinations. Other

cavities. Figure 1 shows the die with a wax pattern still retained in one half.

The patterns were mounted on a rubber base for a 3-in. diameter mold. Eight patterns were mounted on each base for top pouring and a relatively large sprue was provided to feed the castings. Figure 2 shows the sprue base with the specimens mounted ready for investment.

2. Investment—The patterns were invested with a standard proprietary material. This material consists of approximately 70 per cent silica and 30 per cent plaster of paris. Water added to make the pouring slurry was 50 per cent of the weight of the dry investment. The slurry was vacuumed to remove entrapped air bubbles.

The generally recommended method for eliminating the wax and curing the investment is: (1) heat the mold in an inverted position at 180 to 200 F for 2 hr to melt out the wax; (2) raise the temperature gradually to 1200 to 1350 F to calcine the investment and burnout the residual wax; (3) lower the mold temperature gradually to the level desired for the pouring operation. This procedure is successful with copper-base alloys for which the mold temperature at pouring is generally a minimum of 600 F. However, when the molds are reduced in temperature to the 250 to 350 F level as is desirable for aluminumbase alloys, considerable cracking occurs.

In order to overcome cracking, a reduction in burnout temperature was desirable. By this means thermal expansion and contraction effects, which are associated with cracking, could be kept to a minimum.

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However, even a small reduction in burnout temperature increased the burnout time prohibitively. If a greater amount of wax could be removed than is accomplished by the standard procedure, an appreciable reduction in the burnout temperature would result. The function of the burnout oven would then be restricted to calcining the mold rather than oxidizing the considerable amount of carbonized wax that ordinarily remains in the investment.

When the mold is subjected to the preliminary heating necessary to liquefy and drain the wax, much of the mechanically held water in the investment is driven off, leaving a network of voids throughout the mold. As the wax becomes liquid, these capillaries apparently draw much of it into the mold before the wax has had an opportunity to drain, thus accounting for the small volume of wax collected during this cycle.

It is apparent that the capillaries should be prevented from forming until the liquid wax can drain completely from the mold. By more efficient wax elimination, a reduction in the final burnout temperature would be possible because (a) the amount of carbonaceous residue would be reduced and (b) the residue would be confined to the surface of the mold cavity and consequently could be more rapidly oxidized.

Accordingly, several invested molds were heated in an atmosphere saturated with water vapor. Under these conditions, an equilibrium was established between the moisture in the atmosphere and that in the mold which kept the mold saturated with sufficient water to prevent the capillaries from forming. In

Fig. 1—Exploded view of tensile specimen die showing the wax patterns imbedded in one half.

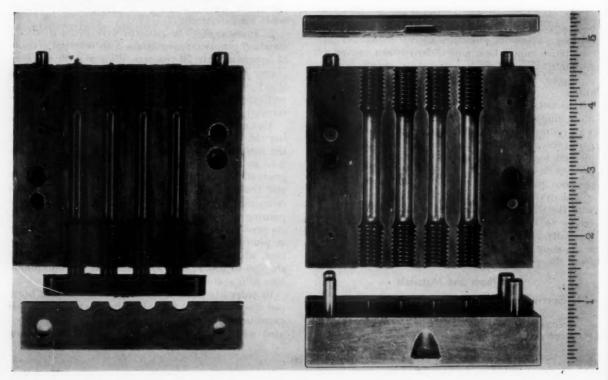


Fig. 2—Arrangement of wax tensile specimen patterns on rubber sprue base.

order to facilitate further the removal of wax, the meltout temperature was raised by performing the operation in an autoclave with superheated steam at 15 psi pressure (approximately 250 F).

The wax recovered from this operation accounted for 98 per cent of the original weight of the wax sprues and patterns.

With molds prepared under these conditions, it was possible to complete the final burnout operation at 600 F for 24 hr or 800 F for 2 hr. When these molds were cooled to the proper temperature for



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casting (250 to 400 F), cracking in most instances was eliminated entirely or confined to very small hairline cracks.

3. Melting and Casting—The metal was melted in a No. 2 clay-graphite bottom-pour crucible by means of 30 kw high frequency induction furnace. No flux was used and the oxide skin which formed on the surface of the molten metal was skimmed away before each pour.

Pressure for casting was obtained by means of a suction casting machine which has been described.³

Various pouring temperatures were used ranging from 10 to 450 F above the liquidus. A mold temperature of 250 F was standardized for this work although several castings were made using molds at room temperature and at 400 F. However, it is not usually practical to allow the flasks to cool much below 250 F because of the hygroscopic nature of the investment.

The effects of copper chills in contact with the castings was determined at the 250 and 400 F mold temperatures. The chills, approximately 1/4-in. diameter and 1/2 in. long, were placed at the extreme end of the specimen by attaching them to the wax and investing the assembled cluster. After melting out, the chills were retained in the investment and were located so that one face would come in contact with the molten metal as it filled the mold.

4. Alloy Composition—The first of the aluminum group to be investigated was the widely used alloy containing 5 per cent silicon. Since this alloy had already found extensive use in this country for sand, permanent mold and die casting, the properties developed by investment casting can be readily compared with the properties developed by the other methods of casting. This alloy is designated in ASTM Spec. B179-51T as Alloy S5A.

In selecting an alloy to represent the ternary aluminum-copper-silicon system, it was decided to determine the effect of the addition of 4 per cent copper to the binary composition being studied. This alloy is designated in ASTM Spec. B179-51T as Alloy SC64A.

Chemical and spectrographic analyses were made of the ingot metal used to cast these specimens and they were found to conform with the specifications. The ingot compositions are shown in Table 1.

#### 1. Tensile Strength and Elongation

a. Al-Si Alloy. Tensile strength and elongation de-

TABLE 1-CHEMICAL ANALYSIS OF INGOT METAL

	Composition, %					
Element	Al-Si Alloy ASTM Spec. B179-51T Alloy S5A	Al-Si-Cu Alloy ASTM Spec. B179-51T Analysis Alloy SC64A Anal				
Cu	0.1 max	0.05	4.0-5.0	4.10		
Si	4.5-6.0	5.22	5.0-6.0	5.05		
Fe	0.6 max	0.53	0.8 max	0.45		
Mg	0.05 max	0.01	0.1 max	0.05		
Mn	0.3 max	0.01	0.5 max	0.05		
Sn	0.05 max	0.03		0.05		
Ni	0.05 max	0.05	_	0.05		
Zn	0.3 max	0.03	1.0 max	0.05		
Cr	0.05 max	0.03	-	0.03		
Ti	0.2 max	0.03	0.2 max	0.03		
Al	Remainder	Remainder	Remainder	Remainder		

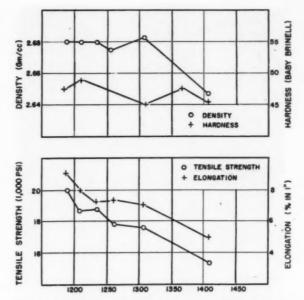
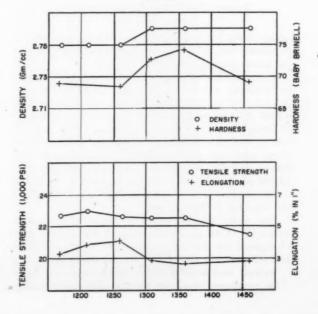


Fig. 3—Effect of pouring temperatures on properties of investment-cast 5Si- Al-Si alloy. Mold temp.—250 F.

creased with increase in pouring temperatures. The data are shown graphically in Fig. 3 and each point represents the average of four to six determinations. For the 250 F molds the highest average values of 19,900 psi tensile strength and 9.1 per cent elongation were recorded for the castings poured at 10 F above the liquidus (1175 F). The lowest values, 15,500 psi tensile and 5.1 per cent elongation were obtained for the castings poured at 450 F above the liquidus (1615 F).

These values compare favorably with those obtained in sand castings although they are approxi-

Fig. 4—Effect of pouring temperatures on properties of investment-cast 4Cu, 5Si Aluminum alloy. Mold temp.—
250 F.



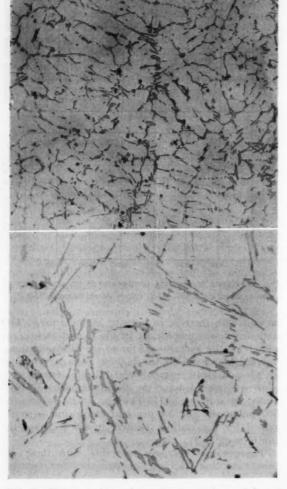


Fig. 5—Microstructure of transverse sections of investment-cast 5Si aluminum alloy tensile specimens. Mag. 50x (top) and 250x (bottom.) Mold temp. 250 F; Pouring temp.—1265 F. Etch—Keller's reagent.

mately 5000 psi lower than the tensile strength of permanent mold castings and approximately 10,000 psi lower than die castings.4

It was possible to produce slightly stronger castings by utilizing a room temperature mold. However, this practice is ill advised because any moisture absorbed by the mold at room temperature will turn to steam during the casting operation. This may result in defective castings. The mold poured under these conditions gave specimens which averaged 20,-200 psi and 6.5 per cent elongation as compared with 18,800 psi and 7.7 per cent elongation for the 250 F mold. Both molds were poured with metal 50 F above the liquidus (1315 F).

The mold which contained chills gave average values of 19,500 psi tensile strength and 8.4 per cent elongation for the 250 F mold and 19,200 psi tensile and 9.3 per cent elongation for the 400 F mold.

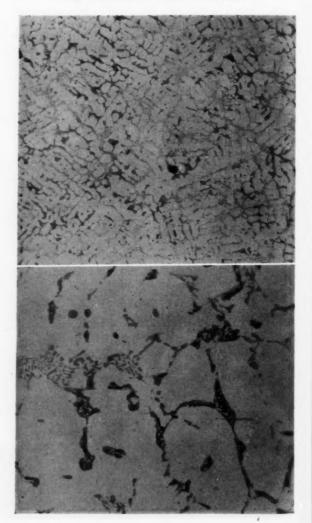
Sodium modification of this composition was tried. This treatment improves the ductility of the alloyconsiderably. A small addition of sodium metal to the melt just before pouring increased the elongation values to 13.0 per cent and also moderately improved the tensile strength. By heat treating the sodium-modified castings at 1040 F for 48 hr, average elongation values of 22.4 per cent were obtained; however, the tensile strength declined to less than 15,000 psi. These data are shown in Table 2.

Table 2—Effect of Sodium Modification on 5% Silicon-Aluminum Alloy

Pour Temp, F	Mold Temp, F	Tensile Strength, psi	Elongation,% (1-in. gage)	Tı	eatment
1250	350	20,100	13.0	Na	Modified
1250	350	14,470	22.4	Na	Modified and
				He	at Treated

b. Al-Si-Cu Alloy. Tests on this alloy revealed that the tensile strength is only moderately affected by

Fig. 6—Microstructure of transverse sections of investment-cast 4Cu, 5Si aluminum alloy tensile specimens. Mag. 50x (top) and 250x (bottom). Mold temp.—250 F; Pouring temp.—1260 F. Etch—Fe(NO₃)₃.



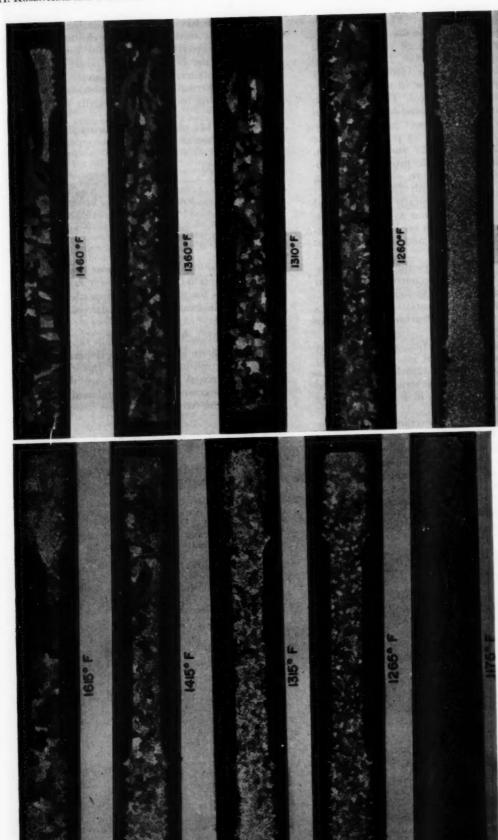


Fig. 7—Macrostructure of longitudinal sections of 5Si aluminum alloy tensile specimens cast into 250 F molds at various pouring temperatures. Etch—NaOH+H₂O.

Fig. 8—Macrostructure of longitudinal sections of 4 Cu, 5 Si aluminum alloy tensile specimens investment-cast into 250 F molds at various pouring temperatures. Etch—NaOH+H₃O.

increase in pouring temperature. The highest average values obtained, 23,000 psi for the 1210 F pouring temperature, exceeded the average values obtained with pouring temperatures of 1160, 1260, 1310, and 1360 F by less than 500 psi. The 1450 F temperature did produce a significant decrease in tensile strength, the specimens averaging 21,500 psi. The highest ductility (4 per cent in 1 in.) was found in the specimens poured at 1260 F, with the lower casting temperatures yielding between 3.2 and 3.8 per cent and the higher temperatures between 2.7 and 2.9 per cent. The best combination of tensile properties was obtained at the 1210 F temperature, elongation being 3.8 per cent and tensile strength 23,000 psi.

These data are shown graphically in Fig. 4 and also represent the average of four to six determinations. Hardness readings, taken on longitudinal sections of the tensile specimens, and density are also

shown in Fig. 4.

Examination of the microstructure of the transverse sections of the tensile specimens revealed that the structures produced by the various casting conditions were similar. The only variation noted was a slight coarsening of the microstructure with increase in the pouring temperature.

Figure 5 shows the microstructure of the 5 per cent silicon specimen which was cast into a 250 F mold from a pouring temperature of 1265 F. These micrographs reveal a coarse eutectic structure consisting of plate-like silicon constituents in a matrix of alpha

solid solution.

Figure 6 shows the microstructure of the 5 per cent Si-4 per cent Cu specimen which was cast from 1260 F melt. This specimen was etched with Fe (NO₃) 3 and the structure is shown at 50 and 250 magnifications. The higher magnification micrograph shows approximately equal areas of silicon and CuAl₂ constituents in a matrix of alpha solid solution. The dark etching constituents is the CuAl₂ and the gray etching constituent is the silicon.

#### 2. Macrostructures

The macrostructures of the tensile specimens cast from various pouring temperatures are shown in Figs. 7 and 8. Coarsening of the grains appears to be a regular function of the pouring temperature. The fine grain size for the lowest pouring temperature is in marked contrast with the extremely coarse structure found above 1450 F.

In view of the similarity which exists between the microstructure of the specimens, the grain size may be responsible for the variations in tensile properties previously discussed.

Some porosity in the castings may be noted although this condition is not thought to be of sufficient severity to affect the properties appreciably.

#### Discussion

Results obtained on tensile properties indicate that solidification rate is the most significant factor which affects properties. This factor also is apparently responsible for the major differences in properties obtained by the various methods of casting (sand, investment, permanent mold and die).

The comparative thermal properties of various

mold materials have been illustrated by Taylor and Wick⁵ in their work showing the chilling power of these mold materials. Their results show the vast difference in the heat extracting power characteristics of sand, plaster, graphite and steel. The plaster has the least thermal conductivity with sand, graphite, and steel having increased conductivity in the order named.

Mondolfo⁶ has placed the following limits on the tensile properties of the aluminum-copper-silicon and aluminum-silicon alloys:

	Al-C	Cu-Si	Al-Si			
	Tensile Strength 1000 psi	Elonga- tion % in 2 in.	Tensile Strength 1000 psi	Elonga- tion % in 2 in.		
Sand Cast	20-25	1-3	17-25	3-8		
Permanent Mold	25-35	1-5	20-28	3-10		
Die Cast	30-35	1-3	25-32	1-5		

Reference to Figs. 3 and 4 indicate that the investment-cast properties fall in the lower range of Mondoldfo's limits for the sand-cast properties. This correlates with the heat conductivity data of Taylor and Wick.

# Acknowledgment

The authors are indebted to the Department of the Army for permission to publish this report. The opinions expressed in the paper are those of the authors and are not necessarily endorsed by the Department of the Army.

## References

1. S. Lipson, H. Markus and H. Rosenthal, "Precision-Cast Copper-Base Alloys," *Iron Age*, Nov. 1945, p. 46-49; Nov. 8, 1945, p. 64-73.

2. H. Rosenthal, S. Lipson and H. Markus, "Precision-Cast Heat-Treatable Copper-Base Alloys," Iron Age, Sept. 19, 1946,

. 48-54

3. H. Rosenthal and S. Lipson, "Precision Casting Aluminum in Moist Investment Molds," Transactions, A.F.S., vol. 58, p. 460 (1950).

4. Aluminum Casting Alloys and Alloys for Other Purposes, Aluminum Co. of America, Pittsburgh, Pa., 1941.

5. H. F. Taylor and W. C. Wick, Insulating Pads and Riser Sleeves for Bronze Castings, Transactions, A.F.S., vol. 54, p. 262 (1946).

6. L. F. Mondolfo, Metallography of Aluminum Alloys, p. 222.

# DISCUSSION

Chairman: HIRAM BROWN, Solar Aircraft Co., Des Moines, Iowa.

Co-Chairman: W. E. Sicha, Aluminum Co. of America, Cleveland.

Recorder: HIRAM BROWN.

HIRAM BROWN (Written Discussion): ¹ The authors are to be congratulated on the excellent data presented in their paper. However, this is not unexpected since most of us are already familiar with the excellent work they have done on the investment casting of copper-base alloys. The authors should also be congratulated upon using the title "Investment Casting" rather than "precision casting" which is so commonly used by others. Most of the comments presented in this discussion are more in the nature of corroboration than disagreement.

In the work which we did in the investment casting of aluminum alloys, we used the direct pour set up, having the test bars fastened directly to the button type sprue as shown in the authors' paper. We found this to be much better than indirect types of pouring for aluminum alloys. Instead of eight bars per sprue, however, we used only four, and we used direct

¹ Company Chief Metallurgist, Solar Aircraft Co., Des Moines, Iowa.

pressure applied from a pressure head rather than suction. Our investment was essentially the same consisting of 70 per cent silica and 30 per cent plaster. These remarks are made to show that our procedures were similar, so it is felt that some comparison of the results is warranted.

The use of atmosphere saturated with water vapor to prevent the formation of capillaries during wax removal seems an excellent idea. The use of an autoclave with superheated steam at 15 psi pressure and 250 F is even better and approximates the practice used by other people who melted out wax patterns by the use of steam impingement on the mold. The presence of the saturated atmosphere raises the question of whether the authors reheated the molds after this treatment to drive off the moisture before placing the molds in the furnace for burn out at 600 F. If not, did they encounter many explosions due to escaping steam during burn out which might, in turn, have affected surface finish. The addition of air or oxygen in the furnace during burn out has also been found helpful in eliminating carbon residue with either wax or plastic patterns. The use of steam to prevent capillaries does not appear necessary when plastic patterns are used due to the fact that viscosity of the plastic prevents its soaking into the capillaries during the pattern melt out operation.

We also found it true that the properties of investment cast aluminum alloys would normally meet the requirements of sand cast specifications for these same alloys. It should be pointed out, however, that this takes into account the fact that the investment cast test bars are 0.252 in. in diameter and 1 in. in gage length as compared with the normal 0.505 in. diameter and 2 in. gage length standard sand cast bars. It is very much to be doubted that ½-in. diameter investment cast test bars would neet the sand cast requirements for that same size of bar. However, due to the fact that most investment castings are thinner and smaller than sand castings, the ¼-in. diameter test bar is usually considered representative of this process.

The authors state: "It is well known that the properties of cast aluminum alloys are quite sensitive to the rate of freezing. This indicates the desirability of using low mold temperatures and minimum pouring temperatures." It is believed that this should be qualified a little since much data indicates that there is a fairly wide range of pouring temperatures and mold temperatures for some alloys that will give comparable results. The authors found that the properties of their 5 per cent silicon alloy dropped rapidly with increased temperature but the 5.5 silicon and 4.5 copper alloy did not change a great deal insofar as tensile strength is concerned over about a 200 F pouring range. Table I attached to this discussion indicates results obtained on three different aluminum alloys in which the pouring temperature was varied in each case 75-100 F. The tensile and

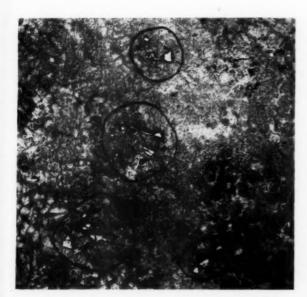


Fig. A—Note beta phase aluminum-magnesium encircled.

TABLE 1—EFFECT OF METAL TEMPERATURE ON TENSILE PROPERTIES

Alloy		Metal Temp.	Tensile,	
ASTM B179-51T	Commercial	° F	psi	% in 1 in.
C4A—As-cast	195-As-cast	1350	13,000	1.0*
		1325	22,500	2.5
		1300	22,800	4.0
		1275	22,700	3.5
		1250	22,100	4.0
SC51A—As-cast	355—As-cast	1350**	-	_
		1300	26,719	1.0
		1275	26,762	1.0
		1250	26,262	1.0
		1225	25,900	1.0
ZG61A—	40E_10 Day	s 1350	35,200	12.0
10 Days Room	Room Tem-	1325	35,200	11.0
Temperature	perature	1300	34,500	11.0
		1275	34,600	12.0
		1250	35,100	11.0

* Coarse, dendritic structure.

** Temperature allowance too low and metal was 1300 F when it reached first mold.

ductility of two of them were affected very little over this range, whereas the third one indicates that its pouring range for optimum results cannot extend more than 50 F. This indicates that some alloys are more sensitive than others in regard to pouring temperature.

Table 2 indicates the effect of mold temperature on tensile properties of the same three alloys shown in Table 1. The mold temperatures used were 400 F and room temperature, respectively. The same two alloys that showed relative insensitivity over the pouring range shown in Table 1 also indicated relative insensitivity to the variation in mold temperature from

TABLE 2—EFFECT OF MOLD TEMPERATURE ON TENSILE PROPERTIES

Alloy ASTM B179-51T	Commercial	Metal Temp.	Mold Temp.	Tensile, psi	Elonga- tion % in 1 in
C4AAs-cast	195—As-cast	1300 1300	400 Room Temp.	19,400 23,800	4.0 3.0
SC51A—As-cast	355—As-cast	1300 1300	400 Room Temp.	29,072 29,226	1.0 1.0
ZG61A—10 Days Room Temp.		1300 1300	400 Room Temp.	33,400 33,000	11.5 10.0



Fig. B-Note almost complete absence of beta phase.

400 F to room temperature, whereas the other alloy again showed that it was affected by this difference in temperature.

One of the alloys which had shown relative insensitivity to pouring temperature and mold temperature in Tables 1 and 2 was selected for further experiments in which the mold temperature was varied from 800 F to room temperature. It was found that no detrimental effect of increased mold temperature was found until the temperature of 800 F was reached. Since these sets of bars were poured from different melts of metal at different times, no direct comparisons can be made but the gvidence would seem to indicate that in the case of this alloy at least, the quite wide range of mold temperature was possible without seriously affecting tensile properties.

In examining the microstructures of the specimens discussed in Table 3, several interesting facts were observed. First of all, grain size did tend to be slightly coarser as mold temperature increased. Secondly, the microstructure changed as mold temperature varied. With the mold temperature of 800 F, there were large areas of what appeared to be beta phase aluminum-magnesium. This is shown in Fig. A. The presence of this phase decreased as mold temperature decreased and was practically non-existent at 400 F as shown in Fig. B.

There is a great deal of work yet to be done on the investment castings of aluminum alloys, particularly in regard to obtaining x-ray sound castings. I am sure that work such as that

TABLE 3-EFFECT OF MOLD TEMPERATURE ON 40E ALLOY

ASTM	Alloy Commercial	Aging	Metal Temp.	Mold Temp.	Tensile psi	Elongation
ZG61A	40E	21 Days R.T.	1300 1300	800 600	33,400 35,400	8.0* 12.0
ZG51A	40E	21 Days R.T.	1300 1300	600 400	33,600 32,900	8.0 6.0
ZG61A	40E	21 Days R.T. R.T.	1300 1300 1300	600 500 400	34,200 33,600 32,700	10.0 9.5 7.5
ZG61A	40E	21 Days R.T.	1300 1300	400 R.T.	33,600 34,000	10.0 11.5

* Bars showed shrinkage porosity. Each set poured from different heat of metal.

being done by Mr. Rosenthal and Mr. Lipson will be of great assistance in this field.

MR. LIPSON: In answer to Mr. Brown, the paper stated that after the wax had been removed by steam in the autoclave, the pressure was reduced and the flasks were allowed to stand overnight at room temperature. This appeared to be sufficient insofar as drying was concerned, and no trouble was encountered from metal reaction with steam after this treatment.

# SUBSTITUTION OF BORON FOR ALLOYS IN CAST STEELS

## By

Ray A. Dyke, Jr., and C. K. Donoho*

#### ABSTRACT

Boron in cast steels substantially increases the hardenability in carbon and low-alloy compositions essentially as it does in wrought steels. With tempering temperatures as high as 1200 F the effect of boron is largely lost. "Fading" of boron has not been found to be a serious problem with normal precautions. The effect of boron on mechanical properties is about as would be expected from the hardenability increase. One useful application is discussed in which a boron steel is substituted for a 4340 type composition with substantial savings in alloys and in cost.

The use of boron in wrought steels to conserve other alloys such as manganese, chromium, nickel and molybdenum has been thoroughly tested and well substantiated in practice. A few thousandths per cent of boron increases hardenability as much as several tenths per cent of the usual hardenability alloys. This use of boron in the steel mills is well justified economically as well as for alloy conservation. In the smaller heats of the steel foundries the savings are less obvious, and relatively little information has been developed on cast boron steels.

#### Hardenability

Portions of several heats of acid electric steel were treated with boron to test the effect on hardenability and other properties. These heats were deoxidized at tap with 2 lb per ton (0.10 per cent) of aluminum. Boron was added in shank ladles as 17.5 per cent ferro-boron along with 0.05 per cent additional aluminum. About the maximum boron effect was found with an addition of ½ lb per ton (0.025 per cent) of ferro-boron, which is 0.0045 per cent boron.

Three types of specimens for the standard endquench hardenability test were tried. Specimens were obtained by (1) machining from normalized and tempered keel block coupons; (2) casting the 1in. diameter specimen in split cast iron molds; and (3) sand casting the specimen in baked oil sand cores. If the cast specimens (2) and (3) are normalized before end-quenching they are found to be as reliable and reproducible as the machined specimen. For convenience the sand cast specimen was used in



Fig. 1-Oil sand mold and cast end-quench specimen.

these tests. A typical mold and cast specimen are shown in Fig. 1.

Figure 2 shows end-quench hardenability curves for five typical steels with and without boron. These were determined on 1-in. round sand cast bars normalized from 1650 F before end-quenching from 1600 F. As in wrought steel experience, the hardenability is definitely increased by the boron in every case. Some quenching tests (Table I) on bars and castings confirm the evidence of the curves.

Wrought steel experience has shown that much of

TABLE 1-STEEL QUENCHING AND HARDENABILITY

		Brinell Hardness					
Steel	Casting	Treatment	No Boron	With Boron			
1025	1-in. coupon	Water quench	223	388			
1045	1-in. coupon	Oil quench	302	627			
8645	4-in. section castings	Oil quench	387	532			

Metallurgical Assistant and Chief Metallurgist, respectively, American Cast Iron Pipe Co., Birmingham, Ala.

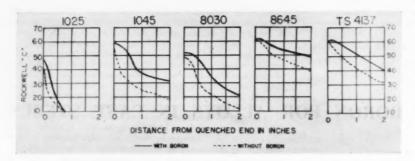


Fig. 2—End-quench curves for typical cast steels.

the boron effect in quenched steel is lost when high tempering temperatures are used. Figure 3 shows hardenability curves of a low alloy cast steel with and without boron, as-quenched and after tempering at the indicated temperatures. At 800 F tempering the boron effect is marked, but at 1200 F the effect is almost completely lost. This is a serious limitation since many cast steel parts normally are tempered at around 1200 F.

# Boron Recovery and Fading

In the heats where granular ferro-boron is added to the shank ladle with some aluminum shot very consistent recovery is obtained. Spectrographic analyses (which show total boron) indicate boron recoveries of over 90 per cent. Boron can be added at the tap if precautions are used to insure that the alloy gets into the steel rather than in the slag. On one 4000-lb heat of 86B45 steel the boron (0.0045 per cent B) was added with the aluminum at the tap. Spectrographic samples at 5, 8 and 18 min after tap showed 0.0042, 0.0037 and 0.0043 per cent boron, respectively, for approximately 90 per cent average recovery. Aluminum content of this heat was at about the 0.12 per cent range.

Another arc furnace heat of 86B45 steel samples at 1, 9 and 20 min gave boron contents of 0.0030, 0.0032 and 0.0031 per cent, for a lower recovery but still no evidence of "fading" (Table 6).

A test heat was run in an acid-lined induction furnace to determine the extent of the "fading" of boron with time. Boron and aluminum in somewhat higher than normal amounts were added to the molten bath in the furnace. With the bath held at 2900 F, successive samples for spectrographic analysis were cast for 30 min. Figure 4 shows that there is no serious loss of boron until the aluminum content had dropped below 0.02 per cent. Apparently the presence of aluminum in the steel tends to protect the boron from oxidation.

## Mechanical Properties

The effects of boron on tensile properties are generally as would be expected from the increased hardenability. Table 2 shows some typical results for several steels as obtained from 1-in. section keel block coupons. Quenched and tempered bars of boron steels tend to be higher in strength and hardness and correspondingly lower in ductility. In the higher hardenability steels (No. 4) there is some strength improvement even with normalized bars. When the boron treated steel is tempered at a higher tempera-

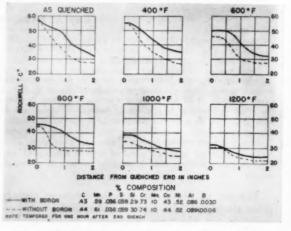


Fig. 3—End-quench curves for a low alloy cast steel.

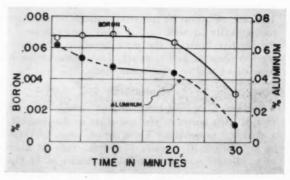


Fig. 4-Fading of boron at 2900 F temperature.

TABLE 2—TENSILE TESTS ON PLAIN AND BORON TREATED CAST STEELS

No.	Steel Type	Heat Treatment	T. S., 1000 psi	Y. S., 1000 psi	El.,	R. A.,	Bhn
(1)	1025	WQ-800T	93.6	71.2	19.6	34.8	201
	1025 + B	WQ-800T	108.7	70.9	15.6	27.9	233
(2)	1330	OQ-1100T	130.0	104.2	15.3	30.0	262
	1330 + B	OQ-1100T	139.2	121.6	12.0	26.2	273
(3)	8030	OQ-1200T	117.8	93.7	18.2	34.5	248
	8030 + B	OQ-1200T	129.5	106.7	14.3	37.4	262
(4)	8645	Normalized	126.2	69.3	14.3	29.3	233
	8645 + B	Normalized	137.5	90.0	8.1	24.2	277
(5)	1045	OQ-800T	145.6	102.0	8.6	8.5	302
	1045 + B	OQ-800T	145.8	93.7	11.9	20.3	293

TABLE 3—IMPACT TESTS ON PLAIN AND BORON TREATED CAST STEELS

	Steel	Heat			ct—ft/lb y V-Notch
No.	Type	Treatment	Bhn	at 70 F	at -40 F
(1)	1025	WQ-800T*	201	37	17
	1025 + B	WQ-800T*	223	17	9
(9)	1330	OQ-1100T*	262	36	19
(2)	1330	06-11001	202	31	18
	1000 · D	00 11007*	079	20	10
	1330 + B	OQ-1100T*	273	23	8
(4)	8645	OQ-1100T*	302	15	10
(4)	0043	OQ-11001	302	19	14
	0048 . 70	00 110075	910	18	10
	8645 + B	OQ-1100T*	316	14	12
(5)	1045	OQ-800T*	302	12	10
, ,	1045 + B	OQ-1000T*	297	12	5
(0)	*****	THO 000TH 6	041	28	
(6)	1025	WQ-800T**	241	27	
	1025 + B	WQ-800T**	341	12	
	1025 + B	WQ-1000T**	241	30	
	1045 + D	11 6-10001	211	31	

* Heat treated as 1-in, coupon bar.

** Heat treated as 10-mm Charpy bar-then notched.

ture to give hardness equivalent to that of the plain steel there may be an improvement in ductility (No. 5 of Table 2).

A limited number of Charpy V-notch impact tests are shown in Table 3. The effect of boron generally is to increase the hardness of the heat treated bars and consequently to lower the impact strength. When the hardness is considered the specific effect of boron on room temperature impact does not appear to be deleterious. As shown in test No. 6 of Table 3, when the boron steel is tempered to an equivalent hardness the room temperature impact strength may be increased. There is some indication that even at a given hardness level the low temperature impact properties of the boron treated steels are lower.

Most of these boron steels contained an excessive amount of aluminum, and it is possible that improved deoxidation techniques can be developed for boron steel that will give improved ductility and impact strength. In this connection it is possible that some of the complex boron addition agents will give better results than the ferro-boron which was used here to limit the number of variables as far as possible. Table

5

4 gives chemical compositions of the heats from which tests were shown in Tables 2 and 3. Numbered lines in Table 4 correspond with those in Tables 2 and 3.

### **Applications**

One definite useful application for boron cast steels is in production use at this foundry. Some equipment castings such as wheels and rollers of 3- to 4-in. section have been cast of 4340 alloy steel for hardening by oil quenching (Fig. 5). It has been found possible to substitute 86B45 steel, which has equivalent hardenability to 4340, at considerable saving in alloys and in cost. A comparison of the alloy content of these two steels is shown in Table 5.

Net dollar saving by substituting the boron steel amounts to over \$18.00 per ton. It is not claimed that the 86B45 steel is in every way equivalent to 4340, but for simply hardening in an oil quench there is no apparent difference. Figure 6 shows hardenability curves of seven of the 86B45 cast steels compared with the extremes of hardenability of SAE 4340 wrought steel composition. The cast boron steels



Fig. 5-Equipment castings made in 4340 or 86B45 steel.

TABLE 5-ALLOY CONTENT COMPARISON

		Alloy Conte	ent, Per Ca	ent
Steel	Ni	Cr	Мо	В
4340	1.75	0.75	0.25	_
86B45	0.55	0.65	0.11	0.005
Saving, %	1.20	0.10	0.14	
\$ per ton saving	14.40	0.60	3.70	0.60

TABLE 4-CHEMICAL COMPOSITION OF PLAIN AND BORON TREATED STEELS MECHANICALLY TESTED

	Steel					Compo	sition, Per	Cent				
No.	Type	C	Mn	P	S	Si	Cr	Мо	Ni	Cu	Al	В
(1)	1025	0.25	0.48	0.037	0.039	0.55	0.044			0.14	0.057	
	1025 + B	0.25	0.48	0.037	0.039	0.55	0.044			0.14	0.096	0.0038
(2)	1330 (Mod.)	0.30	1.40	0.039	0.029	0.69	0.26		0.27	0.13	0.089	< 0.0006
	1330 + B	0.28	1.39	0.039	0.029	0.67	0.28		0.25	0.12	0.101	0.0040
(3)	8030	0.30	0.80	0.029	0.045	0.57	0.25	0.140	0.42	0.11	0.090	
, ,	8030 + B	0.30	0.82	0.026	0.040	0.58	0.25	0.140	0.42	0.11	0.090	0.0018
(4)	8645	0.43	0.81	0.025	0.034	0.66	0.67	0.097	0.72	0.12	0.074	< 0.0006
	8645 + B	0.43	0.84	0.025	0.045	0.70	0.61	0.102	0.72	0.12	0.116	0.0048
(5)	1045	0.46	0.74	0.020	0.042	0.57	0.04			0.11	0.110	
. ,	1045 + B	0.46	0.74	0.020	0.042	0.57	0.04			0.11	0.160	0.0035
(6)	1025	0.27	0.37	0.031	0.026	0.42	0.05			0.10	0.028	< 0.0005
. ,	1025 + B	0.27	0.37	0.031	0.026	0.42	0.05			0.10	0.052	0.0045

TABLE 6-CHEMICAL COMPOSITION OF 86B45 CAST STEELS

					Compo	sition, Per	Cent					Time
No.	C	Mn	P	S	Si	Cr	Мо	Ni	Cu	Al	В	min
57846	0.446	1.02	0.032	0.031	0.72	0.70	0.148	0.65	0.14	0.122	0.0042 0.0037 0.0043	5 8 18
38701	0.436 0.441	0.84	0.039	0.038	0.51	0.73	0.12	0.61	0.12	0.100	0.0030 0.0032 0.0031	9 20
K202	0.429 0.424	0.67	0.037	0.042	0.75	0.91	0.078 0.077	0.71	0.134	< 0.015	0.0046	
38747	0.450 0.445	0.87	0.87	0.041	0.57	0.74	0.23 0.22	0.68 0.73	0.13	0.031	0.0036	
38748	0.422 0.422	0.92	0.92	0.028	0.58	0.78	0.26 0.26	0.76 0.73	0.13	0.050	0.0035	
38752	0.449 0.444	0.95	0.020	0.034	0.52	0.68	0.12	0.66	0.12	0.054	0.0024	
38644	0.430	0.84	0.025	0.045	0.70	0.61	0.102	0.72	0.13	0.116	0.0048	
AISI • Ela	0.43/ 0.48 psed time af	0.75/ 1.00 ter FeB ad	ldition.			0.55/ 0.75	0.08/ 0.15	0.40/ 0.70				

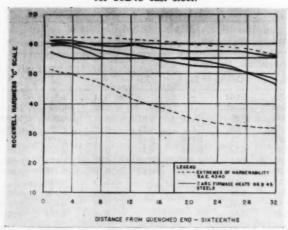
are consistently to the high side of the hardenability range. Table 6 shows chemical composition of these heats.

While these 86B45 cast parts harden by oil quenching just as well as did the previously used 4340 composition, there is an advantage in the boron steel in the earlier heat treatment stage of softening for machining. While the 4340 composition is difficult to soften, requiring an extended holding time of about 1200 F, the 86B45 parts soften much more readily in a routine normalize and temper heat treatment and have given rise to few complaints for machineability.

#### Conclusions

- Boron is an effective alloy for the steel foundry for achieving required hardenability with minimum alloy content and at reduced cost.
  - 2. Fading of boron during normal casting opera-

Fig. 6—Graph showing end-quench hardenability curves for 86B45 cast steel.



tions is not a serious problem if ordinary precautions are used.

- 3. Some study of deoxidation practice is required to determine the best practice to give maximum ductility and impact strength.
- 4. Replacement of a 4340 type composition with a 86B45 steel for castings has been found to be convenient and economical.

# DISCUSSION

Chairman: C. B. JENNI, General Steel Castings Corp., Eddy-

Co-Chairman: Dale Hall, Oklahoma Steel Castings Co., Tulsa, Okla.

Recorder: L. H. HAHN, Sivyer Steel Castings Co., Chicago.

E. A. LORIA (Written Discussion): ¹ This is one of the first papers on the use of boron in cast steels. The effects of boron on hardenability are unique in several respects. These can be enumerated: (1) An extraordinarily small amount has a powerful effect on hardenability. (2) Its effect on hardenability varies considerably with the carbon content of the steel. (3) The nitrogen content, the choice of deoxidizers, and the negating effect of high temperature heat treatment all have a definite influence on the effectiveness of boron. Considerable information on heat-treating grades of automotive wrought steels has been acquired and it appears that only in ineffectively treated steels have the real effects of boron been somewhat obscured. The comments on this paper will follow closely the sequence of topics therein.

The cast steels under consideration were deoxidized with 3 lb per ton of aluminum. The relationship between the amount of aluminum used and the desired boron hardenability effect is an important one. In the case of wrought steel grades from 0.20 to 0.70 pct carbon, an improved boron hardenability effect is obtained with higher amounts of aluminum. The linear trend for the boron multiplying factor for 50 pct martensite vs lb aluminum addition is shown in Fig. A.

Recent work indicates that the use of two or more addition agents to the steel ladle instead of a single ferroboron alloy has considerable merit and the authors do acknowledge that some of the complex addition agents might give better results. It would be worthwhile to know what was the time elapsing be-

¹ Senior Engineer, Metallurgy, The Carborundum Co., Niagara Falls, N. Y.

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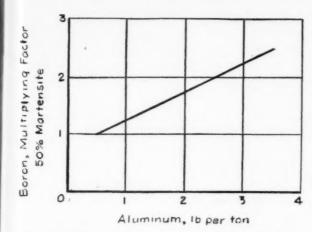


Fig. A—Relationship between boron hardenability factor and aluminum deoxidation for wrought steels of 0.20-0.70 pct carbon.

tween the aluminum-boron additions and the pouring of the hardenability test bars and particularly the equipment castings made from the 86B45 steel. Also what was the pouring temperature in each case?

The increased Brinell hardness values for the boron steels in Table 1 are noteworthy, particularly for the 4 in. section castings. In regard to the latter do the two values represent surface hardness or center hardness? Could the center hardness for a 6 to 8-in. section of 86B45 vs 8645 steel be given?

The tempering experiments are interesting. The loss of the boron effect on tempering at 1200 F may be a serious limitation of boron steels for castings. The boron constituent in steel consists of a row of fine dots and it is absent in boron steels which have lost their boron effect through a high temperature treatment. Is the boron constituent absent in the 1200 F tempered steel? Would it be possible to get a better hardenability effect for the 1200 F tempered steel by raising its chromium content and perhaps lowering its carbon content, since the effect of boron decreases with carbon content?

The writer would like to have seen both the boron analysis and hardenability fading results in arc furnace steel heats where the fixation of nitrogen because of the arc was accentuated and have the results not only for 1-in. section bars but also for heavier, simultaneously cast sections. Since boron has an affinity for nitrogen like aluminum, what is the tendency for boron-teated steel to produce rock-candy fracture in cases where there is nitrogen absorption during melting and no titanium addition is made? Melting with a vigorous boil and a short time in the furnace after melt-down are conducive to low nitrogen absorption. No nitrogen values are given in the paper and no mention is made of the necessity to make nitrogen inactive or insoluble with the addition of titanium.

In any of their improperly or ineffectively treated steels, did the authors observe any streaks of very fine inclusions which are associated with rock-candy fractures? The streaks, observed with 0.0025 pct boron, are hard to find in an unetched polished section but are attacked by almost any etching reagent. They are faintly visible in the unetched condition whereas the boron constituent associated with hardenability is never visible unless etched. These particles will not dissolve on austenitizing up to 2000 F or be affected in quantity by heat treatment which is again contrary to the observations with respect to the boron constituent. They are believed to be oxidized boron formed during or shortly after the final stages of solidification. Perhaps the nitride may be actually formed as the streaks do resemble aluminum nitride which is associated with rock-candy or concluded fracture.

It should be noted that the mechanical properties were obtained from 1 in. section coupons. Have the authors observed, in any of their work on improperly treated boron cast steels, a coarse dendritic structure which may produce an angular type of fracture with lower ductility? This has been noted on occa-

sion with 0.0030-0.0035 pct boron and 0.01 pct aluminum. Hot pouring temperature may be the contributing factor. Cooling velocity should be controlled as it has been shown that cooling velocity, after casting, may be an important factor and the greater the velocity the less is the embrittling tendency. For example, in test bars cast from the same heat of steel, those cut from heavy sections, say 6 to 8 in., will show conchoidal fractures while the standard 1 in. section bars will not show the embrittlement.

In regard to the impact test results, did the authors study possible grain size variations? Experience in wrought steels indicates that 0.5 lb more aluminum than used without boron is necessary to get grain size control. Can the authors speculate on the cause for their conclusion that even at a given hardness level the low temperature impact properties of the boron treated steels are lower? Also what type of fracture is observed in such instances and would they comment on the inclusion structures? Perhaps some of the phenomena that have been described may be occurring.

The statement is made that the boron steels studied contained an excessive amount of aluminum which probably means that type III sulphides were obtained in the test bars, and consequently ductility and toughness should have been adequate. Usually with intermediate amounts of aluminum, these properties are lowered because the inclusions are then in the form of thin envelopes partially outlining the primary grains. Also would not nearly the amount of aluminum used be necessary to prevent pinhole porosity when producing green sand castings? In regard to improved deoxidation technique, the element of time in respect to final deoxidation may be of importance. In commercial heats, records should be kept on type of additions and the time of additions in relation to the tapping time.

Finally the authors' admission that the 86B45 steel is not in every way equivalent to 4340 steel should be accentuated. They have developed their own useful application. Nickel can be eliminated or at least used much more economically in many case-hardened and heat-treated steels. For heavier sections than their 3 to 4 in. equipment castings, its deep hardening effect cannot be attained with other elements such as manganese. In how heavy an alloy steel section have the authors' been able to get the required hardenability with the partial substitution of boron for the other alloying elements?

# DISCUSSION BY W. W. AUSTIN

WM. W. Austin, Jr. (Written Discussion): ^a The authors of this paper have done a commendable job in presenting in concise, useable form, several important conclusions concerning the use of boron in cast steels. Their findings on the relation between aluminum content and fading effect, and their confirmation of adverse effects of high tempering temperatures, are particularly noteworthy.

The use of consecutive ladle additions of aluminum to protect the boron from oxidation or fading while pouring in the foundry is a simple and expedient method of insuring effectiveness of boron treatment. It is clear from the data shown that no serious loss of boron occurs as long as the aluminum content of the metal is maintained at 0.04 to 0.05 per cent.

As mentioned by the authors, further work on deoxidizers for use in conjunction with boron treatment should reveal even more effective means for protecting boron and improving strength and ductility. It is believed that deoxidizers like cerium or calcium-manganese-silicon would prove definitely beneficial in this respect.

It is highly probable that any beneficial effect resulting from cerium or calcium-manganese-silicon deoxidation would also be reflected in improved impact properties. As pointed out by the authors, most of their boron treated steels contained comparatively large quantities of aluminum, which is not conducive to maximum impact properties. Consequently, additional work on other deoxidizers should prove of value with respect to improved impact properties as well as in protecting the boron addition from fading effects. Work of this nature is presently being conducted in the writer's laboratory but results are not yet available for publication.

The influence of tempering temperature upon mechanical

² Senior Metallurgist, Southern Research Institute, Birmingham, Ala.

properties of boron treated wrought steels has been studied with considerable interest. The authors finding that boron treated cast steels suffer much greater loss of strength at a given tempering temperature than comparable alloy cast steels, serves to emphasize this important limitation of boron steels in general. With adequate knowledge of tempering behavior, however, boron cast steels should find ample usefulness in many applications in spite of this characteristic.

The substantial savings in alloy costs listed by the authors through substitution of boron for strategic and expensive alloys is most impressive. This factor combined with the advantage of relatively greater ease of softening for machining shown by boron steels over conventional alloy steels, presents an extremely convincing case in favor of the growing use of boron treatment

for cast steels.

С. К. Donoнo (Authors' Closure): Mr. Loria's discussion is very much appreciated and he brings up most of the "things we don't know about boron cast steels." Our study was intended to be a preliminary survey only, and, as such, is far from complete.

We used 3 lb per ton of aluminum to reduce the chance of boron "fading" and in this we were apparently successful. In Table 6 data is given for the first two heats on elapsed time after boron addition where equipment castings were made from 86B45 steel. The pouring temperatures averaged around 2850 F.

The hardnesses in Table 1 for the 4-in. section castings were taken near the surface in each case. However, a fairly accurate estimate of the hardness to be expected in any portion of a heavy section can be made from the hardenability curves in Fig. 6 using the well established cooling rate data relating Jominy position to equivalent section size in oil or water quench.

It has been rather well established that nitrogen in steel adversely affects the hardenability-increasing effect of boron. It is probably just fortuitous that we did not encounter trouble from this source with our raw materials and our melting practice. We have since encountered certain melting stock which did not give the expected hardenability increase with ferroboron additions, but responded normally to titanium-bearing boron addition agents.

The fact that the low temperature impact properties of the boron treated steels are not high may be due to a number of causes such as inclusion type, hydrogen content, freezing rate,

etc. This phase particularly needs careful study.

Dr. Austin's discussion is welcome and to the point. We are glad that Southern Research Institute is conducting a fundamental study of deoxidation techniques and will be most interested in the results obtained.

We agree with Mr. Strauss that inclusion distribution is more significant in cast than in wrought steels. Elimination of aluminum will normally improve the inclusion type and give increased ductility but we believe we would have to have some other strong deoxidizer as protection against boron fading.

Mr. Tisdale's comment as to improved machinability of boron steels is well taken. With the same quenching hardenability, the boron steels, as normalized, will be easier to machine

Mr. Berry's experience with low hardenability in 10 per cent of his heats, which can be cured by using titanium, is very interesting. Until control of nitrogen content is worked out better the use of titanium may be necessary for consistent results. The use of calcium alloys for improving ductility and impact strength is attractive and promising, as is also the use of rare earth metals.

# FLOWABILITY OF FOUNDRY SAND

By

Loris M. Diran,* Amos J. Shaler** and Howard F. Taylor**

#### ABSTRACT

Techniques developed to date for measuring the flow characteristics of molding sands have not taken into consideration the flow mechanism involved; no true index of flowability exists today whereby foundry sands can be dependably evaluated or compared.

Strain studies have been conducted in the past on metal powders. Deformable metal grids were used to determine the mechanisms by which these powders flowed when compressed in a die. A similar approach seemed indicated in the case of molding sands, but a new technique had to be developed, using grids formed by introducing lines of graphite into the sand specimens. The results of the strain analyses of cylindrical sand specimens were combined with compression-density data to study the flow-mechanism of molding sands. In order to rationalize the fiindings of flowability, compressibility, and moldability. These follow:

(1) "Flowability" is the deformation or flow molding sand undergoes on applying an external force. The flow is due almost entirely to the deformation of the clay-water film surrounding the sand grains.

(2) "Compressibility" is the ratio of densities of a molding sand before and after packing.

# I. Introduction

The basic material used in the foundry is silica sand. It is the cheapest high-temperature refractory known and extensive high-grade deposits are available for foundry use. Pure silica sand has no bonding ability; it cannot be worked into dense, self-supporting shapes. For this reason, binders are added to the sand to provide the necessary bond. Plastic clays,† when properly hydrated, have proved to be excellent binders.¹ Other materials, such as cereals, are often added in small amounts to improve the strength properties of molding sands.

A molding sand must not only possess sufficient refractoriness and bond strength; it must also have the ability to flow into and around patterns or dies, to form a dense mass at the sand-pattern interface. Thorough knowledge of the flow characteristics of a molding sand is therefore important to the foundry(3) "Moldability" is a measure of the ability of molding sand to flow and pack into a dense mass under the influence of an externally applied force.

Grid analysis studies showed foundry sand flow consists of two different mechanisms. Flow first occurs by deformation of the clay-water films that coat the sand grains. The second stage of deformation is due to fragmentation of sand grains, but is not important in molding practice.

The deformation process is materially affected by the type of hydrated films about the sand grains, by the size distribution of the sand grains and by the method of loading.

Compressibility measurements showed that the densification of foundry sands depends upon the deformation process. Molding sands of good flowability have low compressibility.

Compressibility and "shear-angle" tests, based on the grid analyses, should prove to be practical test methods. Results have been obtained that are in excellent agreement with grid analysis results.

A combination of sowability and compressibility indices may prove to be a satisfactory measure of the moldability of foundry sands.

man. Molding sands are known to vary in the relative ease with which individual sand grains move into position about the face of a pattern. Good flow characteristics are essential for dense packing at the sand-pattern interface as well as for easier and therefore less costly ramming. Low packing densities, generally erroneously associated with molding sands having poor flow characteristics, usually cause poor surface finish or severe metal penetration when the casting is poured.

#### II. Review of Previous Work

Several testing techniques have been developed for the purpose of evaluating the strength, compacting properties and flowability of foundry sands. It is characteristic of most of them that no attempt has ever been made to separate these properties and measure each independently of the others; therefore, the molder is often unable to make an intelligent sand selection from the test results.

Dietert and Valtier² have attempted to evaluate the flow characteristics of molding sands empirically. In their testing procedure, a known weight of sand is placed in a smooth-walled steel die³ and the de-

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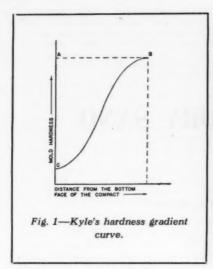
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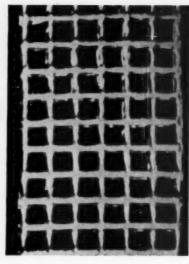
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Research Associate and
 Associate Professor, Department of Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

⁺ Sodium bentonite, calcium bentonite and plastic fireclays (illite- and kaolinite-bearing clays).





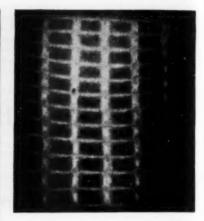


Fig. 3-Photo of a distorted lead grid.

Fig. 2—Photo of an undistorted lead grid.

formation occurring between the fourth and fifth blows of a 14-lb weight is measured by means of a dial gage. The percent "flowability" is given by the following equations:

Percent Flowability =  $1000 (0.10 - \Delta h)$ , where  $\Delta h$  is the deformation in thousandths of an inch occurring between the fourth and fifth blows of the rammer.

Aside from obtaining more nearly uniform results, there is no apparent reason for using the deformation occurring between the fourth and fifth rams to compute the flowability index. Dietert and Valtier have neglected to consider the possibility that a number of molding sands may attain their maximum densities prior to the fourth ram. These molding sands all have a flowability index of 100 and it is impossible to determine their relative flow characteristics.

Mold hardness depends, to a great extent, upon the ease with which a molding sand flows together under an applied force. The more readily the claycoated sand grains flow together into a densely packed mass, the greater is the mold density and consequently, the higher is the mold hardness. Based upon this premise, Kyle⁴ has used mold hardness surveys to evaluate flowability of different molding media.

Kyle has used two methods to express flowability of molding sands. In the hardness differential method, a known weight of molding sand is placed in a smooth walled steel die and a 14-lb weight is dropped through a one-quarter inch distance once. The compact is stripped from the die and hardness readings are taken at the top and bottom of the test specimen. The ratio of these hardness values serves as the flowability index. If the hardness ratio is equal to one, the molding sand is considered to possess ideal flow characteristics.

In the hardness gradient method, a steel die, 1 in. by 12 in. by 123/4 in., is filled with the test mixture. The sand is compressed into a compact. The side plate of the die is removed and mold hardness measurements are taken along the 123/4-in. height of the

test specimen. The mold hardness values are plotted as a function of the distance from the base of the test specimen. A typical curve is shown in Fig. I. In terms of flowability, an ideal molding sand has been assumed to be one that is uniform in hardness throughout the specimen. The curve AB represents the hardness values of such a hypothetical molding sand whereas the curve BC represents the shape of the mold hardness curve that has been obtained for typical molding media. The ratio of the areas under the curves, BC and AB, is considered to be a measure of flowability. Should the curves, AB and BC, coincide, the ratio of the areas is equal to one and the molding sand is assumed to have ideal flow characteristics.

The flowability index obtained by the hardness differential method has little selectivity. It is possible to obtain the same flowability index for a number of different molding sands even though there is a pronounced difference in their respective mold hardness values. This possibility that Kyle has neglected to explain is clearly indicated in Table 1. It does not appear reasonable the above-mentioned molding sands with such great variations in mold hardness should all possess identical flow properties.

Similarly, the flowability index obtained by Kyle's second method, the hardness gradient method, is not a true evaluation of flowability. As stated above, the flowability index is numerically equal to the ratio of areas under the actual and ideal mold hardness gradient curves. There is a possibility the ratio of

TABLE 1—FLOWABILITY INDICES
Computed by the Kyle Hardness Differential Method

Mixture	Hardne	ess Values	Flowability
No.	Тор	Bottom	Index
1	80	60	75
2	60	45	75
3	50	37.5	75
4	40	30	75

the areas under the actual and hypothetical hardness gradient curves for a number of molding sands may be identical even though these curves are appreciably different in shape from one another. These sands would then all possess identical flow characteristics. Nevertheless, it does not seem plausible molding sands with appreciably different mold hardness patterns should have identical flow properties.

Foundry molding sands and metal powders are alike in some respects; each consists of discrete, incompressible particles that can be packed into strong, dense compacts. Kamm, Steinberg and Wulff⁵ and Kamm⁶ have studied the behavior of metal powders at high pressures. They have used a deformable lead grid to determine the mechanism of metalpowder flow as well as the density pattern of pressed metal compacts and die-wall friction effects in metal compacts. In their technique, a thin lead grid, Fig. 2, is accurately centered along the vertical axis of a cylindrical die; metal powder is carefully poured into the die cavity and the powder is then pressed into a dense compact. Radiographs of the pressed compact, Fig. 3, clearly indicate the deformation of the lead grid. Strain analysis data are obtained from photographic enlargements of the deformed lead-grid radiographs. Strain analyses alone do not, however, satisfactorily explain the flow of molding sands because the mechanism is complex and different flow phenomena are dominant under different conditions of triaxiality of the stress pattern.

#### III. Statement of the Problem

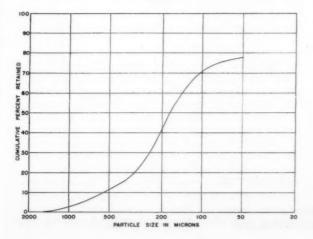
A review of published work does not disclose how clay-water film coated sand grains flow into a densely packed mass; neither does it yield a dependable tool for evaluating the flow characteristics of different molding media. The present research is designed to:

- Determine the mechanism of molding-sand flow, and
- Develop a simple flowability test for foundry molding sands.

#### IV. Experimental Procedure

# A. Preparation of Molding Sands-Three grades of

Fig. 4—Grain size distribution of 80-1 natural molding sand.



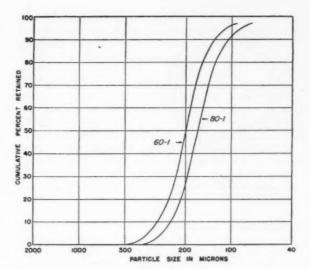


Fig. 5—Grain size distribution for 60-1 and 80-1 grades of silica sand.

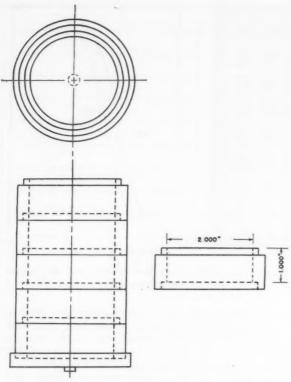
molding sands have been used to determine the flow mechanism of foundry sands in cylindrical dies. One is a natural sand whose particle size distribution and clay content are reported in Fig. 4. Seven per cent water is added to develop the desired bond strength. The other two are synthetically-bonded sands; each contains 4 per cent sodium bentonite and 4 per cent water. The essential difference between the two synthetically-bonded sands is the particle size distribution of the sand phase. These data are shown graphically on Fig. 5.

The natural sand is mulled dry for 5 min in a sand muller to break up sand lumps; the required amount of water is added and the mixture is mulled wet for an additional 5 min. To prepare synthetic molding sands, clean and dry sand and bentonite are mixed dry for 1 min; the water necessary for bonding is mulled-in wet for 5 min. All sand mixtures are tempered in air-tight containers for 1 hr before use.

B. Preparation of Grid Specimen—Initially, the lead grid technique^{5,6} was used to study the deformation of molding sands in cylindrical dies. Examination of the sand compacts showed that the enclosed lead grids had been displaced from their original positions and were also undesirably distorted. These grids could, therefore, not be used to obtain strain distribution data and it was necessary to modify the original grid technique.

The original one-piece die was replaced by a multisectional steel die, a sketch of which is shown in Fig. 6. The first section of the five-piece die was filled by screening molding sand through a six-mesh sieve into the die cavity. The excess sand was carefully removed with a straight-edge so that the sand grains loosely packed in the steel die would not be disturbed appreciably.

A brass template in which  $\frac{1}{32}$ -in. parallel slits had been milled one-quarter inch apart was placed on the top surface of the bottom die section. Powdered graphite was sprinkled on the template; the carbon passing through the slits made parallel markings on



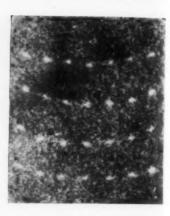
the top face of the molding sand in the die section. The same series of operations were repeated for each section until all five sections of the die were filled. The loosely packed molding sand in the assembled

Two methods were used to compress such grid-containing compacts in order to determine what effect the mode of packing had upon the flow mechanism. In the first method, a 14-lb weight was dropped upon

die was then rammed or pressed into dense compacts.

Fig. 6(left)—Sketch of die for making carbon - grind, sand compacts.

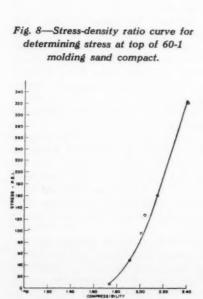
Fig. 7(right)—Photo of sand specimen showing deformed carbon-grid, sand



the loosely packed sand in the die. Some grid specimens were rammed once; others were rammed four times. In the second method, a pressure of 225 psi was applied to produce the compacted grid-bearing specimens.

After compression, sand compacts were stripped from the die and dried in air overnight. The specimens were carefully sawed along a longitudinal plane containing a diameter normal to the original carbon markings. The cut surface clearly showed a series of carbon dots which were easily visible against the white sand background. A photograph of such a section is shown in Fig. 7. The dots formed the corners of a deformed grid of which the dimensions of the original rectangular meshes were readily deduced from the die and template measurements.

Reflex-copy prints were made of the sliced sections and enlarged to a magnification of two in order to make for accurate grid measurements. Four grid specimens were used for each enlargement to obtain representative sites of the grid points. The measurements required for the grid analysis of the molding sands studied are included in Tables 2 to 5.



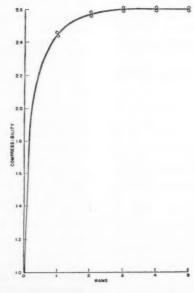
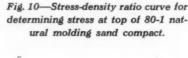


Fig. 9—Compressibility-stress relation for 80-1 synthetic molding sand.



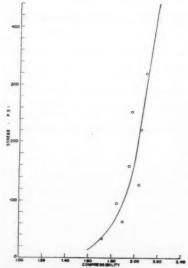


TABLE 2-GRID MEASUREMENTS 80-1 Synthetic Sand; 1 Ram

Point	ra	$\mathbf{b_i}$	d
2	0.500	0.238	75.40
3	0.432	0.246	71.40
4	0.493	0.252	74.50
7	0.515	0.269	80.00
8	0.510	0.267	77.50
9	0.458	0.260	78.90
12	0.395	0.228	86.50
13	0.545	0.249	85.75
14	0.518	0.267	85.60

TABLE 3-GRID MEASUREMENTS 80-1 Synthetic Sand; 4 Rams

Point	ra	$\mathbf{b_{i}}$	d
2	0.435	0.260	73.70
3	0.440	0.255	74.60
4	0.425	0.235	82.30
7	0.475	0.265	82.00
8	0.455	0.280	85.60
9	0.400	0.290	87.50
12	0.500	0.230	86.00
13	0.435	0.245	90.00
14	0.390	0.255	89.20

C. Preparation of Compression-Density Compacts-A strain analysis alone does not furnish all the data necessary to determine the flow mechanism of molding sands. Compression-density data are also required to convert strain distributions to stress distribution by Siebel's7 method. Before presenting the results calculated from grid data, we must describe how to obtain data on compaction in a die of large diameterto-height ratio. For this purpose a circular die, 2 in. in diameter and 1 in. high, was filled by passing molding sand through a six-mesh screen which was positioned approximately 12 in. above the top face of the die. The excess sand was carefully removed with a straight-edge. Just as in the preparation of the deformed grid compacts, the wafers were made by ramming and by pressing. The height of the resulting compacts was measured in the die by means of a micrometer. The final-to-initial density ratio is also the ratio of the initial height to the final height. The results are also expressed as true strain values (natural logarithms of the density ratio). These data are given in Fig. 8 to 13.

D. Grid and Compression-Density Tests as a means of Obtaining Flowability Indices—Grid analyses for strain distribution can be converted to a stress distribution by using Siebel's method⁷ in conjunction with compression-density test results as detailed in Appendix A. From the stress and strain distributions, it is possible to compute for each compact a shear stress-shear strain curve at various values of density. It was found the curves were not appreciably affected by density changes provided a considerable densification had been realized. The curves then consisted of two essentially linear portions—one showing appreciable initial shear strain at very low stress and the other, a rapid rise in stress with little further shear

strain. The intersection of these two lines, or more simply, the intercept of the second line with the strain axis gave an index of flowability of the sand. These values in terms of shear strain are given in Figs. 14 to 17.

This index of flowability involves a laborious calculation; while it sheds light on the flow mechanism of these sands, it is of little practical use in the foundry laboratory. Other test methods have been developed; they are simple in operation and yet they preserve the virtue of correlating closely with gridanalysis results. One test proved to be the compression-density test itself. Accordingly, a number of other sands were tested in the manner described in Section C. The mixtures were so selected as to bring to light the effects of various types of clay binder, of the size distribution of the sand particles and of the moisture content. For close control, it is necessary the results from such a test indicate slight variations in the above-mentioned variables.

Sodium and calcium bentonites and a common fireclay were used to determine the effect of binders upon the density ratio. Only commercially feasible sand mixes were used. The compositions tested are given in Table 6.

Four grades of silica sand were used to study the effect of particle size distribution upon the density ratio. Particle size distribution data are included in Table 7.

It was necessary to have the same bond intensity in all four molding sands for a true evaluation of the grain size distribution effect. In this series of experiments, A.F.S. 60 sand was assumed to be "ideal." Five per cent sodium bentonite and 3 per cent water were added to develop the desired bond in the sand. It was assumed the hydrated clay particles coated the sand grains completely. Sodium bentonite and water

TABLE 4—GRID MEASUREMENTS 60-1 Synthetic Sand; Applied Pressure—225 psi

Point	ra	$\mathbf{b}_{i}$	d
2	0.275	0.453	72.00
3	0.283	0.477	73.50
4	0.273	0.473	81.50
7	0.275	0.460	86.00
8	0.273	0.480	85.00
9	0.270	0.433	87.25
12	0.310	0.500	92.00
13	0.320	0.483	92.00
14	0.333	0.410	86.00

TABLE 5—GRID MEASUREMENTS 80-1 Natural Molding Sand; Applied Pressure-225 psi

Point	ra	$\mathbf{b_i}$	d
2	0.250	0.428	82.50
3	0.238	0.470	80.50
4	0.218	0.470	77.50
7	0.248	0.463	84.75
8	0.268	0.425	83.00
9 0.283	0.488	88.75	
12	0.225	0.483	84.50
13	0.225	0.483	91.00
14	0.225	0.480	86.75

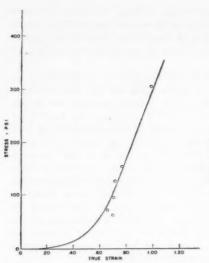


Fig. 11—Stress-strain curve for 60-1 synthetic molding sand.

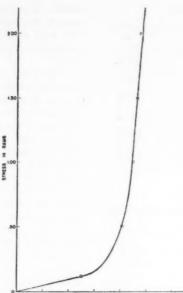


Fig. 12—Stress-strain curve for 80-1 synthetic molding sand.

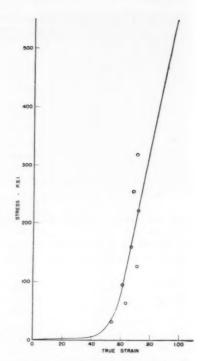


Fig. 13—Stress-strain diagram for 80-1 natural molding sand.

additions were made on a surface area basis so the bond coating in each sand mix was the same. The sand compositions are reported in Table 8.

E. The "Shear-Angle" Test for Flowability — Another method, more applicable to a foundry control laboratory was used for indexing the flow characteristics of molding sands. This test is based upon the flow-mechanism results of the grid analyses. The shear apparatus is shown in Fig. 18.

The shear compacts are made in the two-piece die. The top section, which is open at top and bottom, fits over the bottom section of the die. The bottom section consists of two detachable side plates; the two ends are hinged to the base plate so the sand within the die can be deformed. The molding sand compositions tested are given in Tables 6 and 8.

The following test procedure provides consistent results. The die is filled by passing tempered molding sand through a six-mesh screen. A plate is inserted into the top die opening and the molding sand is pressed into a compact. The packing pressure used here which give a resproducible reading is 7 psi. The die is taken from the press and the top half is carefully removed from the lower section. The portion of the sand compact extending above the top surface of the bottom section of the die is cut away with a thin saw-blade. A steel plate is then placed on the top surface of the compact and the die is again centered in the hydraulic press as shown in

Fig. 14—Stress-strain curve in shear for 80-1 synthetic molding sand, one ram.

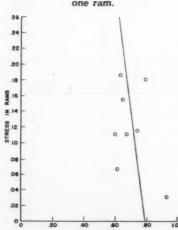


Fig. 15—Stress-strain curve in shear for 80-1 synthetic molding sand,

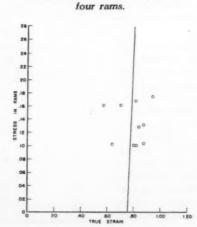
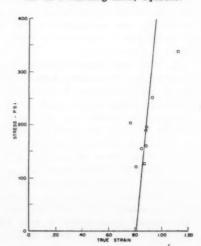


Fig. 16—Stress-strain curve in shear for 60-1 molding sand; squeeze.



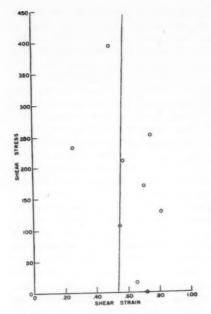


Fig. 17 (left) — Stress-strain curve in shear for 80-1 natural molding sand; squeeze.

Fig. 18 (right) — Sketch of "shearangle" die.

Fig. 19. The "shear" compact is again compressed to 7 psi. A shearing load of 25 lb is then applied as shown in Fig. 20. It is observed the compact is rapidly deformed in shear up to a critical amount after which the stress required for further strain rises abruptly. As soon as the shearing operation is completed, the applied loads are released. The die is removed from the press and one side of the die is detached. The angle through which the hinged ends move is measured with a protractor. This angle is the flowability index and corresponds to the shear strain obtainable with little stress, i.e. the intersection of the two linear portions of the shear stressshear strain curve. The magnitude of the angle of shear obtained in the two methods differs appreciably because in the shear test, conditions of simple shear are not met. The validity of the test is based upon the correlation between the two values. In Fig. 21, the "shear angle" is plotted as a function of the moisture content for several clay binders. The "shear angle" is plotted as a function of the particle size distribution of several sodium bentonite bonded sands in Fig. 22.

It may be seen in these figures and tables the test is sensitive to changes in molding sand characteristics. The property that is measured bears a direct rela-



Composition 1

5 lb 60-1 sand

19 grams sodium bentonite

The moisture ranges from 3.0 to 6.0%

Composition 2

5 lb 60-1 sand

119 grams calcium bentonite

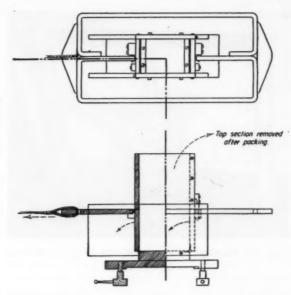
The moisture ranges from 3.0 to 8.0%

Composition 3

5 lb 60-1 sand

401 grams Pedro fireclay

The moisture ranges from 4.0 to 8.0%



tion to the flowability as defined from the results of the grid analysis, i.e., the amount of shear strain which the sand sustains at low shear stresses such as are encountered in molding practice.

# V. Discussion of Results

A. Mechanism of Molding Sand Flow in Cylindrical Dies—A molding sand consists of silica sand and a hydrated, plastic clay binder. Sufficient clay is always

TABLE 7—PARTICLE SIZE DISTRIBUTION OF SANDS USED IN DENSITY RATIO AND "SHEAR ANGLE" TESTS

Mesh Size	Grade 30–2	Grade 60-1	Grade 80–1	Grade 160-1
20	0.80	_	_	-
30	3.80	anne.	_	_
40	24.20	0.20	0.20	0.10
50	40.20	2.40	2.90	0.10
70	26.00	17.90	19.50	0.40
100	5.00	42.50	37.40	8.60
140	0.40	28.50	30.00	36.60
200	_	1.20	7.50	31.60
270	_	_	1.20	11.00
Pan	_	_	1.00	10.40

TABLE 8-SAND COMPOSITIONS USED IN DENSITY RATIO AND "SHEAR ANGLE" TESTS

Grade 30-2

5 lb 30-2 washed silica sand 106.51 grams sodium bentonite 62.70 milliliters water

Grade 60-1

5 lb 60-1 washed silica sand 119.0 grams sodium bentonite 70.0 milliliters water

Grade 80-1

5 lb 80-1 washed silica sand 119.45 grams sodium bentonite 70.30 milliliters water

Grade 160-1

5 lb 160-1 washed silica sand 123.67 grams sodium bentonite 72.80 milliliters water

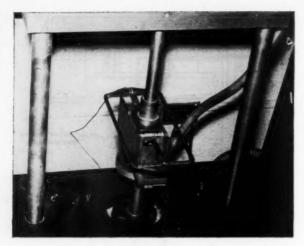


Fig. 19—Photo of "shear-angle" die in press prior to "shearing."

added to sand so each grain is completely covered with a hydrated clay film. The thickness of the film about each sand grain depends upon the kind and amount of clay in the sand mixture. The clay film also serves as a lubricant so the molding sand, when compressed, flows into a compact mass.

In an aggregate consisting of silica sand the grains are in direct contact with one another. Voids exist between sand grains, and their volumes depend upon grain shape. Little flow takes place when such a sand is compressed. Like most non-metallic crystalline substances, silica sand is brittle. It can be slightly deformed elastically; however, a sand grain fractures at stresses near its elastic limit with little, if any, plastic deformation. The fragments of the fractured crystals fit into the voids between sand grains. Elastic deformation is recoverable as soon as the applied stress is removed. Then, the net flow in silica sand is only due to the fragmentation of sand grains and the subsequent filling of intergranular voids with crystal fragments.

A clay slip is a suspension of clay particles in water. Occasionally other materials – for example, silica, alumina or titania – are added in varying amounts. Molding sands consist of silica sand, clay and water and they may be considered highly viscous casting slips. Norton, Johnson and Lawrence⁸ have studied the flow properties of clay-water suspensions. Their results show that clay-water suspensions possess low yield point values and are very plastic.

On the basis of these observations, molding sand flow is due to the plastic deformation of the claywater films about sand grains. The stress-strain curves in shear, presented on Figs. 14 to 17, not only verify this conclusion but also describe concisely how molding sands flow under pressure in a cylindrical die.

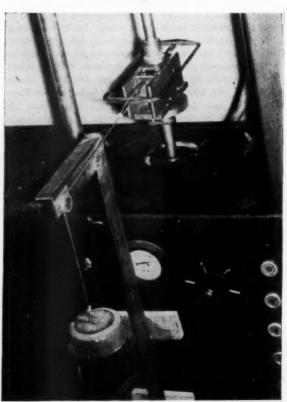
Foundry sand flow is the result of two different mechanisms. Initially, packing of loose molding sand in the die is due almost exclusively to the deformation in shear of clay-water films which have very little shear strength. This is evident since appreciable plastic deformation (d_o) is realized in the absence of appreciable shear stresses. In the second

stage of flow, deformation is dominantly due to the fragmentation of sand grains. In the first stage of the process, both sand grains and their surrounding clay-water films are displaced from their original positions. The primary stage of flow ceases when the clay-water films have sheared to a maximum extent and the majority of contact points are between sand grains only. Any subsequent flow is due to the fragmentation of the sand grains at these contact areas. Clay-water films have been forced from their positions about the sand grains into intergranular voids and have no further role in the deformation process. The transition between clay flow and sand fragmentation is sharp because only a very low shear stress is necessary to cause the total possible clay flow.

The sand grains do not fracture appreciably during molding operations. The average stresses obtained in these studies are well below the fracture stress for pure crystalline silica; however, some sand-sand contact areas are always present that can be fractured at low average stresses. Therefore, the stress-strain curve in shear increases after  $d_0$  with a finite slope.

The curves in Figs. 16 and 17 indicate that a sodium-bentonite bonded sand undergoes a greater initial deformation  $(d_0)$  than natural molding sand. Sodium-bentonite and kaolin, the binder in natural molding sand, behave altogether differently in an aqueous medium. Sodium-bentonite absorbs water within its lattice and then expands. This expansion of the crystal lattice lowers its lattice bonding forces and it then becomes possible to deform a completely

Fig. 20—Photo of "shear-angle" die in pres after "shearing."



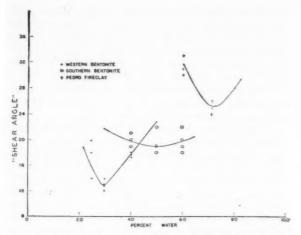


Fig. 21—Effect of moisture upon "shear-angle" for 60-1 and bonded with several clay binders.

hydrated sodium-bentonite particle at a very low shear stress. A hydrated kaolinite crystal is not capable of swelling. Consequently, its lattice bonding forces are not changed upon hydration and it is not deformed as easily as a hydrated sodium-bentonite crystal. Although the percentage of sodium bentonite clay is low compared to others used in molding sand mixtures, swelling increases its volume an amount dependent upon the extent of hydration.

Sand grain-shape appreciably affects the second stage of sand flow in cylindrical dies. Photomicrographs in Figs. 23 and 24 show that sand grains in natural molding sand are rounded whereas they are angular in the synthetic sand to which sodium-bentonite was added. The stress concentrations at contact points are very much higher in an angular sand; consequently, fragmentation must begin at lower applied stresses and the slope of the stress-strain curve in shear must be less in an angular sand. This difference is clearly shown in Figs. 16 and 17.

Fig. 23—Photo of round sand grains in a natural molding sand.

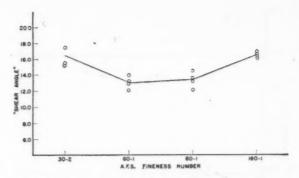


Fig. 22—Effect of sand-grain size distribution on "shearangle."

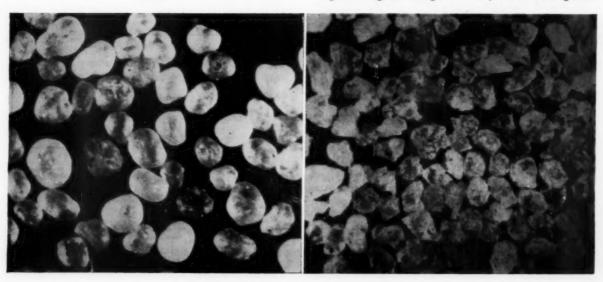
Figures 14, 15 and 16 indicate the method of loading—ramming or squeezing—does not materially affect the primary stage of deformation. The particle size distribution of the sand grains does not appreciably influence the initial deformation. This is reasonable since packing of loose molding sand is due almost exclusively to the deformation in shear of the claywater films about the sand grains.

The method of loading appreciably affects the fragmentation stage of molding sand flow in cylindrical dies. This is verified by the slopes of the curves in Figs. 15 and 16. The slope of the curve in Fig. 15 (rammed specimen) is the greater and, therefore, the molding sand has suffered less fragmentation for a given applied stress. On the basis of maximum deformation, squeezing rather than ramming is more effective for making sand compacts.

The density ratio,  $\rho/\rho_0$ , is plotted as a function of the shear stress per unit strain to determine the packing efficiency of molding sands in cylindrical dies. These plots are included in Figs. 25 to 28. The density ratio at the end of the primary stage of deformation is designated as  $(\rho/\rho_0)_0$ .

The plots in Figs. 25 and 26 are for sodium-ben-

Fig. 24—Angular sand grains in a synthetic molding sand.



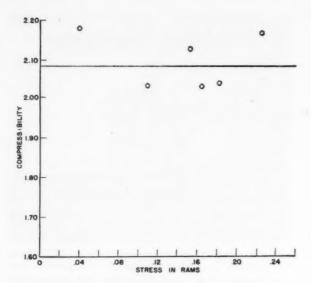


Fig. 25—Compressibility curve for 80-1 synthetic molding sand; one ram.

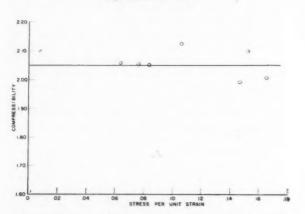


Fig. 26—Compressibility curve for 80-1 synthetic molding sand; four rams.

tonite bonded sands which have been deformed by one and by four rams respectively. The density ratio  $\rho/\rho_0$ , reaches its maximum value early in the deformation process. The slopes of the curves in Figs. 25 and 26 are zero. They clearly show there is no increase in packing even though we observe some deformation by fragmentation of sand grains. One ram is sufficient to produce maximum packing; the three additional rams have no beneficial effect with respect to densification.

The density ratio after pressing is 1.5066 and after ramming 2.0815 and 2.0512. The slope of the curve for rammed specimens is zero and it is greater than one for a pressed specimen. These values indicate a difference in packing; maximum density is attained early in the deformation process by ramming whereas the maximum occurs late in the deformation process for pressed specimens. On the basis of packing alone, ramming is more effective and is, therefore, preferable

The curves in Figs. 27 and 28 show packing of molding sands is materially affected by the type of

clay-water films about the sand grains. After squeezing, the density ratios of natural molding sand and sodium-bentonite bonded sand amount to 2.1777 and 1.5066 respectively. The slope of the curve for natural molding sand is less than that of sodium-bentonite bonded sand. On the basis of these observations, a natural molding sand approaches its maximum packing density much sooner in the deformation process than a sodium-bentonite bonded sand.

A comparison of Figs. 14 to 17 and 25 to 28 shows a relationship between the compressibility of a molding sand and its deformation process. The compressibility is great for those sands in which most of the total possible deformation occurs early in the flow process. If high compressibility is desired, sands which undergo maximum deformation early are then to be used. It is evident from the results described above that a sand of good flowability has poor compressibility. On this basis, tests may be designed for either property and, where properly interpreted, give information on both.

B. Density Ratio—A Measure of Compressibility—It is necessary first to determine the sensitivity of the density ratio method for evaluating the compressibility of different molding sands. In a satisfactory test, slight variations must be easily detected in the type of binder used, in the size distribution of the sand grains and in the moisture content. Figure 29

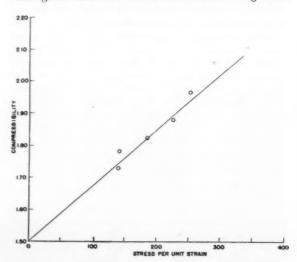


Fig. 27—Compressibility of 60-1 synthetic molding sand; squeeze.

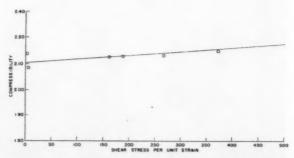


Fig. 28—Compressibility curve for 80-1 natural molding sand; squeeze.

shows the effects of the type of clay binder used and of the moisture content upon the compressibility of molding sands. Figure 30 illustrates grain size distribution effects upon the packing abilities of molding sands. Of the three clays used to bond 60-1 silica sand, sodium-bentonite permits maximum densification; calcium-bentonite and Pedro fireclay are not as effective with respect to compressibility. The density ratios are sensitive to slight changes in the moisture content of the molding sand. Maximum packing occurs in the neighborhood of 3 per cent water for sodium-bentonite bonded sand, at 5 per cent water for calcium-bentonite bonded sand and at 71/4 per cent water in fireclay-bearing sand.

The curve in Fig. 30 shows the effect of grain size distribution upon the compressibility of molding sands. Four grades of silica sand have been studied; the size distribution data are reported in Table 7. Molding sand compressibility increases with decreas-

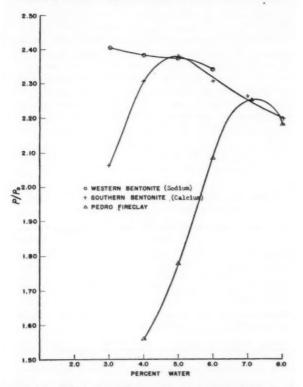


Fig. 29—Effect of moisture content on compressibility of 60-1 silica sand bonded with several clays.

TABLE 9-DENSITY RATIO DATA FOR 5 PER CENT SODIUM BENTONITE BONDED SANDS

	1 Ram	2 Rams	4 Rams
Height, in.	0.432	0.411	0.412
Weight, grams	31.77	31.14	31.90
Height, in.	0.406	0.437	0.397
Weight, grams	30.72	33.00	30.60
Height, in.	0.447	0.408	0.408
Weight, grams	32.90	30.84	31.60

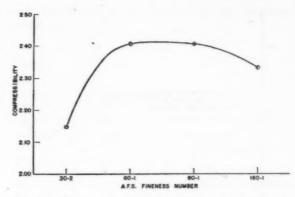


Fig. 30—Effect of grain size distribution on compressibility of molding sand.

ing particle size. Maximum packing is realized by molding sands that contain base sand grades, 60-1 and 80-1. These two molding sands are made from base sands which do not vary very much in the size distribution of the sand grains. Figure 30 also shows there is an optimum size distribution for maximum packing. Any further decrease in particle size causes the compressibility of the resulting molding sand to decrease measurably.

C. Classification of Flowability by the "Shear Angle" Test—Figure 21 presents the effects of the type of clay binder and the moisture content upon the "shear angle" of a 60-1 molding sand. The curves in Fig. 21 show sodium-bentonite bonded sands "shear" least. Calcium-bentonite- and fireclay-bearing sands shear appreciably more. The "shear angles" also vary with the moisture content in these sands. Minimum shear

TABLE 10-DENSITY RATIO DATA FOR 5 PER CENT SODIUM BENTONITE BONDED SANDS

	1 Ram	2 Rams	4 Rams
Height, in.	0.422	0.408	0.423
Weight, grams	31.00	30.82	32.75
Height, in.	0.417	0.442	0.409
Weight, grams	30.75	33.43	31.64
Height, in.	0.431	0.414	0.421
Weight, grams	31.75	31.25	32.60

TABLE 11-DENSITY RATIO DATA FOR 5 PER CENT SODIUM BENTONITE BONDED SANDS

	1 Ram	2 Rams	4 Rams
Height, in.	0.426	0.431	0.402
Weight, grams	31.65	33.00	31.40
Height, in.	0.442	0.432	0.410
Weight, grams	32.87	33.00	32.10
Height, in.	0.443	0.417	0.413
Weight, grams	33.10	32.00	32.20
Height, in.	0.418	0.413	
Weight, grams	31.10	31.65	

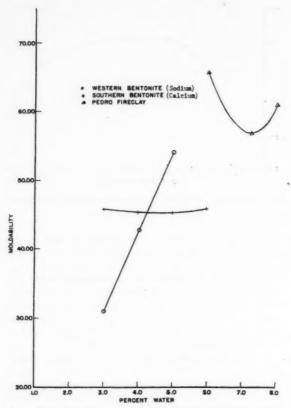


Fig. 31—Effect of moisture on moldability of 60-1 silica sand bonded with several clays

angles occur at 3 per cent water for sodium-bentonite bonded sands, at 5 per cent water for calcium-bentonite bonded sands and at 71/4 per cent water in fire-clay-bearing sands.

Figure 22 illustrates grain size effects upon the "shear angle." The angle decreases with decreasing particle size. The minimum "shear angle" is obtained for molding sands that contain silica sand grades, 60-1 and 80-1. Any additional decrease in particle size causes an increase in the "shear angle."

This test does indicate slight changes in moisture content, moderate changes in size distribution and differences between various clay binders. The nature of the results indicates sufficient sensitivity so that the test has practical application.

D. Moldability of Foundry Sands—From a practical point of view, molding sands should possess good

Table 12—Density Ratio Data for 5 Per Cent Sodium Bentonite Bonded Sands 6.0% Water

	1 Ram	2 Rams	4 Rams
Height, in.	0.445	0.409	0.425
Weight, grams	33.55	31.55	33.65
Height, in.	0.431	0.416	0.408
Weight, grams	32.25	32.00	32.31
Height, in. Weight, grams Average weight of s	0.418 $31.42$ samples = 32.41	0.410 31.35 grams.	0.417 32.90

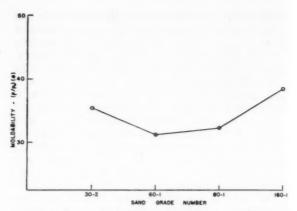


Fig. 32—Effect of grain size distribution on moldability of sands bonded with sodium-bentonite.

flowability as well as high compressibility. Unfortunately, such a combination is not possible since flowability is the converse of compressibility. The combination of compressibility and flowability values  $(\rho/\rho_0 \times \theta)$  provides an index which may be termed the index of moldability. This index then makes it possible to select the best combination of clay binder, sand-grain size distribution and moisture content, as is shown in Figs. 31 and 32. Sodium-bentonite bonded molding sands containing 3 per cent water yield maximum moldability. In a 5 per cent calcium-bentonite molding sand maximum moldability is found in a sand containing about 41/2 per cent water and in a fireclay-bearing sand ideal moldability occurs at 71/4

TABLE 13—DENSITY RATIO DATA FOR 5 PER CENT CALCIUM BENTONITE BONDED SANDS 3.0% Water

	1 Ram	2 Rams	4 Rams
Height, in.	0.489	0.487	0.483
Weight, grams	35.40	35.60	35.55
Height, in.	0.508	0.468	0.469
Weight, grams	36.52	34.66	35.37
Height, in.	0.491	0.479	0.483
Weight, grams	35.52	35.34	36.60
Average weight of	samples = 35.61	grams.	

TABLE 14—DENSITY RATIO DATA FOR 5 PER CENT CALCIUM BENTONITE BONDED SANDS

	1 Ram	2 Rams	4 Rams
Height, in.	0.424	0.440	0.427
Weight, grams	30.90	33.13	32.67
Height, in.	0.447	0.433	0.425
Weight, grams	32.61	32.65	32.55
Height, in.	0.428	0.428	0.416
Weight, grams	31.73	32.20	32.00
Height, in.	0.432		
Weight, grams	31.60		
Height, in.	0.452		
Weight, grams	33.40		

per cent water. Of the synthetic sands tested, a sodium-bentonite bonded sand with 60-1 silica sand as base material provides maximum moldability.

#### VI. Conclusions

Results of grid analyses show the flow of molding sands under pressure is the result of two different mechanisms. In the initial deformation, flow is due almost entirely to shearing of the clay-water films that coat the sand grains. When the clay-water films have sheared as much as possible and sand grains are in direct contact, any additional flow is due to fragmentation of sand grains. The transition between primary and secondary flow is sharp because only a low shear stress is required to cause maximum flow of the hydrated clay films about sand grains. These results also reveal the flow associated with the shearing of clay-water films is responsible for most of the total deformation.

Flow depends upon the type of bond used, upon the shape of the sand grains and upon the method of loading. Sodium-bentonite bonded sands undergo greater deformation than kaolin-bonded sands. Molding sands containing angular grains flow appreciably more than round-grained sands. Squeezing produces

greater overall flow than ramming.

Compressibility measurements also indicate the packing of molding sand is a function of the type of clay-water film used for bonding and of the method of loading. Kaolin-bonded sands pack more effectively than sodium-bentonite sands. Ramming rather than squeezing produces greater densification in a foundry sand.

There is a relationship between the packing of foundry sand and its deformation process. Sands of good flowability possess low compressibility.

Compressibility and "shear-angle" tests based upon grid-analysis results should prove to be practical test methods. These results are in excellent agreement with grid analysis results—namely molding sands of good flowability possess poor compressibility.

A combination of compressibility and flowability values provides an index of moldability. Such information should make it possible to select molding sands which posses the best combination of flow and

packing properties.

The writers have not included the laborious grid analysis (stress-strain) calculations in order to make this paper as brief as possible. They will furnish these calculations upon request.

#### Acknowledgment

The authors are indebted to George E. Schmidt, Jr., Research Assistant, Metallurgy Dept., M.I.T., for preparing the illustrations used in this paper, and to others of our foundry colleagues for assistance of various sorts.

#### APPENDIX A

The analysis of deformed grids, initially rectangular, for the magnitude and directions of the principal strains is the one developed by Seibel⁸ and by Kamm.⁶ They have shown that if the two dimensions of the deformed grid element—taken in the coordinate system of the original rectangle—are  $r_a$  and  $b_1$ , then the

new dimensions in the direction of the principal strains are:

$$r^2{}_{1,2} = \frac{r_a{}^2 + b_1{}^2}{2} \pm \sqrt{\frac{r_a{}^2 + b_1{}^2}{4} - r_a{}^2 \ b_1{}^2 \sin^2\!\gamma} \ (l)$$

These are also the semi-axes of the strain ellipse (Fig. A-1) of which one of the initial axes is a grid edge length. The angle d is the acute angle of the deformed grid element. In deformed grid elements that are not parallelograms, the quantities  $r_a$  and  $b_1$  and d are averages for the element.

Having determined the values of r, r_a and d, the magnitude of the principal strains are obtained from

the relations,

$$\epsilon_1 = \ln \frac{r_1}{r_0}, \qquad (2)$$

$$\epsilon_2 = \ln \frac{\mathbf{r}_2}{\mathbf{r}_2}$$
 (3)

and the direction of the principal strains with respect to the original coordinate system is found by

$$\tan \alpha = -\frac{\tan \alpha}{2} \left( 1 - \frac{r_2^2}{r_1^2} \right) \pm \sqrt{\tan^2 \alpha \left( 1 - \frac{r_2^2}{r_1^2} \right) - \frac{4r_2^2}{r_1^2}}$$
(4)

The values  $\epsilon_1$ ,  $\epsilon_2$  and  $\alpha$  are averages for each of the elements. They are assigned to their centers which are found by intersecting the diagonals of each element. In these calculations,  $\alpha$  is the angle between  $\epsilon_1$  and the direction of the axis of the compact.

Vectors representing the computed magnitudes and the directions of principal strains are plotted in the sectioned plane of the compact as shown in Fig. A-2. Strain trajectories are then drawn upon the sectioned plane of the compact by eye; the two families of trajectories must intersect only at right angles. A typical strain trajectory pattern is shown in Fig. A-3. If the material is assumed to obey the first rule of plasticity, the stress trajectories will coincide with the strain trajectories and the directions of the principal stresses may be obtained from strain trajectory plots.

Let us also assume that a relation similar in form to the generalized Hookes' law* exists at each set of principal stresses. Then it becomes possible to use the Laué-Maxwell equation to obtain the differences

$$Y_1 = f(\epsilon_i) \left[ \frac{R^{\circ}_{1\gamma}}{R_{1\gamma}} + \frac{1}{R_{1\gamma}} \right]$$
 (5)

between the magnitudes of the principal stresses at any two points along any trajectory. For this purpose, points are selected in such a way that three of them lie along each of several trajectories. The trajectories are chosen to form a grid of a density comparable to that of the original grid. The magnitudes

^{*} Third rule of plasticity.

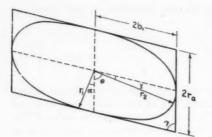


Fig. A-1—Strain ellipse of deformed grid.

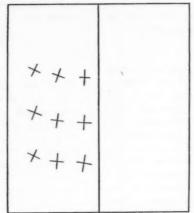


Fig. A-3 — Strain trajectory pattern for a molding sand specimen.

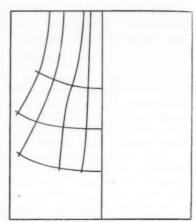


Fig. A-2—Sketch showing vectors and their directions in a deformed grid.

of the principal strains are found for each of these "selected points" by graphical interpolation. Let us assume that there is a continuous relation between the stress and strain tensors. Then it is possible to convert data from a separately determined stressstrain curve into the function  $f(\epsilon_1)$  which appears in the Laué-Maxwell equation. Such a stress-strain curve is constructed from a series of experiments in which thin wafers are rammed or pressed in a die; the wafers are thin enough so that the die walls have no effect on the flow of the material. The symbols Roiy and R₁ respectively represent the initial and final radii of curvatures of the stress trajectories perpendicular to the two families that are mapped in the sectioned plane of the compact. In the present case,  $R^{o}{}_{i\gamma}$  and  $R_{i\gamma}$  become the distances from the compact axis to the initial and final positions of the "selected points." Rik is the radius of curvature of the mapped trajectory passing through the "selected point" and is perpendicular to the trajectory along which the stress difference, Y₁, is computed.

In this manner, we have obtained a matrix of differences between every pair of neighboring "selected points." There now remains the problem of finding the absolute values of these stresses at the boundary points. The stress values at any other point may then be found by the integration—starting at the boundary—of the stress differences, Y₁, along the appropriate trajectories.

No information is available regarding the flow mechanism of molding sands; neither are there any known points at which a principal stress is zero or some other finite value. For this reason, it is necessary to make another assumption concerning the boundary conditions. It is assumed that the layer immediately below the plunger is not subjected to any lateral stress by friction against the plunger. Density determinations of the grid elements in this layer are then converted into vertical compressive stresses by means of stress-strain curves of thin wafers. This provides absolute values,  $\sigma_{10}$ , of the principal stress,  $\sigma_1$ , at the top of the  $S_1$  family of trajectories. The values of  $\sigma_1$  are obtained at every "selected

point" by integrating  $Y_1$  values along the  $S_1$  trajectories and by using the proper  $\sigma_{10}$  values as constants of integration. The values of  $\sigma_1$  at the centerline of the compact are then obtained by extrapolation along the  $S_2$  trajectories.

The boundary conditions can now be derived for  $\sigma_2$ . Cylindrical symmetry is assumed to exist throughout the specimen and the continuity of the relation between the stress and strain tensors is assumed to be valid. Now, should the observed density of the material along the centerline differ from that expected of the stress,  $\sigma_1$ , the difference is then probably due to the equal stresses,  $\sigma_2$  and  $\sigma_3$ . If  $\sigma$  ( $\rho/\rho_0$ ) is the stress corresponding to the observed density,  $\rho/\rho_0$ , at a point on the centerline where  $\sigma_1$  is  $\sigma$  (CL), then the value  $\sigma_{20}$  of  $\sigma_2$  at that point is

$$\sigma_{20} = \sigma \left( \rho / \rho_0 \right) - \sigma \left( \text{CL} \right) \tag{6}$$

The stress,  $\sigma_2$ , can now be evaluated at any "selected point" in the compact by the integration of  $Y_2$  along  $S_2$  trajectories and by using the appropriate  $\sigma_{20}$  values as constants of integration.

In this manner are obtained the values of the principal axial and radial stresses. The principal tangential stresses are equal to the principal radial stresses since cylindrical symmetry is assumed throughout the compact. For simplicity in drawing conclusions from the results, the principal stresses and strains are converted into shear stresses and shear strains thus,

$$T = \frac{\sigma_1 - \sigma_2}{2} \tag{7}$$

$$d = \epsilon_1 - \epsilon_2 \tag{8}$$

Stress-strain diagrams in shear are made for three molding sands. Since it is expected that the ratio, T/d, may vary with the density, T/d ratios are plotted as functions of the density ratio,  $\rho/\rho_0$ .

#### VIII. Bibliography

 L. M. Diran and H. F. Taylor, "The Bonding of Foundry Molding Sands," to be published.

2. H. W. Dietert and F. Valtier, "Flowability of Molding

Sand," Transactions, American Foundrymen's Society, vol. 42, pp. 193-210 (1935).

3. Testing and Grading Foundry Sands and Clays, American

Foundrymen's Society, 1938.

4. P. E. Kyle, "The Flowability of Molding Sands," TRANSAC-TIONS, American Foundrymen's Society, vol. 48, pp. 175-192

5. R. Kamm, M. Steinberg, J. Wulff, "Plastic Deformation in Metal-Powder Compacts," T.P. No. 2133, AIMME, February

6. R. L. Kamm, "Plastic Deformation in the Cold Pressing of Metal Powders," Sc.D. Thesis, 1947, Massachusetts Institute of Technology

7. E. Von Siebel, "Plastic Forming of Metals, III-Investigation of Distribution of Deformation and Stress in Drawing. Pressing and Piercing," Steel, April 19 and March 19, 1934.

8. F. H. Norton, A. L. Johnson and W. G. Lawrence, "Fundamental Study of Clay: VI, Flow Properties of Kaolinite-Water Suspensions," American Ceramic Society, Journal, vol. 27, pp. 149-164 (1944).

9. Sosman, The Properties of Silica, 1927, pp. 480-490.

10. J. M. Harian, Jr. and H. W. Christopher, "The Determination of the Clay Content of Naturally-Bonded Molding Sand and Strength Characteristics of Synthetically-Bonded Molding Sands with the Aid of the Euelin Apparatus," B. S. Thesis, 1949, Massachusetts Institute of Technology.

#### DISCUSSION

Chairman: O. J. Myers, Archer-Daniels-Midland Co., Minne-

Recorder: F. S. Brewster, Harry W. Dietert Co., Detroit.

V. M. ROWELL: 1 What is the relationship between force necessary to shear and force necessary to ram the sample?

Dr. TAYLOR: The compressive force of 7 psi is used merely to hold the sand together during shearing.

CHAIRMAN MYERS: Is there any relationship between the present flowability test and the one proposed in the paper?

Dr. TAYLOR: There is no relationship between the proposed

test and our previous concept of flowability.

H. W. DIETERT (Written Discussion):* The authors study of the compressibility of molding sands adds much more information on the subject of the moldability of sands. There are, however, several statements made in this paper with which I do not

In their review of previous work, they state that a number of molding sands may attain their maximum densities prior to the fourth ram. I have yet to find any molding sand that has reached its maximum density after the fourth ram of the A.F.S. sand rammer. Some may begin to reach maximum densities near ten rams while the majority require in the neighborhood of 20 to 50 rams (see Table A). Very few molding sands ever

TABLE A.—GREEN HARDNESS VS RAMMING

No. Drops A.F.S. Rammer Weight	Specimen Green Hardness
6	83.5
10	90
15	92
25	94
35	95
45	95.2
55	96
70	96

¹ Service Engineer, Archer-Daniels-Midland Co., Cleveland.

reach a flowability percentage of 90 with the Dietert-Valtier flowability indicator method after the fifth ram (see Table B). Thus, the statement made that "These molding sands all have a flowability index of 100 and it is impossible to determine their relative flow characteristics" is incorrect.

TABLE B-DIETERT FLOWABILITY VS RAMMING

No. Drops A.F.S. Rammer	1	Dietert Flor	wability	
Weight	3	5	6	8
1	0	0	0	0
2	14	6	5	10
3	52	49	49	41
4	68	65.5	67	61
5	77	76	76	72
6	82.5	81	82	79.5
7	86	85	85	83
8	88.5	87	87.5	86
9	90.5	90	90	88
10	92	91	91	90

A statement is also made that "mold hardness depends, to a great extent, upon the ease with which a molding sand flows together." One must not be led to believe that sands which show greatest mold hardness under equal ramming energy will necessarily possess better mold wall surfaces in reference to large void spaces not comparable with their grain size. In other words, a strong bonded sand with high compressive strength will possess a high mold hardness but possess large voids at mold surface while a lower bonded and will possess a lower mold hardness and much smaller voids at mold surface-each sand being of same grain size and rammed with same energy. (Referenceillustration page 198, A.F.S. TRANSACTIONS. vol. 48, 1940).

One should differentiate between the mold hardness obtained and false void spaces on mold wall surface. The mold hardness ability of a sand could be called rammability while the false void spaces have been termed flowability since 1935. Endeavoring to treat both mold hardness and false void space under one term is confusing and incorrect since hardness obtained and false voids obtained are two distinct properties.

MR. TAYLOR (Authors' Clesure): The authors are grateful for Mr. Dietert's remarks concerning the flowability of foundry sands.

On applying an external force upon a closed, rigid system filled with molding sand, the ccated sand particles move closer together and so assume a smaller volume. Initially, the flow is due to forcing the clay-water films coating the sand grains into inter-particle voids. This deformation occurs at low stress and proceeds until the voids between coated sand grains are filled with hydrated clay particles or until the sand grains contact one another. Further flow can only take place by fracturing sand grains; such a deformation requires applying relatively high stresses.

The extent to which foundry sand flows depends on the magnitude of stress required to deform the film coating the sand particles, on inter-particle void size, and on film thickness. Foundry sands which are bonded with films requiring low deformation stresses and having large inter-particle voids volumes as well as moderately heavy bonding films, undergo the greater deformation or flow. A sodium-bentonite bonded sand meets these requirements. Sands bonded with an oil-water emulsion do not flow appreciably under applied stress. This is due to the relatively low stress required to effect contact between sand grains. The Dietert-Valtier technique has yielded numerous values of 100 for sands of this kind, and so does not permit an accurate evaluation of their flowabilities.

The authors agree with Mr. Dietert that the term "flowability," which has been so loosely used since 1935, should be more carefully defined.

^{*} Harry W. Dietert Co., Detroit.

# METALLURGY AND MECHANICS OF HOT TEARING

By

H. F. Bishop,* C. G. Ackerlind* and W. S. Pellini**

#### ABSTRACT

Solidification conditions at the time of hot tearing were determined by simultaneous radiography and thermal analysis of solidifying castings. It was established that hot tearing occurs at near solidus temperatures at which state solidification is essentially complete except for liquid films. Corrolary strength tests coupled with thermal analysis to establish the state of solidification at the time of the tests showed that the development of high strength followed immediately upon the elimination of film regions. Variables such as C, P, S and pouring temperature were investigated by the above procedures and also by the use of castings of varied restraint design. The contributions of mechanical and metal variables to hot tearing were thus established. A new generalized theory of the mechanism of hot tearing is presented.

#### Introduction

Hot tearing, one of the most common and serious defects encountered in castings, is particularly difficult to control inasmuch as it involves an interaction of mechanical and metallurgical factors. While the primary mechanical factors of mold and casting design which favor hot tearing are generally understood, there is a lack of agreement concerning the effect of such metallurgical factors as metal and solidification variables. It is an accepted fact that hot tears occur when the contraction of the solidifying castings is excessively restrained by the mold or cores and that the tears occur in weak areas where the strain resulting from the contraction is concentrated. These weak areas are localized hot spots such as gate and riser contacts, re-entrant angles or enlargements in the castings. Hall1 has shown the relation of hot zones and restraint to tearing by casting round bars with enlargements at their center and flanges on each end. He found that hot tearing was eliminated when the degree of strain concentration in the hot zone was sufficiently reduced by increasing the length of the central bulb.

A number of investigations of metallurgical factors contributing to hot tearing have been reported in

the literature. However, the conclusions are for the most part highly contradictory. Gelperin2 considered that hot tears are caused by the allotropic transformation of the delta phase into the gamma phase since this transformation is accompanied by a marked volume contraction. He accordingly concluded that as the carbon is increased the steel is less prone to tear. Singer³ also concluded that high carbon steels tear less readily than low carbon steels, and ascribed the difference to the wider solidification ranges of the high carbon steels which increase solidification time thus permitting temperature equalization within the casting. Juppenlatz4 on the other hand claimed that a low carbon steel will not tear as readily as a high carbon steel because it has a narrow solidification range and solidifies faster than a high carbon steel thus acquiring strength and ductility earlier. Lutz and Hickey⁵ concur with Juppenlatz's findings. Briggs and Gezelius^{6,7} measured the amount of contraction of both unhindered and restrained solidifying steel bars. They investigated plain carbon steels having carbon contents from 0.08 to 0.90 per cent and also several alloy steels and show that under like restraint the contraction characteristics of all of the steels in the austenite temperature range were approximately the same. Thus, insofar as contraction factors are concerned neither carbon content nor alloy content should influence the tearing properties of steel.

It is generally agreed that sulphur and phosphorus favor hot tearing in steel although the exact mechanism by which this occurs is not definitely known. Phillips⁸ has observed that steel will tear more readily when sulphides are present as stringers rather than as globules and suggests that periodic aluminum additions be made to the ladle while pouring to prevent stringer inclusions. Hall^{1,9} reports that both sulphur and phosphorus markedly lower the temperature at which steel acquires ductility in cooling. Singer and Benek¹⁰ observed hot tears which had been partially filled with sulphur and phosphorus-rich liquid, indicating that these elements cause the retention of liquid segregates to temperatures at which hot tears

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The opinions expressed in this paper are those of the authors and do not necessarily reflect the opinions of the Navy Dept.

There is a general acceptance that hot tearing of steel occurs considerably below the solidus temperature. Korber and Schitzkowski¹¹ found that the contraction of restrained bar castings having diameters of 20 to 50 mm and containing from 0.15 to 0.39 per cent carbon deviates from normal at about 1300 C (2370 F) indicating fracture. Hall^{1,9} and Briggs¹² made tensile tests on solidifying bar castings and found that a medium carbon steel at 1300 C had a strength of 1800 psi and an elongation of 5 per cent. Such high ductility at accepted hot tearing temperatures is difficult to reconcile with the brittle appearance of hot tears.

Middleton and Protheroe¹³ recently conducted tests on cast steel bars restrained by means of springs and found that steels containing 0.07 to 0.40 per cent carbon fractured at temperatures between 1420 and 1365 C (2590 and 2490 F). The bars which were 24 in. long and had a cross-sectional area of 1 sq in., broke under loads of 10 to 200 lb. Middleton and Protheroe's work is difficult to interpret in that the tearing temperatures which they report are the temperatures at which their bars fractured completely; no temperatures were reported for the bars which were only partially torn.

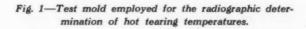
While all of the investigations on steel indicate that tearing occurs at temperatures appreciably below the solidus, Vero¹⁴ and Lees¹⁵ by determining the temperatures at which aluminum alloys acquired strength concluded that hot tearing in aluminum alloys occurs near the solidus temperature. While the temperatures of hot tearing in steel as determined by Middleton and Protheroe are appreciably higher than the temperatures determined by Korber and Schitzkowski, they are still considerably lower than the

solidus temperature of 0.07 to 0.40 per cent carbon steel which varies between approximately 2710 and 2650 F.*

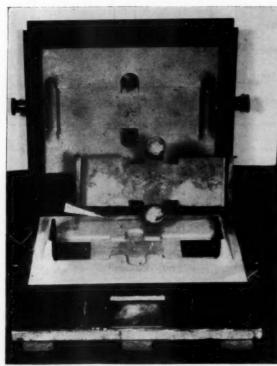
The effect of pouring temperature on hot tearing is debatable with many such as Singer¹⁰ believing that hot pouring will minimize hot tearing; while others believe that cold pouring is beneficial. Middleton and Protheroe conclude from their studies that the temperature of hot tearing is lowered and the ease of tearing is reduced by lowering the pouring temperature.

It is evident that the underlying difficulty in the interpretation of the many investigations is a lack of basic information as to the mechanism of hot tearing. Such information is prerequisite to any approach aimed at separating and evaluating the contribution of various foundry variables. An investigation was accordingly planned to determine by direct experiment the conditions of solidification at which hot tearing occurred in order that the mechanism of hot tearing could be established. This investigation was conducted by x-raying castings during solidification and relating the time of origin of hot tears to the existing condition of solidification deduced by thermal analysis. A simple casting system was utilized which permitted a wide range of control over the

[•] The difficulties encountered in the measurement of temperatures in small steel castings are great and unless adequate precautions are taken the measurements will be erroneous. To minimize conductivity losses through even fine thermocouple wires, the couples should be immersed in the casting for a distance of at least 1 in. Experiments at Naval Research Laboratory have shown that unless this is done the readings may be several hundred degrees low. The thermocouples cannot be used unprotected because of the ease with which they are contaminated and yet the protection around the couples, usually quartz tubing, must be sufficiently small to prevent a chilling effect upon the metal and also to minimize the heat loss through the tube. The temperature measurements reported by many of these investigators are highly questionable since these precautions were generally not followed.







contraction conditions. This system was also used to investigate the effect of metal variables on the severity of contraction (mechanical) conditions which the casting hot spot can withstand. In addition the development of strength and ductility through he period of solidification was determined by tensile type tests at various stages of solidification utilizing again a sample casting system amenable to exact thermal analysis. The various findings were finally interrelated to deduce the interactions of metallurgical (metal and solidification) and mechanical (stress and strain) factors which determine the development of hot tears.

# Temperature of Hot Tearing

Method of Procedure-The temperature at which hot tears originate was determined by means of a series of radiographs of a casting while it solidified and cooled in the mold. Tests were first made on 4 per cent copper-aluminum alloy because of its low x-ray density and then extended to steels. The test casting and molding procedures used are illustrated in Fig. 1. The castings were 1 in. thick for aluminum and  $\frac{1}{2}$  in. thick for steel. Restraining bars were placed between the flanges to provide a reproducible condition of hindered contraction and chills were placed against the flanges and over the horizontal portion of the castings adjacent to the flanges to accelerate cooling in these regions and hence cause a rapid contraction of the casting. The resultant strain was thus concentrated at the hot central portion of the casting between the two risers. The mold above and below the central portion of the casting between the risers consisted only of 1-in. thick layers of sand reinforced with perforated 1/16 in. thick steel sheet.

X-ray exposures through the central sections of the castings were obtained at intervals of 30 to 60 sec on film placed in the opening beneath the casting. Type K film was used and each exposure required 10 sec at 10 milliamperes; the kilovoltage for the aluminum and steel castings was 120 and 220 respectively. Since the x-rays had to penetrate a total of 2 in. of sand and ½ in. of reinforcing plate in addition to the casting itself, the sensitivity obtained was approximately 5 per cent of the casting thickness. The first stages of hot tear formation could be readily determined with this sensitivity since the direction of the x-ray beam was optimum for recording hot tearing.

The cooling cycles at the center planes of the castings between the risers where tearing occurs were determined by means of thermocouples connected to automatic recorders. Twenty eight gage chromelalumel couples were used in the aluminum castings; the thermocouple wires were separated from each other by means of two-hole alumdum tubing and protected with  $\frac{3}{16}$ -in. op quartz tubing. To minimize conductivity losses the end of each thermocouple assembly within the mold cavity was bent at a right angle so that at least 1 in. of the thermocouple lay along the center plane of the aluminum plate.

Twenty-six gage Pt-Pt-Rh (13%Rh) thermocouples were used to measure steel temperatures; since the steel castings were ½ in. thick and cooled more

rapidly than the aluminum castings protection tubes of only ½6-in. on were used in order to permit the couples to be surrounded with the maximum amount of molten metal. These small tubes were bent in the shape of a square "U" with the bottom of the "U" being about 3 in. long and located in the casting center plane. The thermocouple bead was positioned at the center of the "U" with a thermocouple wire leading out through each end of the tube. Thus the thermocouple was immersed in the center of the casting for the equivalent of at least 1½ in.

Temperatures were measured at the casting center plane because of the greater accuracy which could be obtained. It was found that in both the aluminum and steel castings the temperature differential between the surface and center plane did not exceed 10 F and that the center temperature lagged that of the surface by only a few seconds because of the rapid cooling at the critical solidus temperature range. Thus, the temperatures measured at the center plane may be considered to be effectively the temperature of the entire section. Typical cooling curves for aluminum and steel castings are presented in Fig. 2.

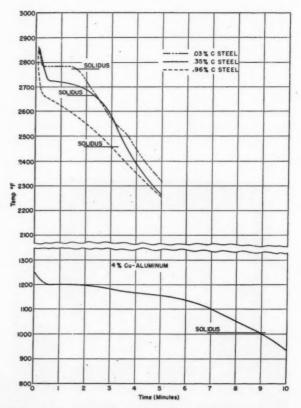


Fig. 2—Typical cooling curves for aluminum and steel test castings.

#### Hot Tearing in Aluminum

Approximately 20 test castings were made with the 4 per cent copper-aluminum alloy melted in a high frequency induction furnace and poured at 1300 F. The temperature at which the first tears were ob-

served as well as the temperature of the preceeding radiograph which did not show tears are listed in Table 1. It can be noted that the tears are present at temperatures between 1000 F and 1035 F. The solidus temperature of the aluminum alloy as determined by inverse rate curves was between 1000 and 1010 F. Thus tearing begins at or slightly above the solidus temperature at which time solidification is very nearly complete and the remaining liquid is present as interdendritic films. Figure 3 shows the temperatures of the two critical x-ray determinations for all the test castings in relation to the solidification state of the metal.

Table 1—Determination of Hot Tearing Temperature in 4% Copper-Aluminum Alloys by X-ray Methods

Test No.	Sound	Castings	Castings	With	Shrinkage
	lst Hot Tear Noted, °F	Preceding	Tear No.	Tear	Temp. of Preceding X-ray, °F
29A	1035	1075	17	1030	1110
29B	1020	1065	22	1075	1125
32	1025	1050	24	1050	1115
34A	1000	1025	28	1030	1080
34B	1010	1050	31	1075	1105
35	1015	1050	33	1040	1090
40	1030	1065	36	1040	1070
40	1030	1070	39	1060	1105
***	2000		43	1070	1130

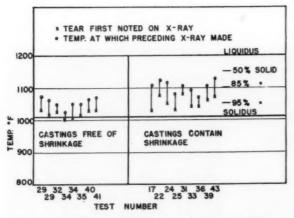


Fig. 3—Summary data indicating hot tearing temperature range in 4-pct Cu-Al alloys.

In the course of this work a few castings were found to contain shrinkage pockets and in these castings tearing began at somewhat higher temperatures than in the completely fed castings. In order to determine more definitely how the presence of shrinkage affects the temperature of hot tearing of aluminum, several castings were deliberately made to contain shrinkage by reducing the riser diameter from 3 to 2 in. and the riser contact to half the plate thickness. The temperature at which tears were first noted in the inadequately fed castings varied from 1020 F to 1075 F as shown in Table 1; the average tearing temperature is in the order of 30 F higher than for the shrink-

age-free castings. At these higher temperatures the solidification conditions differ from those of temperatures nearer the solidus primarily in the width of the liquid film regions, as may be deduced from Fig. 3. Figures 4 and 5 show series of radiographs for castings with and without shrinkage. It can be seen that the hot tears originate in the shrink area which was invariably the case for castings which developed shrinkage.

# Hot Tearing of Steel

Similar radiographic observations and thermal measurements were made on steel castings of various carbon contents ranging from 0.03 to 1.00 per cent. The steels were melted in a basic induction furnace using an armco iron charge. All steels contained approximately 0.60 per cent manganese and 0.40 per cent silicon; the carbon was adjusted to the desired level by additions of Mexican graphite. Except for the case of special additions the sulphur and phosphorus content was 0.02 per cent. Final deoxidation was accomplished by the addition of 0.10 per cent aluminum in the ladle and the castings were poured with a superheat of approximately 150 F.

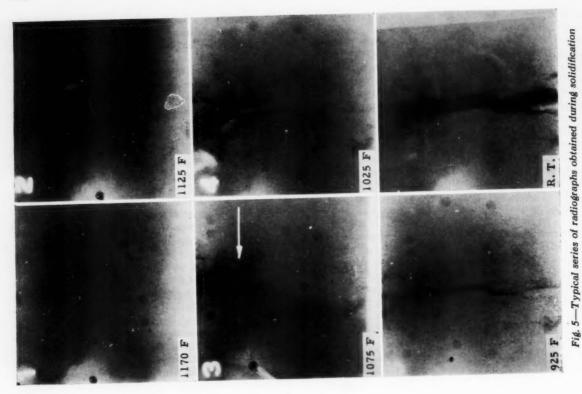
The steel castings, because of their smaller section thickness and higher solidification temperatures, cooled at a much faster rate than the aluminum castings, Fig. 2, and the temperature differences between successive x-rays were accordingly somewhat greater. As with aluminum, tearing was considered as starting between the temperature at which the tear was first noted on the radiograph and the temperature at which the preceding radiograph was taken. These two temperatures for each test are listed in Table 2. Typical series of radiographs made during the solidification of a 0.17 and 0.33 per cent carbon steel are shown in Figs. 6 and 7 respectively.

Table 2 shows that as carbon content increases

TABLE 2-DETERMINATION OF TEARING TEMPERATURE IN STEELS BY X-RAY METHODS

Heat No.	C, %	1st Tear Observed °F	Temp of Preceding X-ray, *F		18
92	0.03	2765	2780	2775	
80	0.07	2685	2760	2750	
73	0.17	2695	2725	2715	
68	0.22	2675	2700	2710	
61	0.25	2610	2660	2700	
58	0.28	2660	2670	2680	
82	0.31	2610	2675	2680	
101	0.31	2650	2690	2680	*No Restraining Bar
63	0.32	2600	2625	2675	or Horizontal Chills
62	0.33	2600	2670	2675	
67	0.33	2635	2660	2675	
79	0.35	2645	2685	2670	No Restraining Bar
74	0.41	2635	2685	2645	or Horizontal Chills
66	0.57	2600	2660	2580	
76	0.71	2515	2560	2545	*No Restraining Bar
86	0.78	2495	2540	2505	
95	0.96	2470	2515	2470	
100					

*Test castings from a series aimed at deducing the possible effect of decreased levels of contraction; all other tests conducted with standard casting having restraining bar, horizontal and vertical chills.



1065 F 1020 F 940 F 950 F 950 F 500 F

Fig. 4—Typical series of radiographs obtained during solidification and cooling of 4-pct Cu-Al test casting.

and cooling of inadequately risered 4 pct Cu-Al test casting.

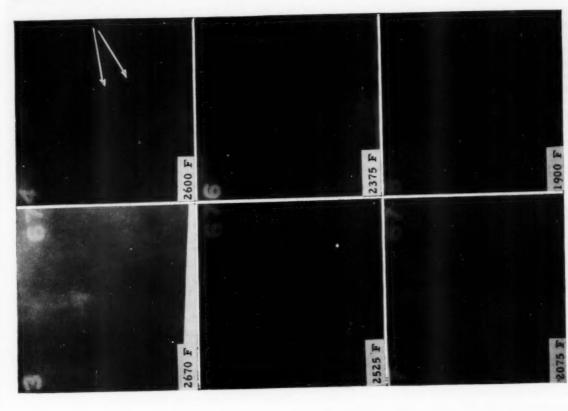


Fig. 7—Series of radiographs obtained during solidification and cooling of 0.33-pct C steel test casting.

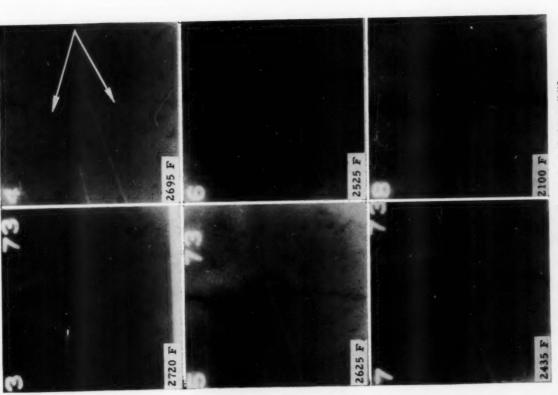


Fig. 6—Series of radiographs obtained during solidification and cooling of 0.17-pct C steel test casting.

there is a gradual and consistent lowering of the hot tearing temperature. When these data are plotted on the Fe-C constitution diagram as tearing temperature range vs carbon content, Fig. 8, it is noted that tearing over the whole range of carbon contents begins in the vicinity of the solidus temperature. It was observed in this series of tests that the tears become somewhat less pronounced as the carbon content is increased.

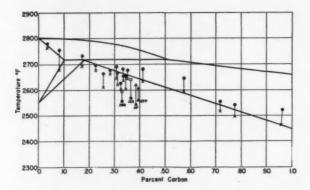


Fig. 8—Summary data indicating hot tearing temperature range in steels as related to the Fe-C constitution diagram.

In order to relate the effects of sulphur and phosphorus to the temperature of hot tearing, the contraction conditions of the test casting were modified to the point where the low sulphur and phosphorus steels no longer tore. This was accomplished by reducing the length of the test casting from 24 to 16 in. and by eliminating the horizontal chills and restraining bars between the flanges. The end chills were retained in order to produce an early solidification and rigidity in the flanges. The actual steps taken in the development of test casting having a hot tearing tendency slightly below that required to produce tears in the base analysis clearly illustrate the

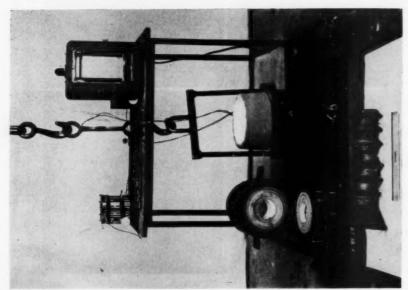
effects of mechanical variables. The restraining bars were first removed to reduce the level of restraint; this had little effect on hot tearing, indicating a high level of restraint by the sand. Next the horizontal chills were removed to decrease the time-rate of contraction; this resulted in an appreciable decrease in the hot tearing produced. Finally the length of the casting was reduced to 16 in. to decrease the amount of total contraction; this final change was sufficient to completely eliminate hot tearing. These variations in the mechanical condition did not affect the temperature of hot tearing, Table 2.

The shortened test casting was then poured with steels to which phosphorus or sulphur was added. Abnormally high quantities were included in the series in order to aggravate the effects produced by these elements so that clear-cut trends would be indicated. It was found that each of these elements produced tears in the shortened castings, but the tears began at a temperature approximately 50 F below the solidus (Fe-C) temperature. The temperatures at which the two critical radiographs were taken in each of the tests are listed in Table 3 and are also indicated on the iron-carbon diagram, Fig. 7.

The effect of inadequate feeding on the temperature of hot tearing was not studied in the steel castings because of the small section thicknesses involved which made it impossible to create gross shrinkage; thicker sections could not be used because of the limitations of the x-ray equipment.

TABLE 3-EFFECT OF IMPURITIES ON HOT TEARING TEMPERATURE RANGE DETERMINED RADIOGRAPHICALLY

Heat No.	C, %	S, %	P, %	First Tear Observed, °F	Preceding	Fe-C Diagram Solidus Temp., °F
83	0.36	0.12		2575	2635	2670
96	0.33	0.09		2615	2640	2675
91	0.38		0.07	2535	2635	2660
98	0.38		0.11	2550	2610	2660



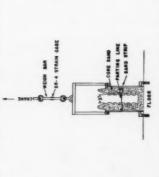


Fig. 9-Tensile test apparatus for soliditying castings

11

# Strength Tests of Solidifying Steel

To determine the time and rate at which steels acquire strength as they solidify, 3-in. diam tensile bars were cast into the system shown in Fig. 9 and loaded in tension when they had cooled in the molds for different time intervals required to develop known thicknesses of solid skin. The mold consists of two strong cores fitted into a round two-part flask, one forming the cope, the other the drag. Circumferential ribs on the cope and drag halves provide the necessary grip action required to transfer the load from the sand to the casting. The drag half of the flask is fixed to the floor while the cope half is attached to a crane through an SR-4 weigh bar. As the load is applied to the casting the cope is lifted away from the drag and the casting fractured near the mold parting line. The load necessary to cause fracture is measured by means of a direct reading photo-electric potentiometer connected to the SR-4 strain gage on the weigh bar. Such tests were made on 0.25 to 0.30 per cent carbon steels containing less than 0.02 per cent sulphur and phosphorus and also on similar steels which contained additions of sulphur and phos-

The strength data obtained from these tests are plotted in Fig. 10. The cooling cycles at four locations along the midplane of the test bar as well as the "start of freeze" and "end of freeze" curves de-

duced from the cooling cycles¹⁶ are also shown so that the relation of casting strength to the condition of solidification can be observed.

It can be seen that the load-bearing capacity of the cast bar of base analysis is virtually nil until a solid skin as indicated by thermal analyses begins to form. Thereafter, the load-bearing capacity rises rapidly and at a time that 1/4 in. of solid skin has formed approximately 400 lb load is required for rupture.

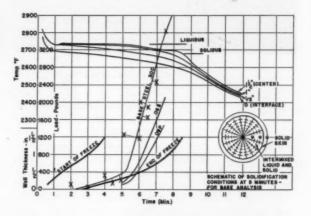


Fig. 10—Summary data indicating the development of strength of various steels with solidification conditions.

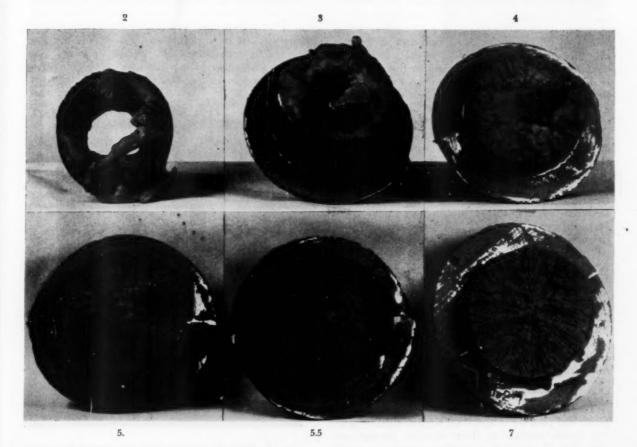


Fig. 11-Fracture appearance of base steel test bars broken at the indicated times after pouring.



Fig. 12-Dendritic appearance of hot tear fractures; crystalline areas were untorn.

At the time 1 in. of solid skin has formed the fracture load of the 1-in. thick ring is approximately 4000 lb. (Sand core fractures occurred above this load level). These tests indicate that low strength at temperatures considerably below the solidus cannot be considered as a factor leading to hot tears. In this respect it should be noted that at a time when the 1-in. solid skin is supporting a load of 4000 lb, the surface regions of the casting have cooled a maximum of only 60 F below the solidus. It can be noted that the effect of sulphur and phosphorus additions is to delay the time at which steel acquires strength. For example, the time at which steels containing added sulphur and phosphorus acquire a strength of 400 lb is delayed for approximately 1 min and at comparable times during the remainder of the solidification these steels are inferior in strength to the base steel. It is interpreted that the strength delay effect of these elements results from the development of liquid film segregates which effectively produce a lowered solidus temperature. The true end of freeze curve repressenting complete elimination of liquid is thus shifted to longer times by these elements.

Figure 11 shows a series of fracture surfaces of the test bars broken at different time intervals. There was no gross bleeding in bars broken after 4 min, although drops of interdendritic liquid were observed to bleed from the broken bars up to about 51/2 min after pouring. Liquid metal after 51/2 min is retained in the broken casting by capillary action. All of the broken castings showed interdendritic fractures similar to those of the hot tear test castings, Fig. 12, except close to the surface of the bar castings which had developed an appreciable solid skin. This may be observed from the photograph of the 7-min test bar which shows a 1/4-in. surface zone of fracture having a ductile appearance. This zone was approximately 60 F below the solidus at the time of fracture and must be considered to be devoid of films from the fracture appearance.

#### Metallurgical Factors Affecting Tearing

The qualitative effects of carbon, phosphorus, sulphur, and pouring temperature were studied by pouring molds, containing three separate flanged castings, Fig. 13, designed to produce contraction conditions varying from mild to severe and hence to pro-

duce different hot tearing tendencies due to variations in mechanical factors. The metallurgical effects were therefore evaluated on the basis of the severity of the mechanical condition required to cause tearing. The casting which developed the most rapid contraction was 24 in. long and heavily chilled at both the flanges and the horizontal portions of the casting adjacent to the flanges; an intermediate rate of contraction was obtained in a similar casting which was completely unchilled, while the third casting which developed mild contraction rates was only 16 in. long and also unchilled. These castings will be referred to as I, II and III in order of increasing severity of contraction conditions. Each of the castings was fed by two risers as indicated in Fig. 13; the metal entered the mold through the risers from a common runner basin. Although this gating method resulted in castings having somewhat dirty sur-

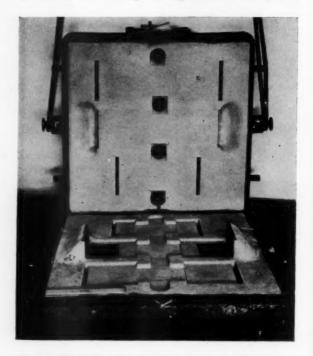


Fig. 13—Mold for triple hot tear test casting, incorporating three levels of mechanical restraint.

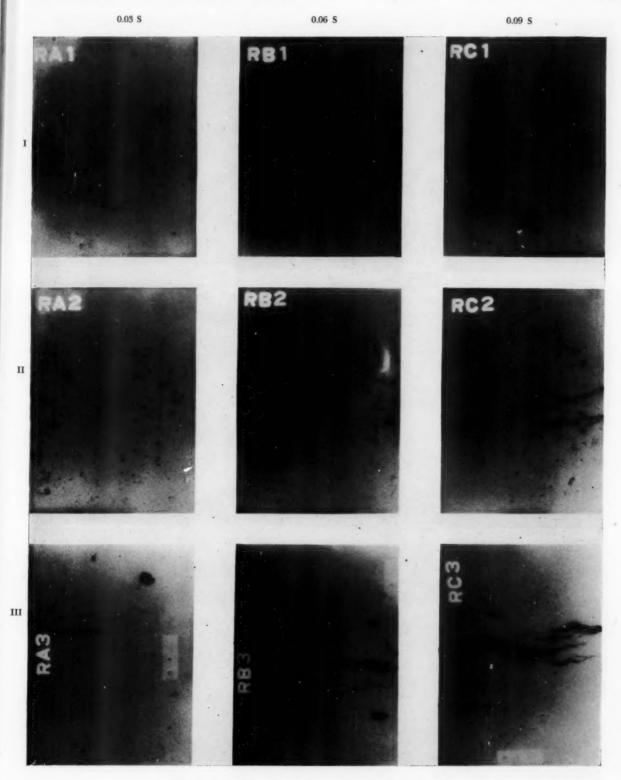


Fig. 14—Radiographs showing the interrelation of increasing sulphur (left to right) and increasing mechanical restraint (top to bottom) on the hot tearing tendencies of an 0.31-pct C steel.

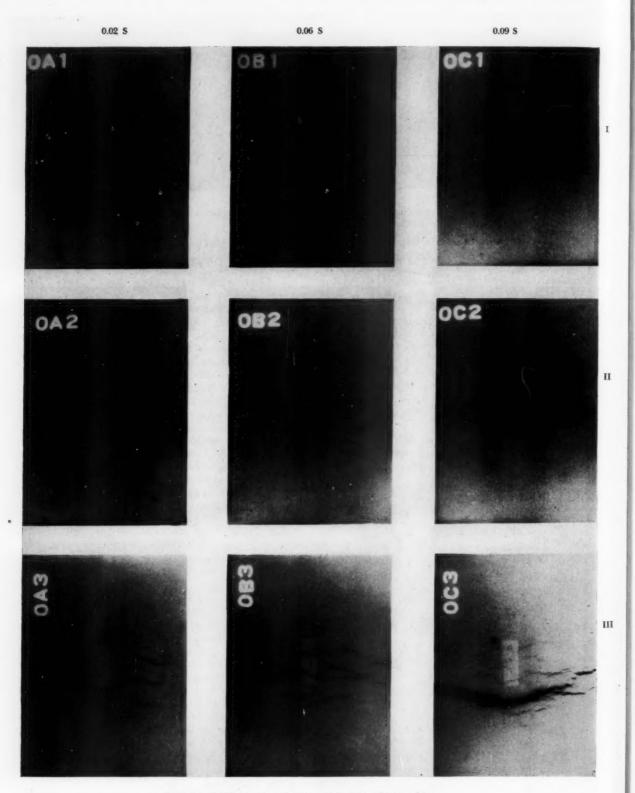


Fig. 15—Radiographs showing the interrelation of increasing sulphur (left to right) and increasing mechanical restraint (top to bottom) on the hot tearing tendencies of an 0.59-pct C steel.

II

I remained untorn.

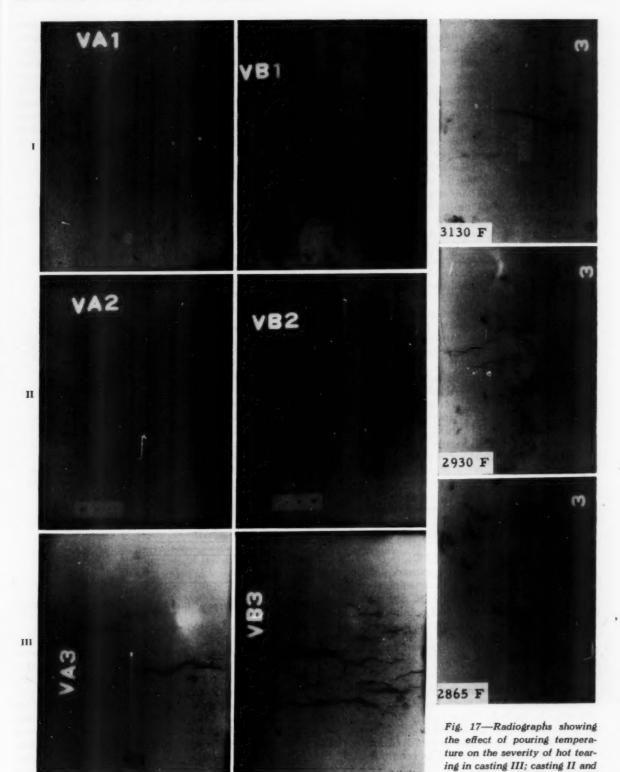


Fig. 16—Radiographs showing the interrelation of increasing phosphorus (left to right) and increasing mechanical restraint on the hot tearing tendencies of an 0.38-pct C steel.

faces it permitted rapid uniform filling of the thin casting section. These castings were poured with superheats of approximately 150 F except for a pouring temperature series. Two or three such molds were poured from one heat of metal, the heat being split into fractions containing varying amounts of the element being investigated. Thus the effect of each increment of the variable element was observed under three conditions of restraint.

The effect of sulphur content on the hot tearing tendencies of 0.30 and 0.60 per cent carbon steels is shown in Figs. 14 and 15 respectively. Each horizontal row of radiographs in Figs. 14 and 15 represents the same contraction condition ranging from the mildest (I) condition at the top to the most severe (III) condition at the bottom. Each of the three vertical columns represents a steel having the same sulphur content with the lowest at the left and the highest at the right. As is to be expected, increasing sulphur content increases the severity of tearing and causes tearing to occur with conditions of lesser severity of contraction.

Figure 16 shows in a like manner the reaction of steels containing 0.03 and 0.07 per cent phosphorus to the three different contraction conditions. It can be observed that phosphorus has an effect approxi-

mately equal to that of sulphur.

Pouring temperature variations were found to have no significant effect on hot tearing in these test castings. A 0.30 per cent carbon, low sulphur and phosphorus steel poured at temperatures between 2865 and 3130 F tore only in casting III representing the most severe contraction condition and these tears were essentially of equal severity as can be noted in Fig. 17.

In these test series as well as the preceding radiographic study series the tears in the high carbon steels appeared to be slightly less severe than in the low carbon steels. The differences, however, were relatively minor and possibly are not significant.

## **General Summary**

The interdendritic fracture appearance of hot tears and their origin at near-solidus temperatures indicate that the basic mechanism of hot tearing is a separation at the film stage of solidification which exists when the solidus temperature is approached and only a minute amount of liquid remains. This conclusion is further supported by the evidence of the high strength and ductility of metal which has passed through the film stage.

A number of interacting variables determine whether or not tearing will actually occur when the critical film stage is reached. These factors fall into the two general categories of metal and mechanical variables. Metal variables cover factors which determine the nature and duration of the film condition and are primarily a function of the metal composition; mechanical variables include the factors which determine hot spot and contraction conditions and are a function of casting design and molding techniques.

The primary requirement for hot tearing is the development of a hot spot which must extend to compensate for the contraction of solid regions. The

rigidity of the mold is a factor inasmuch as yielding of the mold wall is subtractive to the extension forced on the hot spot. The extension forced on the hot spot results in hot tear fractures when the critical film stage condition is reached because of the severe concentration of the overall extension into these zones to produce high unit strains. It is believed that the nature of strain distributions resulting from the extension of the hot zone at various stages of solidification are as shown schematically in Fig. 18.

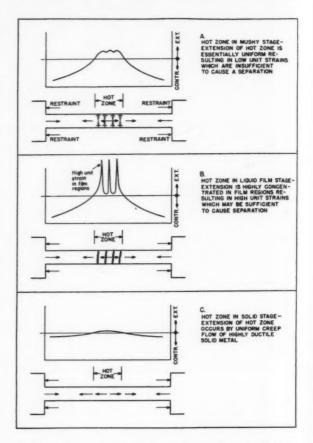


Fig. 18—Schematic illustration of the basic mechanism of hot tearing.

While the metal is in the mushy condition which exists at temperatures considerably above the solidus, Fig. 18A, the extension of the hot zone is distributed relatively uniformly since the interdendritic liquid areas are relatively wide and mass flow of the pasty mass results. Furthermore it is possible that at this stage of solidification any tendency to separation can be countered by flow of feed metal. As the hot zone approaches the film stage the extension in the hot zone necessarily is forced into the liquid films which have no appreciable strength compared to the adjoining solid dendrites. Figure 18B illustrates the high value of unit strains developed in these regions at this time. The width of the overall hot zone in a casting must be considered as a factor in determining if tearing will occur. If the hot zone is narrow and

only a few films exist, each film must accomodate a great amount of extension; a wide overall hot zone will have many films and the extension per film will be less. This factor may explain why Hall could eliminate hot tearing in his test bars by lengthening the bulb which formed the hot zone. If separation does not occur at the film stage of solidification, hot tearing is no longer possible, Fig. 18C, because the strains in the hot spot area are distributed in a relatively uniform fashion through the coherent and ductile solid metal. Stresses will build up in the casting as cooling continues until the low yield point of the hot metal is exceeded after which time creep flow occurs. Tears which occur during the critical film stage enlarge primarily in width as the casting contracts further on cooling below the solidus temperature due to stressing across the notch area created by the tear. Enlargement of the tears in a true hot tearing fashion does not occur at emperatures below the true solidus. The enlargement in width is due to plastic yielding at the end points of the tear.

The various observations of the interrelated effects of metal and mechanical variables resulting from this investigation permit further generalizations of

the hot tearing problem.

1. The time-rate of extension of the hot zone during its film stage is the primary mechanical factor which determines hot tearing. The film condition stage lasts for only a relatively short fraction of the total solidification time as can be observed by the accelerated rate of cooling of the metal as it passes through the solidus temperature. Hot tearing will occur only if a sufficient rate of extension is forced on the film regions during this short time.

2. The rate of extension forced into the film zones is related to (a) the overall extent of the hot zone, (b) the amount of mold restraint, (c) the length and the rapidity with which the adjoining sections cool

relative to the hot zone.

3. The presence of elements such as sulphur extends the film condition to lower temperatures and therefore increases the duration of the film stage.

4. The longer the interval of film stage the greater the total amount of separation which is developed in the films. Thus, increasing the duration of the film stage permits hot tearing to result with lower time-rates of extension than would be necessary to cause hot tearing in the absence of film-forming elements. Tearing moreover occurs at a lower temperature because of the longer time required to build up the extension to the critical value for separation.

5. The presence of sulphur and phosphorus not only causes hot tearing to occur with milder mechanical conditions but also causes enlargement and aggravation of tears which begin at the normal higher temperatures in the presence of severe mechanical conditions. The wider temperature range of hot tearing provided by these elements permits enlargement of the initial cracks in true hot tear fashion. Thus, as is normaly observed, film forming elements produce more numerous and also more severe hot tears.

6. In the light of the foregoing analyses the effect of pouring temperature should be considered highly specific to the type of casting in question.

This conclusion is based on the variety of possible effects from pouring temperature on the nature and extent of the hot spot and contraction conditions resulting from simple modification of gating and risering methods.

# Acknowledgments

The interest and active support of Dr. O. T. Marzke, Superintendent, Metallurgy Division during course of planning and conducting of this investigation is gratefully acknowledged. Contributions of Dr. Blake Loring to the initial experiments and of the assistance of Mr. Thomas Snodgrass, and Mr. William Tilly have been of great value to the investigation and are likewise gratefully acknowledged by the authors.

# References

- 1. H. F. Hall, "The Strength and Ductility of Cast Steel During Cooling from the Liquid State in Sand Molds-Part 2. The Iron and Steel Institute Special Report No. 23, 1938, pp.
- 2. N. B. Gelperin, "Steel Susceptibility to Hot-Tear Formation in Castings,"Transactions, A.F.S., vol. 54, pp. 724-726 (1946).
- 3. K. Singer, "Hot Tears in Steel Castings" Die Giesseri, vol.
- 15, December 1928, pp. 1225-1229.4. J. W. Juppenlatz, "The Iron Carbon System and Its Effect on Castings," The Foundry, vol. 69, Jl. 1951, pp. 104-06. 5. C. J. Lutts and J. P. Hickey, "Special Low Carbon Steel
- for Castings," Transactions, A.F.S., vol. 52, pp. 904-912 (1944). 6. C. W. Briggs and R. A. Gezelius, "Studies on Solidification in Steel Castings II-Free and Hindered Contraction of Cast Steel," Transactions, A.F.S., vol. 42, pp. 449-476 (1934).
- 7. C.W. Briggs and R. A. Gezelius, "Studies on the Solidification and Contraction in Steel Castings—4,—The Free and Hindered Contraction of Alloy Cast Steels," Transactions, A.F.S., vol. 44, pp. 1-31 (1936).

8. W. J. Phillips, "Prevent Hot Tears in Steel Castings,"

- The Foundry, vol. 68, 1940, Jl. pp. 27-28.

  9. H. F. Hall, "The Strength and Ductility of Cast Steel During Cooling from the Liquid State in Sand Molds," The Iron and Steel Institute Special Report No. 15, 1936, pp. 65-93. 10. K. Singer and H. Benek, "Contribution to Hot Tears in
- Steel Castings," Stahl and Eisen, vol. 51, 1931, pp. 61-65.
  11. F. Korber and G. Schitzkowski, "Contribution to Contrac-
- tion in Steel Castings," Stahl and Eisen, vol. 48, 1928, p. 129. 12. C. W. Briggs, NRL Report M-1479 (1938).
- 13. J. M. Middleton and H. T. Protheroe, "Hot Tearing of
- Steel" *JISI*, vol. 168, 1951, pp. 384-398.

  14. J. Vero, "The Hot Shortness of Aluminum Alloys," *The*
- Metal Industry (Lond.), vol. 48 (1936), p. 431 and p. 494.
  15. D. Lees, "The Hot Tearing Tendencies of Aluminum Casting Alloys," The Journal of the Institute of Metals, vol. 72
- 16. H. F. Bishop, F.A. Brandt and W. S. Pellini, "Solidification of Steel Against Sand and Chill Walls," TRANSACTIONS, A.F.S., vol. 59, pp. 435-450 (1951).

#### DISCUSSION

Chairman: R. H. JACOBY, The Key Co., East St. Louis, Ill. Co-Chairman: R. H. FRANK, Bonney-Floyd Co., Columbus,

Recorder: J. R. GOLDSMITH, Crane Co., Chicago.
J. B. CAINE (Written Discussion): The authors are to be congratulated upon another scientific contribution to the science of founding. Their work has at least eliminated the shaky presumption of a brittle range in solid metal just below the solidus and then a sharp increase to very high ductilities. Such behavior is theoretically very improbable.

If the evidence of this paper that hot tearing occurs at the solidus temperature is accepted, a new field opens up in the study of hot tearing, made of solidification as determined from phase diagrams. For example, a number of investigators have

¹ Consultant, Cincinnati, Ohio.

shown that steel with a carbon content of 0.20 to 0.30 per cent hot tears more readily than one with either lower or higher carbon. From the iron-carbon phase diagram it is seen that such steels solidify differently than steels with less than 0.18 per cent carbon, or more than about 0.30 per cent carbon due to the peritectic phase changes. The change from delta to gamma iron should be accompanied by a sharp, almost instantaneous linear contraction that is additive to the normal thermal contraction during cooling. With carbon contents between 0.2 and 0.3 per cent this extra contraction occurs when the metal is close to the solidus and is in, or near, the hot tearing range. Under the 0.18 per cent carbon the delta-gamma transfermation occurs gradually when the metal is solid. Above about 0.30 per cent carbon this transformation and contraction occurs while the steel is still over 50 per cent liquid and the liquid metal can compensate for the contraction.

Could the authors give a few details regarding the amount of restraint imposed on their test casting? Are the restraining plates so arranged that the metal is cast around them? If so, would the restraint be such that the rate of load application could be considered infinite? One last question. Are the reproductions of the radiographs about full size? If they are it should be possible to approximate the amount of contraction for a given drop in temperature by the increase in the width of the hot tear. Is the hot tear study with aluminum more accurate and

informative than with steel?

Messes. Bishop, Ackerlind and Pellini (Written Reply to Mr. Caine): Mr. Caine's remarks are appreciated and the deductions he makes relative to carbon effects on hot tearing are logical.

In the tests for determining the temperature of hot tear origin chills were placed on both faces of the end flanges and four 1½-in. diam restraining bars (2 in the cope and 2 in the drag) were molded between the flanges. The restraining bars fitted tightly against the chills on the inside faces of the flanges, hence the amount of allowed casting contraction was neeligible.

The illustrations of the radiographs were reduced in printing;

the radiographs were originally 6 in. square.

The mechanism of hot tearing is believed to be the same in all metals, ferrous or non-ferrous, and in all cases hot tearing undoubtedly occurs near the solidus temperature. Critical correlation of the precise temperature of hot tearing in relation to the solidus temperature for the different metals is probably not possible due to their variable solidification modes as indicated by the varying slopes and temperature separations of the liquidus and solidus lines in their constitution diagrams. For example the slopes of the liquidus and solidus lines for aluminum-copper alloys are not nearly as steep as those for iron carbon alloys. This indicates, by lever arm relationships, that the temperature range for film stage in Al-Cu alloys is somewhat wider than that of the iron carbon alloys. It is possible that this condition results in hot tearing at temperatures slightly above the solidus for the Al-Cu alloys and very close to solidus for the iron carbon alloys. A longer time interval of film stage (resulting from the wider temperature range of film stage) should result in increased tendency to hot tearing. As Mr. Caine suggests these basic concepts open up a new field of approach to the problem of hot tearing.

U. K. BHATTACHARYA, C. M. ADAMS AND H. F. TAYLOR (Written Discussion): We would like to congratulate Mr. Bishop and his collaborators for developing a new technique for determining the formation of tears in castings by means of radiographs obtained during solidification, and for presenting a theory about tearing in steel on the basis of a critical film of interdendritic

liquid at the near-solidus temperature.

Strain rate is an important factor that should be considered in any tensile test, even a test taken during cooling from the liquid state in a mold. The authors have not mentioned at what rate they pulled the 3-in. diameter tensile bars. Therefore, we do not believe it would be logical to correlate the tensile test data with the tearing temperature unless the rate of testing² is commensurate with the rate at which contraction develops in a casting before it tears. Hall² has shown clearly that strength and ductility of a 0.30 per cent carbon steel varied considerably with different rates of strain; low strain rate (0.046 in. per sec) produced less strength and ductility above 1320 C. Recently, Hamaker² and Wood reported tensile strength of gray iron

during cooling from the liquid state with a strain rate of 0.038  $\pm$  0.001 in. per min.

The authors' specific criticism of Middleton' and Protheroe's work does not seem entirely justified, even though our tests using their bar were not conclusive. We have worked with the British test casting and found that when tearing developed 80 sec after the casting was poured, the metal was so weak that it broke completely within a few seconds. We have shown elsewhere's such casting failures even take place at temperatures well below the solidus.

We believe it would be incorrect to assume a casting can only tear at near solidus temperatures. A simple casting under laboratory conditions may crack in the above range, but it is really difficult to determine at what temperature a complex casting may fail in practice. Therefore, it would be more appropriate to specify a critical temperature below which a cast-

ing will not fail under normal conditions.

The authors contend that since a 1/4-in. zone of the 7-min test bar (Fig. 11) has a ductile fracture, it "must be considered to be devoid of films (liquid) from the fracture appearance" because the temperature at the time of fracture was only 60 F below the solidus. But, non-metallic phases frequently occurring in steel may have melting points well below this temperature. Also, the liquid film theory itself does not explain the cause of relative susceptibility of steels to tearing. It is felt that the tearing tendency depends upon the distribution as well as the existence of a liquid phase below the solidus.

Smith⁶ has shown that the distribution of intergranular liquid is related to the ratio of the energies of the inter-phase and

inter-crystalline boundaries of an alloy (that is,  $\frac{\gamma \alpha \beta}{\gamma \alpha \alpha}$  where  $\gamma \alpha \beta$  is the energy of the interface of the liquid and solid crystals and  $\gamma \alpha \alpha$  the energy of the interface between two solid crystals and  $\frac{\gamma \alpha \beta}{\gamma \alpha \alpha} = \frac{1}{2 \cos \theta}$ , where  $\theta$  is the dihedral angle the liquid phase

makes against two solid grains). Extreme brittleness occurs only if the dihedral angle is zero, because then the liquid will completely "wet" the solid crystals. The above ratio is then equal to or less than 0.50.

As the inter-phase energy increases, the dihedral angle increases and the liquid then exists in more-or-less compact form and may not be as harmful to the ductility. This energy depends upon the composition of the liquid and decreases as the composition approaches that of the solid, and with it the dihedral

angle also decreases.

In his studies on the relative energies of iron/liquid sulphide interfaces in pure iron alloys, Van Vlack⁷ found that the dihedral angle of the liquid iron sulphide decreased to zero at 1300 C. When oxygen was present in the liquid sulphide,  $\theta$  was 15 degrees at the same temperature, because the liquid was less rich in iron. Since mechanical properties at high temperature are functions of  $\theta$ , determinations of  $\theta$  which the liquid phase makes with  $\gamma$ -iron grain boundaries at the tearing temperatures, may provide the answer for the relative tear-resisting properties of steels.

#### REFERENCES

 C. W. Briggs, Metallurgy of Steel Castings, McGraw Hill Book Co., 1946, p. 330.

2. H. Hall, "Strength and Ductility of Cast Steel During Cooling from the Liquid State in Sand Molds," Part I, J. Iron & Steel Inst., Special Report No. 15, 1936, pp. 65-93.

3. J. C. Hamaker and W. Wood, "Influence of Phosphorus on Hot Tear Resistance of Plain and Alloy Gray Iron," A.F.S. Transactions, vol. 60, pp. 501-510 (1952).

4. J. M. Middleton and H. T. Protheroe, "Hot Tearing of Steel," J. Iron & Steel Inst., vol. 168, 1951, pp. 384-398.

5. U. K. Bhattacharya, C. M. Adams and H. F. Taylor, "Hot Tear Formation in Steel Castings," Transactions, A.F.S., vol. 60, pp. 675-680 (1952).

6. C. S. Smith, "Grains, Phases and Interfaces," AIME, 175,

(1948), pp. 15-51.

7. L. H. Van Vlack, "Intergranular Energy of Iron and Some Iron Alloys," AIME, 3, (1951), pp. 251-258.

MESSRS. BISHOP, ACKERLIND AND PELLINI (Reply to Messrs. Bhattacharya, Adams and Taylor): The various questions raised by Messrs. Bhattacharya, Adams and Taylor have been consid-

² Massachusetts Institute of Technology, Cambridge, Mass.

ered in detail in conducting the experiments and in the preparation of the text. The strain rate factor, for example, has been analyzed in detail beyond the simple first approximations presented in the subject comments. Any analysis of this question must include the necessary consideration that the metal at film stage is not homogeneous. Thus, we are not dealing with the simple case of straining of a uniform material but of preferential straining of the liquid film material. The nominal strain rates for the 3-in, diam test bars were 1.0 in. per sec but obviously it is impossible to speak of actual rates of strain in such a case for the significant strain is the separation which occurs in the liquid film itself. The rates of strain quoted in the discussion must be recognized as being approximations which define the rates of movement of the ends of the bar but in nowise represent the rates of strain in the non-strain-sensitive liquid film regions.

It should be recognized that the significance of the load vs solidification state plot lies not in the actual loads measured but in the sharp change of load-bearing capacity with the development of a solid skin. The sharp change noted is in agreement with the basic theory of film formation during the last stages of solidification of metals. The authors do not claim any originality in the concept of film formation for this is an accepted metallurgical fact-the originality of the present paper lies in the demonstrated correlation of hot tearing temperature to the constitution diagram. The 3-in. pull tests merely reinforce the already overwhelming proof obtained by radiographic methods that hot tearing is a film stage problem. As such it opens up new fields of productive rational thinking concerning this

serious foundry problem.

The authors hold an unqualified objection to the past and apparently persisting misconception that hot tearing of steels of normal commercial S and P content can occur to temperatures several hundred degrees below the solidus. In this respect Hall,⁸ Middleton and Protheroe⁴ and Bhattacharya, Adams and Taylor⁵ are mistaken in their interpretation of the temperature records. The common error of inserting a thermocouple directly into a sub-surface of a bar or plate leads to errors in temperature measurement in the order of 150-200 F. This point is discussed in the text of the present paper and does not therefore require further explanation.

The discussors' point regarding distribution of non-metallic constituents if properly analyzed should adequately explain why steels of normal S and P contents should not be expected to hot tear several hundred degrees below the diagram solidus. Simply,

there is not sufficient non-metallic film constituent to bring the average composition of the film to near eutectic ranges which would be required to substantiate the case for abnormally low temperatures of hot tearing.

The case for the theoretical dihedral angle consideration is based on concepts of planar boundaries which have in fact provided the basis for the formulae quoted. It is most difficult to reconcile the complex dendritic form of hot tear surfaces which represent the actual prior surfaces of the liquid films with such simplified concepts. The near zero dihedral angle of low oxygen content iron sulphide liquid in contact with austenite at 1300 C (2372 F) as compared to oxygen rich liquid cannot be translated to signify that similar surface tension differences are developed at near-solidus temperatures. It is entirely possible that reversals or other changes would occur due to the increased temperature. Answers in this respect, if provided, must be based on applicable data.

J. A. RASSENFOSS (Written Discussion): 3 The authors are to be complimented for making a significant contribution to our knowledge of the mechanism of hot tearing. This paper shows evidence of careful work and a thorough approach to the

The study of plain carbon steels at a sulphur content of 0.02 per cent indicates that hot tearing starts at a temperature just below the solidus temperature for the steel in question. When the sulphur and phosphorus contents are increased markedly, the temperature at which hot tears start is decreased. The authors state: "It is interpreted that the strength delay effect of these elements (sulphur and phosphorus) results from the development of liquid film segregates which effectively produce a lowered solidus temperature.

It is presumed that no Type II inclusions were present, since the authors used an aluminum addition which is normally sufficient to avoid the formation of Type II inclusions and no

mention is made of their presence.

Have actual measurements been made of the effect of sulphur and phosphorus on the solidus temperature which demonstrate

that the authors' interpretation is correct?

In short, what evidence do the authors have which demonstrates that the specific effect of sulphur and phosphorus is to extend the film stage and not to have an effect such as an alloving element would give.

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# A STUDY OF THE FORMATION OF NODULAR GRAPHITE

By

#### Fredrik Hurum*

Spherulitic or nodular graphite has long been known to exist and various authors, such as Samuel Epstein, have pointed out that charcoal pig-iron under ordinary foundry conditions sometimes give finely divided nodular graphite, while coke pig irons give long thin flakes.

H. Gröber and H. Hanemann² who investigated the formation of graphite in hyper-eutectic melts prepared from Armco iron and sugar carbon were able to produce spherulitic graphite in melts containing 6.67, 7.40 and 7.70 per cent carbon when chilled.

Such spherulites revealed the characteristic cross under polarized light and had a diameter up to  $6\mu$ .

Messrs. H. Morrogh and W. J. Williams, working able to produce spherulitic graphite when adding cerium to hyper-eutectic melts. When the addition of cerium was followed by an addition of ferrosilicon they could also produce graphite spherulites in hypoeutectic melts.³ Messrs. Millis, Gagnebin and Pilling were able to use magnesium instead of cerium in a dual treatment analogous to that of Morrogh and Williams.⁴

It is known that both cerium and magnesium act as carbide-stabilizers in cast-iron. The genesis of the graphite nodule in cast-iron has been widely discussed, but no theory has apparently found general

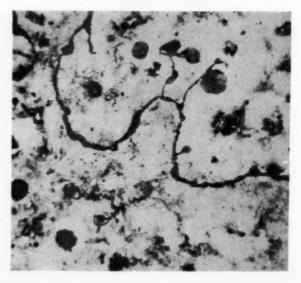
For the present investigation various grades of pig iron were remelted in a 15 kw high-frequency spark-gap induction furnace. Sillimanite crucibles were used for all melting and the charges varied from 2 to 6 lb. Different methods were used to add magnesium and ferrosilicon to the melt. Telescopic test-pieces having diameters of 5,10,20 and 40 mm and test bars having 20 mm diam were cast in graphite and core sand molds. Several melts of synthetic cast iron were also made and treated in various ways. The test bars and telescopic samples were examined under the microscope and usually also subjected to chemical analysis.

An interesting observation was the occurrence of black or dark films in melts having been treated with magnesium and ferrosilicon. Such films do not appear to have been noticed or described by investigators of nodular graphite, although J. E. Rehder is known to have referred to a brownish substance which he believed to be magnesium sulphide.

Figure 1 illustrates such films found in the the 10-mm section of a telescopic test piece of remelted pig iron which contained only traces of sulphur. Sulphur could accordingly not be a part of such films. Ferrite is seen along the borders of the films. M. G. Corson⁵ in his notes on iron-silicon alloys speaks of films and a foam of graphite, while the German expression "Garschaum"—implies that the graphite may have a foam-like appearance.

Of considerable interest was the discovery that graphite nodules existed on these films taking the appearance of sprouts which seemed to detach themselves and drift out into the melt as full-fledged spher-

Fig. 1—Films in nodular cast iron; Melt 39, a sulphur-free electro pig iron with 4.5 C. Photomicrograph made of a 10 mm-diam section. Mag. 600x.



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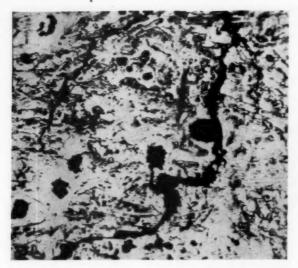


Fig. 2—Films in chill-cast nodular iron. Photomicrograph taken of 5 mm-diam section. Nodules are attached to films. Free cementite and ledeburite are present. Mag. 600x.

ulites. It looked as if these sprouts were drawing nourishment from the films.

These films were also found in considerable size in the chilled section of 5-mm diameter. Figure 2 illustrates such films in white iron of various melts where a ledeburitic structure prevails, sometimes with excess cementite. On, or adjacent to such films are invariably seen the graphite nodules. The films appear to have been formed in a hyper-eutectic melt. Crystals of cementite may be seen to cut into the films at some places, although the ledeburitic matrix has usually vanished along the borders of the films. Of interest is the fact that such films also occur in hypo-eutectic cast irons. The films seemed, on first sight, to be identical in substance with the graphite nodules.

These films were frequently found present in the shape of rings or loops. The matrix inside the loop would frequently have a lighter and brighter appearance than the outside, pointing to a difference in composition. Cementite and even pearlite would sometimes be missing inside the loop, while free cementite might be present in the outside matrix. Figure 3 shows micrographs of some loops and rings. They are evidence of the films being formed on the border between molten strata of different silicon contents. The difference in size between the nodules on each side of the loop is often striking and points to a difference in carbon content. The conclusion to be drawn was that the films were formed on the border between a high silicon, low carbon melt and a liquid of high carbon and low silicon content. The nature of these films is difficult to reconcile with the notion of the crystalline state; it suggested the presence of graphite as a colloidal precipitate.

# The Fractional Diffusion

The author's attention was called to the fact that finely crushed ferrosilicon had a remarkable thickenening effect when added to the molten cast iron. The alloy would rise to the surface and form a fused cover. When this cover was stirred into the metal the thickening, effect became very noticeable even when the metal was very hot. A paste-like substance resulted which would only gradually dissolve in the metal. This substance often remained in the furnace or crucible in substantial quantities if not allowed sufficient time for homogenization before pouring. Thermic considerations alone did not seem to account for this behavior. A microscopic examination of this paste-like substance revealed the presence of massive ferrites which contained interstitial D-type graphite. A sample taken at random was analyzed and found to contain 10.41 per cent silicon.

This observation, it was felt, could explain how the addition of only very small amounts of ferrosilicon to a hypo-eutectic melt would be capable of creating local hyper-eutectic zones such as would be necessary for the production of graphite nodules. The idea that a low carbon melt would render hyper-eutectic zones by the simple empirical addition of a small amount of ferrosilicon did not look as the most convincing explanation.

The microscopic examination of many melts revealed the presence of massive ferrites as remnants of the inocculation with ferrosilicon, provided that the metal had not been kept too long or at too high a temperature before pouring.

Figure 4 shows silico-ferrite in nodular cast-iron. The presence of graphite on the border of the massive ferrite is significant both with respect to appearance and quantity. The author believes that only the principle of what may be termed "fractional diffusion" may explain the efficiency with which a small amount of ferrosilicon is capable of producing substantial zones of hyper-eutectic metal in a hypo-eutectic melt.

Here it may be well to sum up the brief features of

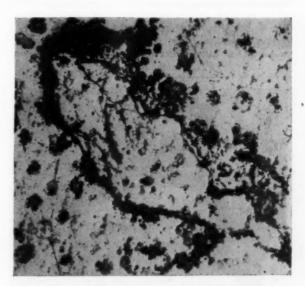


Fig. 3—Films seen as loops in nodular iron. Inside loop seems to have a structure different from that of the outside, indicating a difference in composition. Melt 45 was chill-cast and annealed, 40 mm-diam section. Mag. 180x.

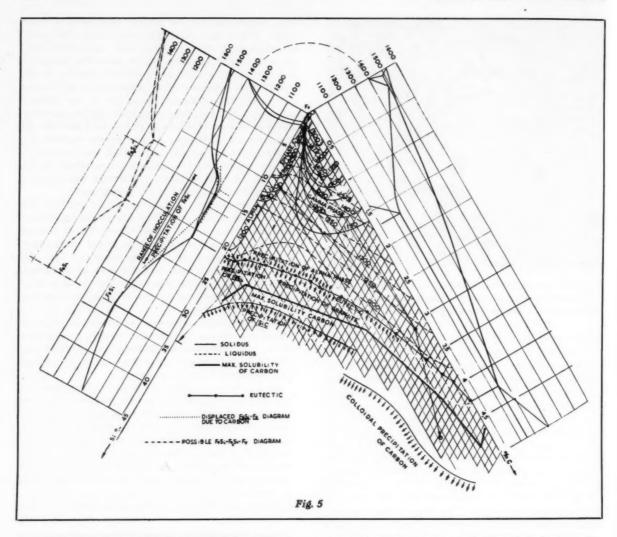
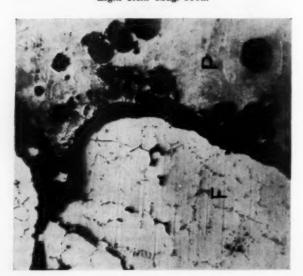


Fig. 4—Massive ferrite precipitated in nodular iron; F, ferrite, P, pearlite. Heavy graphite precipitations seen between ferrite and pearlite result of fractional diffusion.

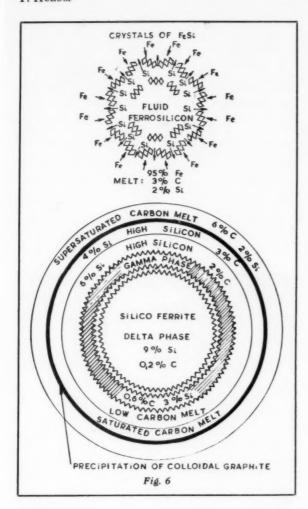
Light etch. Mag. 600x.



the Fe-Si system, the figures being taken from an investigation of Haughton and Becker.⁶

Starting with a 75 per cent ferrosilicon alloy, this will consist of the silicon phase and the zeta-phase corresponding to the compound Fe₂Si₅ with a small amount of silicon in solid solution. The silicon phase and the zeta-phase will form a eutectic at 1208 C (2205 F) and with about 58 per cent silicon. As iron now is added to the system the silicon-phase will disappear and the zeta-phase alone exist. The compound Fe₂Si₅ will melt at 1205 C (2200 F). As more iron is added the compound Fe₂Si₅ will now hold a little iron in solid solution and form another eutectic point, this time with the compound FeSi at 1212 C (2214 F) and with about 51 per cent silicon. Here the zeta-phase vanishes and the epsilon-phase comes into existence corresponding to the compound FeSi which melts at 1410 C (2570 F) and holds 33 per cent silicon.

As will be seen from the above the liquefaction of the ferrosilicon does not present difficulties as far as the temperature goes. As however still more iron is added and the silicon content in the Fe-Si system falls towards 33 per cent it will be necessary to resort



to a temperature above 1400 C (2550 F) to secure liquefaction. As still more iron is added to the system and the silicon content is reduced below 33 per cent, the liquidus curve will again fall until another eutectic point is reached corresponding to about 21 per cent silicon and at 1195 C (2183 F). At this point the epsilon- and the alpha-delta-phases meet.

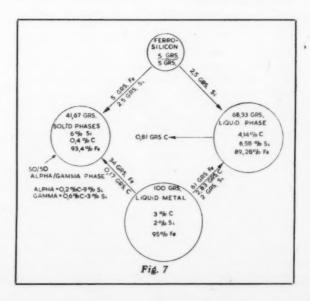
To gain some insight in the Fe-Si-C system the author drew up the isotherms from the available liquidus and solidus points. While the Fe-Fe₃C side may be regarded as thoroughly investigated, the same cannot be said of the Fe-Si side. Very little information is available to demonstrate the effect of carbon. The work of such investigators as Satoh, Honda and Murakami, Kriz and Poboril, Jass and Hanemann, Wüst and Petersen, Gontermann and Hanson were consulted. Figure 5 gives the isotherms. The line of maximum carbon solubility based on the observations of Honda and Murakami. The eutectic line is taken from the work of Jass and Hanemann.⁷

As will be seen from Fig. 5 the alpha-delta-phase is likely to hold far more silicon but less carbon than the gamma-phase when the two phases co-exist. This fact appears frequently to be ignored judging from misleading diagrams used to demonstrate the influence of silicon on the Fe-Fe₃C system. There appears

little reason for assuming that silicon is evenly distributed in the alpha- and gamma-phases over a wide range of temperature and even in the liquid in the presence of the two solid phases. On the contrary it may be expected that a low-silicon austenite may coexist with a high-silicon ferrite and thus serve to withdraw important amounts of iron from the melt under the mechanism of a fractional diffusion.

Figure 6 depicts what may be conceived to happen when ferrosilicon is added to a hypo-eutectic melt holding 3 per cent carbon and 2 per cent silicon. The alloy will first become liquefied and absorb iron from the surrounding melt thereby swelling up considerably under precipitation of alpha-delta iron (silicoferrite) and gamma-iron together with some Fe-Si crystals. Both the silico-ferrite and the FeSi phase would mean a serious impediment to the diffusion of carbon. (So may the formation of SiC prove to be as will be seen later). The formation of the fairly stable FeSi phase may even prove an obstacle to the migration of silicon. The same may be true of the SiC precipitate and to a lesser degree of gamma-phase which has a restricted solubility for silicon. Of the three elements iron is the only one free to travel through all the phases with the exception of SiC. Even the epsilon phase is granted to have a certain solubility for iron.

Figure 7 presents the quantitative results of the fractional diffusion such as might be imagined when one part of ferrosilicon (50 Fe-50 Si) is added to ten parts of cast iron. Two-fifths of the total weight are seen to be precipitated as a low carbon mixture of ferrite and austenite, while the remaining three-fifths would consist of a hyper-eutectic melt capable of liberating an appreciable amount of graphite. Such results would of course not be in conformity with the statics of the phase rule and must be regarded as only transitory allowing the selective diffusion of the three elements involved. As a matter of fact the inocculation with ferrosilicon is in itself only of transitory effect. When the alpha- and the gamma-precipitates are redissolved in the melt, the nodular



graphite will also begin to dissolve.

No claim to accuracy is made for Figs. 6 and 7. Their purpose is only to give an idea of the extent to which the double precipitation of silico-ferrite and austenite may draw iron away from the surrounding melt and enrich it with carbon.

The precipitation of alpha-ferrite containing 9 per cent silicon and 0.2 per cent carbon may truly be expected from a melt containing in the neighborhood of 10 per cent silicon and somewhat less than 0.5 per cent carbon when the temperature is around 1300 C (2375 F). Succeeding this precipitation, or simultaneously with it, the formation of gamma-iron holding 3 per cent silicon and 0.6 per cent carbon may be anticipated. The only requisite to this result would be the local creation of a liquid holding 10 per cent silicon and less than 0.5 per cent carbon-a condition that easily can be met if admitting the comparative ease with which the iron will diffuse. By way of example it has been assumed that one-half of the silicon in the alloy would diffuse into the melt, moving in opposite direction to the iron. Carbon will of course also diffuse from higher to lower concentration, but this migration is lagging and gains importance only as the precipitated phases begin redissolving.

The author does not wish, of course, to suggest any such procedure as adding one part of alloy to ten parts of metal, but such an exaggerated proportion would nevertheless exert itself locally whenever the inocculant is added to the metal.

#### The X-Constituent

The fractures of many test bars were examined. Successfully treated metal would usually not reveal much of interest, except that graphite nodules would be present in every conceivable shape and size and often be linked together in a number of two or more. The theory advanced that the genesis of the graphite nodule is a certain nuclei on which the graphite would crystallize in a certain way seemed to be only a part of the picture.

Test bars have a tendency to break where imperfections are found and it was a hope that such defects would have an explanation to offer. In some of the fractures black spots or patches could in fact be seen, indicating an uneven distribution of carbon likely to be due to the metal having been poured before the necessary degree of homogeneity had been acquired. See Fig. 8. Under the microscope it could be seen that graphite films sometimes would cut through a considerable portion of the polished section while heavy conglomerations of carbon often would accompany them. Sometimes cavities were discovered and the path of fracture could often be identified with the location of a film.

The etching of the surface in such sections where graphite films and carbon conglomerations occur is often rendered difficult because the reagent will be absorbed in the pores and cracks and it will subsequently discolor interesting details. This difficulty is likely to explain why an important detail apparently has escaped the notice of investigators of nodular cast iron. This important detail for the sake of convenience will be referred to as the X-constituent. It has

an appearance which is different from any constituent of cast iron and steel, although it may sometimes be difficult to discern from pearlite when etched.

It will readily be discovered, however, whenever present, in the carefully polished unetched section and it shows up well when lightly rubbed with cotton moistened with alcohol or acetone until completely dry. It appears in both heat-treated and in the as-cast samples. This X-constituent may be regarded as a pregnant liquid capable of producing enormous numbers of graphite nodules when distributed into the hyper-eutectic melt surrounding it. In this X-constituent is found the genesis of the nodule. This constituent is hard and brittle and several cracks have been found to exist in it.

Figures 9, 10, 11 are microstructures of X-constituent as it appears in various melts. It is seen to consist of myriads of small particles in a metallic

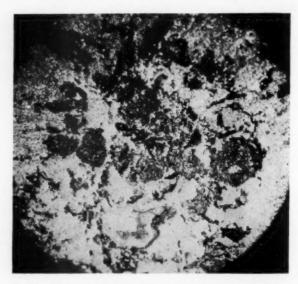


Fig. 8—Defective fracture due to heavy film formations.

Metal not sufficiently homogenized. Mag. 6.5x.

matrix. The graphite is precipitated on these particles, but apparently in different ways.

The fine particles were at first supposed to be pure graphite of a colloidal origin which coagulated and took the shape of films and spherulites, but a further examination made it appear likely that they were of a foreign substance.

Figure 11 gives evidence of graphite being precipitated on a foreign substance. This substance could be traced back to the particles of the X-constituent. The appearance of these particles are, however, usually obscured by the adhesion of a slimy deposit which gives them a dark color and soft contours. They appear like minute clots. The movement of these particles are easy to follow and study.

Figure 9 indicates that the fine particles are not evenly distributed in the X-constituent. They are seen to form concentric bands in an arrangement which very much resembles the pattern of annuals in a tree, and which grow in thickness as they approach the circumference. The size of the particles also dif-

fers. The suspicion that carbon as it diffuses slowly from the outside towards the interior of the liquid X-constituent is instrumental to the formation of the particles as well as their destruction seems substantiated.

The X-constituent appears to contain a high amount of silicon which however may vary. This silicon when diffused into the adjacent melt, is likely to produce a ferritic matrix around the X-constituent. Figure 10 clearly shows ferrite which the outside pearlite may be seen to encroach upon (x100, slightly etched). The X-constituent may often be seen by the naked eye as a bright spot in the etched section of a specimen.

Figure 12 shows how the X-constituent with its multitudes of fine particles will create nodules when disseminated into the surrounding high carbon melt. Fig. 13 shows how the dissemination of nodules proceeds. It was the author's desire to identify the fine particles of the X-constituent. At least some of them appeared very hard and left stripes in the surface when polishing. Several possibilities had to be considered. Would they be carbides or silicides or would both of them be present? New compounds would also have to be reckoned with. The X-constituent resembled finely spheroidized cementite and this compound may be considered first.

It has been a general belief that iron carbides cannot exist in high silicon cast iron, but such investigators as Kriz and Poboril⁸ have reported the presence of carbides in chill-cast specimens containing about 10 per cent silicon and 0.22 per cent to 1.35 per cent carbon. More recently J. E. Hurst and R. V. Riley maintained that iron carbides were present in a 15 per cent silicon iron when chill-cast.⁹ D. Marles¹⁰ reported two different carbide constituents, one of which formed in high silicon melts and would not be heat tinted while it decomposed on annealing

Fig. 9—Photomicrographs showing X-constituent coalescing (right) into a band which is the origin of the films. The X-particles assume a slimy appearance as they approach the outer edge of the X-constituent. Mag.—Left, 100x; Right, 2000x.

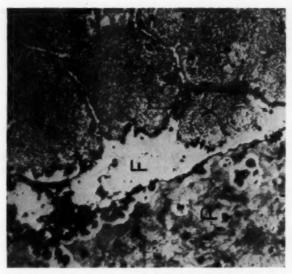
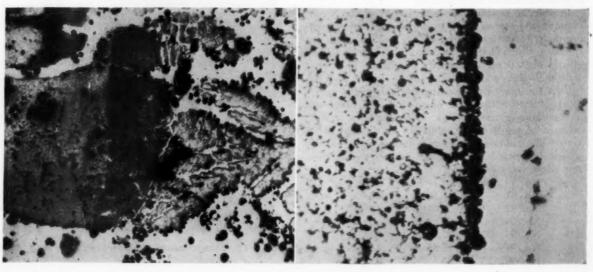


Fig. 10—X-constituent with the formation of films. A loop forms around the X-constituent. Ferrite band, F, present between pearlitic matrix, P, and X-constituent, X. Light etch. Mag. 180x.

at 700 C (1290 F). E. D. Harry¹¹ pointed likewise to a duplex structure of the carbides in a 4 per cent silicon iron.

Unfortunately no accurate knowledge is available with respect to double carbides of iron and silicon, nor do we possess any knowledge of compounds of iron silicides with iron carbides or silicon carbides. The possibility of such compounds being formed during the inocculation with ferrosilicon should be borne in mind. The action of magnesium must also be considered. It may form carbides or double carbides.

Magnesium may be expected to be present in the X-constituent. The compound Mg₂Si melts at 1102 C (2015 F) and is soluble in silicon. When iron enters the system Mg₂Si-Si, the solubility of this silicide is markedly reduced and it will be set free in a fine condition by iron under the formation of FeSi. Mg₂Si is likely to be present in the X-constituent. Nodular



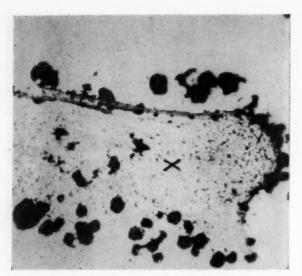


Fig. 11—X-particles coalesced into bands. Slimy fluid is seen as black substance between X-particles. Nodules are seen to form on this fluid (Z-fluid). Mag. 600x.

cast iron will produce a marked odor of carbides and silanes when ground in the presence of water.

The presence of silicon carbide in cast iron does not appear to be recorded. Silicon carbide is known and used as a graphitizer. It has been suggested that this compound takes part in the transformation of cementite to graphite, but no theory has found acceptance.

The presence of the compound FeSi—the epsilon phase—of the Fe-Si system must be anticipated. As will be seen from Fig. 5 the compound FeSi begins to form in a melt with 21 per cent Si. When carbon enters the mother liquid the eutectic point moves to the right and was found at 17 per cent Si with 0.5 per cent C by Jass and Hanemann. Accordingly the precipitation of FeSi could be effected in a melt holding from 17 to 33 per cent Si if carbon were allowed to enter the mother liquid. This precipitation would take the appearance of very fine particles if sufficient nuclei were present.

In order to establish the exact nature of the X-constituent, the author tried to synthesize it by preparing melts from Armco iron with varying amounts of silicon to which charcoal iron was subsequently added. Silicon of 98.9 per cent purity was used. The alloys were melted in a sillimanite crucible in the high-frequency furnace. After pouring a sample of the binary alloy, a small addition of magnesium was made and a new sample was poured. Thereupon a certain amount of pig iron was added in an attempt of inverted inocculation. As soon as the pig iron was dissolved, the melt was stirred and poured. The following alloys were examined:

 No. 90 holding
 10.74 Si and 0.58 C

 No. 91
 " 15.15 Si " 0.38 C

 No. 92
 " 21.04 Si " 0.28 C

 No. 93
 " 22.85 Si " 0.38 C

 No. 94
 " 27.32 Si " 0.11 C

 No. 95
 " 25.86 Si " 0.32 C

 No. 96
 " 25.8 Si " 0.22 C

The addition of magnesium was not seen to affect the microstructure of the binary alloys.

Alloy 90 and 91 were found to consist of silicoferrite (alpha-phase) with eutectic. In alloy 90 fine interstitial D-type graphite was clearly visible while in alloy 91 the graphite appeared as extremely small flakes and in small quantities. The other alloys contained the epsilon phase with eutectic, but without visible signs of graphite.

It was found, however, that well-defined hexagonal crystals of brown color would co-exist with the epsilon phase. Their size would vary considerably and their

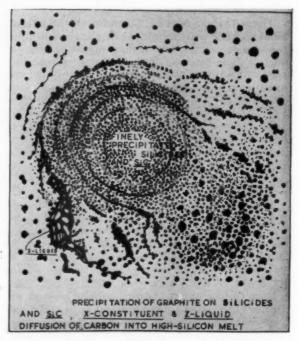


Fig. 12

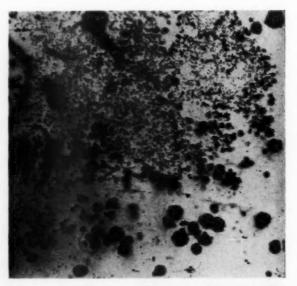


Fig. 13—Dissemination of X-particles into melt. Mag. 300x.

distribution was not always uniform. They were seen to be present in the epsilon-phase and in the eutectic. Under high magnification (x1250) they could also be discovered in alloy 91, but in small quantities. Even in alloy 90 a few of these crystals could be discovered under high magnification.

After an extraction with fuming hydrofluoric acid followed by ignition, the insoluble residues of the alloys were subjected to an X-ray examination. The Debye-Scherrer diagram gave sharp lines corresponding to SiC (II) with a=3.095 Å and c=15.17 Å and SiC (III) with a=3.095 Å and c=10.10 Å. The hexagonal modification SiC (I) was found to be present in minor quantities while the cubic modification could not be traced.

Only the alloys 90, 91 and 93 had dissolved the carbon in the pig iron to the full extent. Alloy 96 was on the contrary found to contain only 0.22 per cent against an amount of carbon in the pig corresponding to 1.6 per cent. The erratic behavior of the carbon may be due to the formation of an amorphous substance in which carborundum is present and which does not dissolve in the metal. This substance seems to form a crust on the crucible and may be found as inclusions in the metal.

Figure 14 shows silicon carbide in the alloys mentioned above. In alloy 90 two small crystals may be seen. (1250x)

A re-examination of a number of test-bars which now followed, proved that silicon carbides were present both inside and outside the X-constituent. The carefully treated insoluble residues from some of these samples were X-rayed and the Debye-Scherrer diagrams clearly indicated the presence of SiC (I), SiC (II) and SiC (III) with the exception of one sample which revealed the presence of only magnetite (Fe₃O₄ cubic, a = 8.37 Å)

Figure 15 shows SiC-crystals as found in nodular cast iron. At high magnification hexagonal crystals

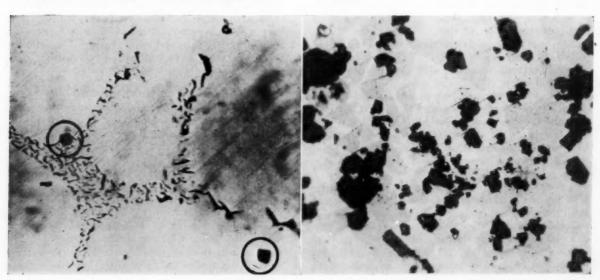
Fig. 14—Silicon carbides in iron-silicon alloys. The silicon carbides were produced by adding a pig iron to the iron-silicon alloys. Mag.—Left, 1250x; Right, 600x.

resembling SiC may be seen among the particles of the X-constituent and there is reason to believe that silicon carbide plays a part in the formation of nodular graphite. There is, however, hardly any justification for drawing the conclusion that the fine particles consist exclusively of SiC. As already pointed out there is good reason for presuming that silicides are present. It may suffice to point out that SiC coexists with FeSi in the alloys prepared and there appears no reason why they should not co-exist in the X-constituent.

It cannot be denied that the appearance of the fine particles in the X-constituent suggested the presence of a graphite-eutectic, nor can it be denied that some of the particles resemble spheroidized cementite, but the author holds the opinion that the nature of the X-constituent is that of a high-silicon liquid phase which has been injected into a high carbon melt. The contour of the X-constituent corresponds to those of a liquid phase which is in movement in another liquid phase. The fine particles of the X-constituent should be attributed to the effect of a sudden penetration of carbon into a high silicon fluid.

This review would not be complete without considering also the presence of Fe₃Si. This compound has been presumed to exist, but unfortunately its existence has not been proven. The liquidus and solidus lines of the Fe-Si system are seen to meet at a point corresponding to the composition Fe₃Si and this irregularity, although vague, has been attributed to the existence of such a compound. Until the matter has been further investigated, it would be wise to allow for its presence, particularly so in view of the possibility that magnesium silicide may form compounds with iron silicides.

Presence of Fe₃Si has been indicated in Fig. 5. If present as fine particles in the X-constituent, the importance of this compound may readily be realized. Also indicated in Fig. 5 is the precipitation of SiC along a line whose position must be regarded as purely speculative in lack of a more thorough investigation.



The author did not succeed in synthesizing an alloy having the appearance of the X-constituent, but the failure may be due to a lack of proper nucleation.

# Colloidal Graphite

As previously mentioned, the discovered graphite films suggested a colloidal origin. A colloidal precipitation may be expected in such complex melts where the orderly crystallization of a component is hindered by the presence of foreign molecules. Factors which otherwise influence the crystallization of a component are the number of nuclei present, the viscosity of the melt, the rate of cooling and the de-

gree of supersaturation.

The border line between two liquid phases, or rather between fluids having a different composition, may naturally be expected to furnish a certain complexity favorable to a colloidal precipitation. This would be the case when a liquid phase supersaturated with carbon meets another liquid phase high in silicon. The supersaturation has been attributed to a fractional diffusion, while the presence of magnesium is probably instrumental to a stabilization of the carbides necessary to achieve this supersaturation. When the X-constituent with its contents of silicon now enters the field with the appearance of the X-constituent, it will break down this stabilizing effect.

If double carbides of iron and magnesium are broken down by the action of silicon, the precipitation of graphite may become too violent to take the shape of an orderly crystallization. A microscopic study of the zones surrounding the X-constituent readily unveils an intense and disorderly precipitation of flocculent graphite. See Figs. 9, 10 and 11. In Fig. 5, the precipitation of colloidal graphite in a hyper-eutectic melt has been marked by drawing a curved line well beyond the eutectic. While the position of such a line-if any-could hardly be ascertained, there seems some justification for its recognition. On Fig. 6 is indicated the presence of a zone where a colloidal precipitation of graphite may be conceived to occur. Here again it must be pointed out that this zone was merely drawn to stake the nature of the problem. It could naturally be imagined to be a very wide one. However, as will be seen, it is the decomposition of the particles of the X-constituent which has a first claim on the creation of colloidal graphite.

A colloidal precipitation may also be caused by the introduction of colloidal particles through the medium of another fluid. An action parallel to that of nucleation may be anticipated. Colloidal particles have the Brownian movement which is caused by the unbalanced impact of molecules. When the particles have acquired a certain size they no longer respond to the Brownian movement and coagulation takes

place.

A coagulation may also take place on the border of a liquid in which colloidal particles are present or in the process of coagulation when such a liquid is injected in another liquid containing colloidal particles or capable of producing such.

For reasons already explained, the X-constituent was at first believed to consist of an iron-silicon solvent with pure graphite as the dispersed phase. This graphite was thought to coagulate and form dots and films. As the particles of the X-constituent, however, were found to be of a different substance than graphite, it seemed justified to explain the formation of graphite films by the simple process of metamorphosis. If the silicon carbides or the silicides would coalesce into bands as they in fact appear to do and the graphite would crystallize on them, the plain result would be that graphite would remain in the shape of bands or films even after the silicides or the silicon carbides had been dissolved or decomposed by the

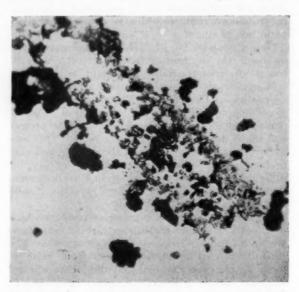


Fig. 15—Silicon carbides in nodular cast iron. Unetched. Mag. 600x.

surrounding melt. Graphite and silicon carbide both crystallize according to the hexagonal system and such a metamorphosis would only be reasonable to expect. The writer does believe that some nodules are formed by direct crystallization of graphite on silicon carbide, but apparently not all. What actually happens, the particles of the X-constituent—the silicides and silicon carbides—begin to decompose as they are disseminated into a melt with a lower silicon content. This decomposition results in a colloidal precipitation of graphite which creates a slimy dark fluid around the decomposing particles.

# The Z- Liquid

For the purpose of discussion, the slimy dark fluid formed around the decomposing particles will be called the Z-fluid. It is likely to consist of iron and FeSi together with silicon carbides and a generous amount of colloidal graphite. The presense of Mg₂Si could be expected. It is probably due to this colloidal graphite that this fluid looks dark and does not behave as an ordinary melt giving a true eutectic of silico-ferrite and FeSi. Silico-ferrite is ejected from this fluid as an emulsoid pointing to incomplete miscibility. Here in particular the complexity of the melt is such that an orderly crystallization is hindered. This applies not only to the graphite and the

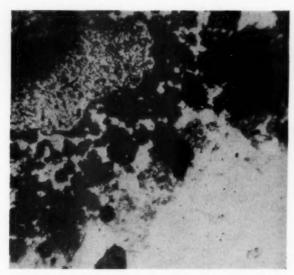


Fig. 16—Z-fluid formed around the X-constituent. Z-fluid forms on borders of X-constituent as the X-particles coalesce. Z-fluid appears as a separate phase. Graphite nodules form on its borders. The Z-fluid appears to hold several components as white droplets along with at least two different kinds of crystals and an amorphous matrix. Unetched. Mag. 600x.

ferrite, but also to the carborundum. It is seen to form amorphous substances with a conchoidal fracture and pieces of such a substance will often break loose and form cavities in the polished surface. This Z-liquid may sometimes be formed in considerable quantities when submitting a high carbon pig iron to the dual treatment with magnesium and ferrosilicon.

Figures 16, 17 and 18 show details of the formation and behavior of this Z-liquid.

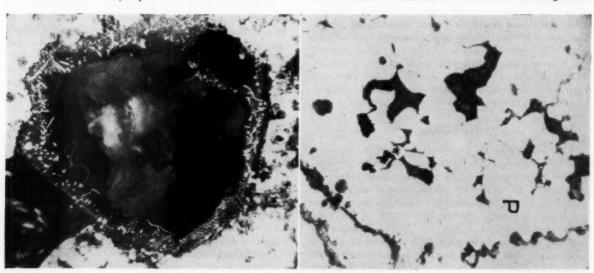
The Z-liquid is seen to form in the X-constituent and gain in size as it approaches the border. Here it may attain a massive appearance. The nature of its formation appears to be that of a coagulation of minute details of slimy liquid created around the fine

particles of the X-constituent by their decomposition. Figure 16 shows how the fine particles of the X-constituent are transformed and flow together into a massive fluid which appears to have only a limited solubility for iron. This Z-liquid is seen to form films which are more or less broken up by the movement in the melt, but which also under circumstances may form heavy conglomerations. Figure 17 shows the appearance of such conglomerations with the characteristic white droplets which are believed to be silico-ferrite. On etching they appear like ferrite and in only a few cases has the writer been able to find pearlite in them which would point to a high silicon content.

It is significant that the Z-liquid has large quantities of graphite on its borders. If the graphite is present as a dispersed phase in the Z-liquid, it would be natural to expect that the coagulation of this element would start on the border. Here the emulsoid particles of silico-ferrite would first coagulate and pass into the surrounding melt with an ensuing concentration of graphite as a result. This concentration would in turn promote the coagulation of graphite, followed by its crystallization. As the Z-fluid in this way becomes enriched with SiC in its interior, crystals of SiC may form or it may solidify as an amorphous substance. Under the microscope may be seen that crystals of SiC are associated with the Z-liquid ,although the appearance of such crystals is usually obscured by an amorphous matrix. Some black crystals having the distinct spearhead form of the tetrahedral FeSi-crystals have clearly been noticed but they are not visible on the photomicrograph.

Figure 18 shows the Z-liquid as it appears in the interior of a loop while the nodules are an unmistakable part of the exterior. In the film itself is seen a black ribbon which is the amorphous remnants of

Fig. 17—Z-fluid as interstitial filling (Left). Dendrites have pearlitic matrix. Unetched. Mag. 200x Right, Center amorphous substance, outer zone white droplets, while nodules are seen on the borders. The white droplets are believed to be a silico-ferrite, while the brownish crystals are believed to be silicon carbides. Unetched. Mag. 600x.



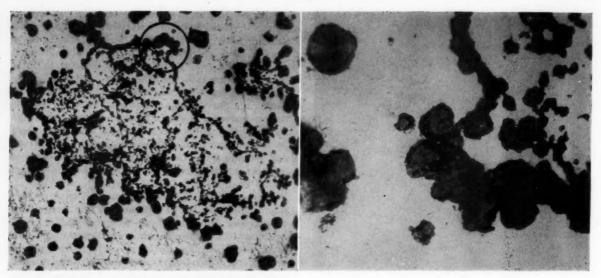


Fig. 18—Nodules seen as precipitate on Z-fluid. Film with black amorphous core and nodules. The black core is remnant of Z-fluid on which graphite was precipitated. Z-fluid inside loop. Nodules on outside. Black amorphous core seen in film-loop. Unetched. Mag.—(Left), 100x; (Right), 600x.

the Z-liquid. Outside the loop are seen nodules with the fine particles of the X-constituent which once filled the loop.

Also seen is a film with nodules and a black ribbon as core as well as the remnants of a film in a melt held too long before pouring. The nodules and the graphite alongside the amorphous core have been dissolved and flakes have later been formed.

The Z-Huid appears to have a surface energy not much different from that of the metal surrounding it. In its original state it does not appear to have any pronounced tendency to form droplets of spherical shape in the melt. This equality of surface tension would be favorable to coagulation on its borders.

The Z-liquid may often be found as interstitial filling among dendritic formations of austenite if present in a hypo-eutectic metal. Due to its incomplete miscibility with the liquid silico-ferrite phase, the iron will not readily penetrate through it and it may be seen to form uninterrupted films around the X-constituent which will shield it over a certain length of time. When the film is finally ruptured the inside will be disseminated into the melt. Crystallization in this complex Z-liquid is apparently a slow one.

# The Nodules

The ideal picture of creating nodular graphite could be imagined as a method by which the particles of the X-constituent could instantly and evenly be distributed in a slightly hypereutectic melt. The nodules would then form by crystallization on these particles. To some extent this dream of an ideal nucleation appears to come true and investigators of nodular graphite have presented pictures of perfect spherulites with a single nucleus in the center.

In most cases, however, the nodules disclose a core

of black amorphous substance and generally a mottled appearance which seems to be caused by the presence of the fine particles of the X-constituent. The mottled strata of the nodules sometimes present themselves as a continuation of the coalesced bands of the X-particles found in the adjacent melt. Figure 19 shows X-particles distributed between nodules and melt.

As already explained the Z-liquid should be defined as an iron-silicon melt in which substantial amounts of graphite are present as a dispersed phase, i.e. in the colloidal state. The presence of silicon carbide must naturally also be expected in this melt. The Z-liquid is created by the decomposition of the X-particles due to their transfer into a melt having a lower content of silicon and a higher content of carbon than the X-constituent. The Z-liquid will accordingly appear as microscopic clots embracing the remnants of an X-particle. These clots have a marked tendency to coalesce. It appears that the nodules are formed on these clots as a colloidal precipitate. The surface of the Z-liquid appears to favor such a precipitation.

While the direct crystallization of graphite on the surface of the Z-liquid cannot be denied and undoubtedly occurs, it seems as if coagulation of graphite is the first step in the formation of nodules, followed by a crystallization as a secondary and somewhat slower occurence. The flocculent deposit of graphite on the surface of the Z-liquid would naturally contain entrapped metal which would segregate, thereby causing concentrations of graphite. As this concentration proceeds, crystallization will begin. This concentration of graphite is likely to be facilitated at the surface where the entrapped metal will join the surrounding metal and here crystallization will begin and proceed towards the interior.

The rate of crystallization may naturally be expected to proceed much faster through the flocculent graphite than would be the case in the Z-liquid or on its borders where the complexity of the fluid would present obstacles to the regimented crystallization. The Z-liquid would be the last to crystallize. The appearance of the nodules would not only be affected

by the points where crystallization starts, but also by the difference in speed by which the different centers

of crystallization propagate.

The nodules often present a striking likeness in structure to that of an ingot where solidification begins at the outer walls. The presence of white particles in the interior, which so often may be noticed, would be the natural result of the metal segregating towards the interior as graphite crystallization proceeds.

While crystallization of graphite would be likely to begin at the surface and at different points more or less simultaneously, it could also begin in the center of the flocculent deposit, or at any point wherever the graphite concentration would favor the orderly crystallization and its speed of propagation.

When crystallization begins at different places, the nodular graphite will take the appearance of this multiplicity. This is evident from Fig. 19.

Although the Z-liquid in its original state does not reveal a surface-energy much different to that of the surrounding melt, the viscosity and character of it will change as the crystallization proceeds and it may take a compact form. When the Z-liquid is broken up by the movement in the melt and the graphite is precipitated on the shattered fragments of the films, the nodules will take the peculiar appearance as shown in Fig. 19. The final shape of the nodule will be influenced by the condition of the metal and its solidification. If the metal is hypo-cutectic, the nodules will begin to dissolve as they move from the carbon-rich strata into the melt.

#### The Dissemination

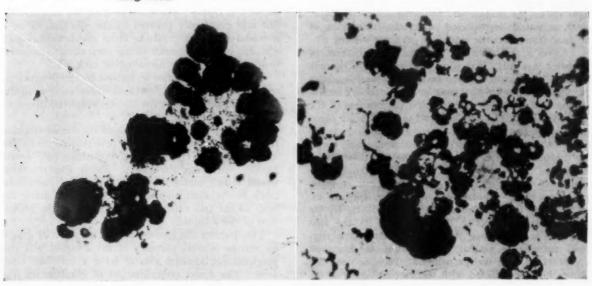
The successful production of nodular cast iron will depend on a thorough mixing of the metal in order to distribute the X-particles properly. The X-con-

Fig. 19—Nodules. The X-particles give nodules mottled appearance. They appear as black spots and cores surrounded by crystalline graphite. Left, unetched, Mag. 740x. Right, Nodules formed on shattered tragments of Z-fluid. Mag. 600x.

stituent is in its origin a low carbon, high silicon fluid around which a high carbon liquid strata has been created by the mechanism of fractional diffusion. As the small particles of the X-constituent, surrounded by Z-fluid, now pass through this high carbon liquid strata, they acquire a heavy flocculent deposit of graphite, whereupon the crystallization of this coagulated graphite begins which may or may not keep pace with the rate of deposit. If the nodules have been sufficiently "fattened" in this high carbon strata, they may be able to survive for some time as they pass into the hypo-eutectic metal. During the period in which the nodules exist in the hypo-eutectic melt, they will be subjected to a "washing." Fragments of films, protruding edges and such shapes having a big surface in relation to the mass, will first dissolve in the melt and the nodules will gradually acquire a cleaner and more rounded appearance.

The solidification in a hypo-eutectic cast iron begins with the precipitation of the gamma-phase which proceeds until the melt has acquired a eutectic composition. The presence of graphite nodules will act as nucleation for the crystallization of graphite and prevent an undercooling of the eutectic melt. Their presence may furthermore be expected to move the eutectic point to the left and raise its temperature according to the established ideas of a dual system of iron-graphite. It is evident that under such circumstances the precipitation of primary austenite will not reach an extent where it appears as massive dendrites. The nodules on the contrary will shape the pattern of solidification.

When the graphite-eutectic point is reached, graphite begins to crystalize on the surface of the nodules and make them grow. The surrounding melt automatically becomes saturated with metal and the precipitation of the austenite-phase will follow an orientation radial to the nodule. A rapidly cooled nodular cast iron will be seen to have a network of cementite grouped concentrically around the nodules, and this may be taken as evidence that the solidification in this case has started around the nodules and pro-



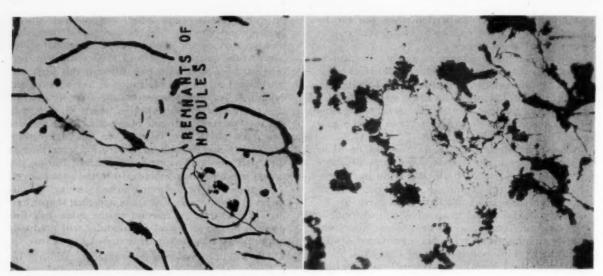


Fig. 20—(Left), Remnants of film in a melt held too long before pouring. Remnants consist probably of amorphous core in film. Mag. 600x. (Right), Z-fluid as interdendritic filling. Mag. 200x.

ceeded radially, leaving finally an interstitial liquid which will solidify white.

Figure 1 shows a rapidly cooled nodular iron with some free cementite which is located at a maximum distance from the nodules in what may be discerned as a concentric arrangement.

A successfully treated nodular iron should have a sufficient number of nodules in proper distribution to safeguard the assimilation of all carbon in excess of that contained in the pearlite. If the distance between the nodules is short enough, the carbon concentration gradient between the nodule and the melt will be sufficiently steep to prevent the formation of free cementite, respectively graphite flakes. Neither will any segregate graphite appear. If the nodules present are too few, cementite and flakes may form in an arrangement concentric to the nodule. If the cooling is slow enough, ferrite will be seen to precipitate on the nodule and assume a spherical shape, such as would be expected under the conception of a dual iron-cementite and iron-graphite system.

Figures 8 and 20 show how the dissemination of nodules proceeds. It is seen to begin at the borders of the X-constituent where myriads of small nodules are carried away by what may be explained as a laminar movement taking place around the borders of the X-constituent. To begin with these myriads are often seen to be surrounded by films which will gradually dissolve in the metal and give the inside free access to the melt. In the company of the nodules may be seen fine particles of the X-constituent completely deprived of any graphite precipitate, but carrying more or less Z-liquid. This occurrence may be explained by the lack of homogeneity existing in the metal surrounding the X-constituent. As the first X-particles move into the high carbon strata, they will provoke a violent precipitation of flocculent graphite on their surface, while at the same time this strata becomes diluted with the mother-liquid of the

X-particles which is very low in carbon and high in silicon. The X-particles flowing in the wake of the first ones will accordingly not find excess carbon to precipitate on them. The presence of such X-particles may be found almost everywhere in the melt if the conditions for their survival are favorable. They may best be seen in the unetched specimens and probably represent micro-porosity in nodular castiron. If more closely studied under high magnification, these particles reveal the appearance of the "slimy" Z-liquid which may account for this apparent stability.

#### Conclusions

The X-particles are likely to consist of carbides and silicides precipitated in a high silicon fluid by the action of carbon and a nucleating effect which may or may not be due to magnesium. The X-particles are seen to decompose when the composition of the mother liquid is changed and they form a slimy fluid described as the Z-liquid which apparently contains graphite as a dispersed phase. The presence of silicon carbides has been established. They may be formed by direct crystallization from the mother liquid and eventually parallel to the silicides, or as a secondary product resulting from the breakdown of silicides in the presence of an increasing concentration of carbon in the surrounding melt.

The Z-liquid may also be formed by the destruction of SiC crystals due to a decreased silicon content in the mother liquid whereby the carbon would be liberated

as a dispersed phase.

The latter view would be based on the assumption that SiC crystals are directly precipitated from the mother liquid. The statics of the phase rule would hardly determine the course which the reactions would follow under the rapidly changing concentrations in the melt and it appears right to allow for both of the two courses when explaining the formation of the Z-liquid.

The nodules are formed by precipitation of graphite on the minute particles of the Z-liquid and this precipitation appears also to have a colloidal character. The direct crystallization of graphite on par-

ticles of SiC or on graphite nuclei formed in the Zliquid should be accepted as a parallel to the precipitation on the Z-liquid.

The nodules are formed in a melt which locally has acquired the condition of a hypereutectic fluid by the mechanism of fractional diffusion. The massive ferrite encountered in nodular cast iron can be ex-

plained only by this fractional diffusion.

The action of magnesium appears to be that of stabilizing the carbides so that a strongly hypereutectic melt may be locally created. A definite minimum content of magnesium in the metal appears necessary to secure this condition. Apart from a nucleating effect, the function of magnesium in the formation of the X-constituent may be due to its carbides or double carbides which on penetration into the high silicon fluid would cause the precipitation of extremely small SiC-crystals as a result of the carbides reacting with the silicides or silicon. Magnesium-silicide would then be formed and retained in solution in the fluid X-constituent.

The inoculation with ferrosilicon appears to have a dual effect: that of creating the X-constituent and that of creating the fractional diffusion.

#### References

1. Samuel Epstein, The Alloys of Iron and Carbon, 1936.

2. H. Gröber and H. Hanemann, Archiv f. d. Eisenhüttenwesen, vol. 11, 1937/38.

- 3. H. Morrogh and H. Williams. J. Br. & Steel Inst., no. 1, 1948.
- Millis, Gagnebin and Pilling. U.S.A. Patents 2.485.760 and 2.485.761.

5. M. G. Corson, Metals Technology, AIME, 1928.

 Haughton and Becker, J. Br. Iron & Steel Inst., vol. 121, 1930.

7. Jass and Hanemann, Giesserei, vol. 25, 1928.

- 8. A. Kriz and F. Poboril, J. Br. Iron & Steel Inst., no. 2, 1930.
- 9. J. E. Hurst and R. V. Riley, J. Br. Iron & Steel Inst., no. 1, 1944.
  - D. Marles, J. Br. Iron & Steel Inst., no. 1, 1948.
     E. D. Harry, J. Br. Iron & Steel Inst., vol. 167, 1951.

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Mr. Howard F. Taylor, associate professor, and Mr. Jack Keverian, research assistant, department of metallurgy Mass. Inst. of Technology, have been kind enough to review the present publication.

#### DISCUSSION

Chairman: J. E. Rehder, Department of Mines & Technical Surveys, Ottawa, Ont. Canada.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

Recorder: C. F. Walton, Case Institute of Technology, Cleveland.

CHAIRMAN REHDER (Written Discussion): This paper is a stimulating one; and although some of the concepts involved cannot be clear-cut at the moment because of lack of data, at least a new approach is developed and some interesting ideas stated. The present writer finds himself in some sympathy with the views of Mr. Hurum, and the remarks to be made are in a constructive direction.

With regard to peculiar features of the microstructures of nodular cast iron, several have been noted by the writer in addition to the still unidentified brownish colored inclusions. The "circular" or ring formation of graphite in nodular iron has been noted frequently, and has been mentioned by De Sy and others in the literature. The films mentioned in the paper have been observed by the writer, although they are usually somewhat thinner than those shown in the paper. A further common occurrence is roughly circular areas which etch much lighter than the surrounding matrix, especially when the matrix is martensitic, and which apparently cannot be explained as a dendritic pattern since the latter is visible independently and is of much lower contrast. It must be confessed that some of the structures shown, such as in Fig. 3 of the paper, have been attributed by the writer to incorrect ladle treatment and insufficient or careless skimming of reaction product from the treated metal, and such correlation is frequently observed.

In Fig. 5 of the paper ternary isotherms are drawn of the Fe-Si-C system, and Mr. Hurum is evidently aware that in view of the unsatisfactory state of the iron-silicon diagram and the complete absence of a silicon-carbon diagram, such constructions

must be regarded as qualitative.

The question of possible existence of silicon carbide in cast irons has interested the writer for some time, and in 1942 calculation of the thermodynamics showed silicon carbide to be very stable at molten iron temperatures, with  $\Delta F = -26,000$  calories at room temperature and -23,600 calories at 1600 C (2912 F). Silicon carbide was identified in samples of 15 per cent ferrosilicon, but many attempts then and since have failed to identify silicon carbide in any cast iron of commercial composition ranges, i.e., under about 3 per cent silicon. It is not considered by the present writer that the presence of silicon carbide in such irons has been proven in the present paper. It may be of interest to note that careful search by chemical and X-ray diffraction techniques has failed to show the presence of silicon carbide in nodules separated from nodular cast iron, or in the ash from such nodules.

If some iron-silicon solution or compound is more stable than silicon carbide, the latter would not be found in such irons, and there is considerable evidence that such is the case. Determination of specific heat-temperature curves for various iron-silicon alloys is all that is lacking for calculation of this possibility, and this data is being obtained in the writer's laboratory. In 1941 Oknov and Moroz (Journal Technical Physics, U.S.S.R. vol. 11, p. 593) reported that an iron rod cemented in granular silicon carbide produced the same results as in ferrosilicon or in a mechanical mixture of silicon and carbon, an example of evidence that iron-silicon is more stable than silicon carbide at least when iron is in considerable excess. On the practical side, the writer has used silicon carbide (in the form of "carborundum fire-sand," a by-product of the manufacture of silicon carbide) extensively as a ladle inoculant in the commercial production of automotive gray cast irons, with excellent results.

The possible existence of a compound Fe_sSi has been of interest and is referred to by Mr. Hurum, Recent opinion is that it does not exist as such in pure iron-silicon alloys, but the question of the effect of carbon is still considered open. The writer showed in a paper ("An Interpretation of the Constitution of Iron-Carbon-Silicon Alloys," A.F.S. Transactions, vol. 55, pp.

77-81, 1947) whether or not Fe, Si exists, all of the available data on ternary compositions and temperatures, at least up to about 4 per cent silicon, can be quantitatively predicted from the ironcarbon diagram simply by assuming that all of the silicon present is held as Fe,Si. This does not prove the existence of Fe,Si, but suggests that the approach should be considered. It might be noted that the narrowing of the liquidus-solidus curves in the iron-silicon diagram in the region corresponding to the composition Fe, Si is in considerable doubt, as examination of the original data will show, and so should not be taken as evidence of existence of a phase.

There are three considerations that appear difficult to explain on the basis of the ideas in the paper, and the author's com-ments would be appreciated. The first is that ladle inoculation of gray iron with ferrosilicon or other graphitizing agents is of maximum effectiveness only when the iron is above a temperature in the region of 2600 to 2700 F, where solution and homogenization would seem to be most rapid. The second is that it is the writer's understanding that a ladle addition of ferrosilicon is of small effectiveness as an inoculant if the ferrosilicon is of high purity, i.e. very low in aluminum content. The third is that a ladle addition of a small quantity of ferromanganese or ferrochromium to a cast iron will inoculate it, i.e., will decrease

the chill depth or produce mottle in a white iron.

E. A. LORIA (Written Discussion): 1 First, I would like to compliment the author on a very excellent and thought provoking paper. In regard to the X-ray examination of the insoluble residues, it is remarkable that type III silicon carbide was observed in the small total quantities extracted from the irons. Type II silicon carbide is much more prevalent. One would have to look at a lot of straight silicon carbide to observe type III, the proportion being roughly 98 per cent type II to 2 per cent type III, The lines that make up the cubic pattern are found in all the hexagonal patterns. Therefore, one could have a mixture of cubic and hexagonal crystals and only by very careful study of the line intensities could they be separated (or distinguished).

The author's work is substantiated by the fact that the M.I.T. investigators, Fulton and Keverian, have established silicon carbide as a stable phase in the Fe-Si-C equilibrium diagram. It appears that the addition of a silicon-bearing material to cast iron cuts through the silicon carbide region of the ternary, which on resolution precipitates graphite, thereby inoculating

A. L. DE SY AND J. VIDTS (Written Discussion): 2 A thorough discussion of the article by F. Hurum would require another article of nearly the same importance; therefore we shall confine ourselves with some remarks.

1. The micrographs appear so different from the thousands nodular iron structures observed by microscopic examination

that it is difficult to judge.

A first glance at some micrographs gives the impression of dirty not decanted iron not completely homogenized. Is there any evidence of the Z-liquid and the X-constituent other than the metallographic indication after rubbing the surface with cotton moistened with alcohol or acetone?

2. The author seems to consider the action of the ferro silicon addition as the primary cause of the nodule formation while this is generally believed as secondary and certainly not abso-

lutely necessary.

According to our knowledge spherulitic graphite results mainly from the addition of Mg, Ce, Li . . . and many other elements, unless the iron was particularly pure and highly superheated.

3. The formation of spherulites by the assembling of colloidal graphite particles is difficult to believe if one considers the

crystallographic structure of spherulitic graphite.

4. The spheroidal shape can not result from partial re-dissolution of irregular graphite nodules; the well known radial structure of the spherulite indicates the growth from the center.

5. According to the author the presence of graphite in the melt should prevent supercooling and provoke solidification

according to the stable system.

As is well known, nodular iron solidifies with supercooling. 6. The author believes that the nodules will shape the pattern of solidification of a hypo-eutectic cast iron. A Wittmoser (1951 —International Foundry Congress) has shown very clearly that the primary austenite dendrites are responsible for the solidification pattern.

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 Laboratorium Voor Metallurgie, Gent, Belgium.

# TIME OF FORMATION OF SPHERULITES IN HYPO- AND HYPER-EUTECTIC IRONS

By

Jack Keverian, Clyde M. Adams, and Howard F. Taylor*

#### ABSTRACT

Since the advent of a cast iron in which the graphite is in spheroidal, rather than flake, form, researchers have been busily studying the mechanism of precipitation of graphite. Cerium added to hypereutectic cast iron spheroidizes the graphite but does not effect hypo-eutectic irons similarly; magnesium serves to nodulize either type. It has been suspected different mechanisms prevail for spheroidizing hypo- and hyper-eutectic irons.

Specimens of hypo- and hyper-eutectic frons, treated with magnesium and post-inoculated with ferrosilicon to nodulize the graphite, were quenched from various temperatures. The results indicate that graphite precipitates in hypereutectic liquid at a temperature above the eutectic temperature; the hypereutectic liquid is thought to be supplied in hypo-eutectic irons (1) as a result of microsegregation of silicon and carbon during formation of austenite dendrites, and /or (2) as a result of post-inoculation with ferrosilicon.

## Introduction

Many points, in the field of spherulitic cast iron, are in need of clarification before the more dramatic problem of the actual mechanism of formation of graphite spherulites can be solved. One such point is determination of the time during solidification when the graphite nodules form. Examination of centrifugal castings made from hyper-eutectic irons exhibit nodules collected near the inner diameter of the casting; so it seems fairly well-established nodules precipitate in the liquid above the eutectic temperature. However, this same segregation is not observed in hypo-eutectic irons similarly cast. This lack of segregation for hypo-eutectic irons does not necessarily indicate nodules are only formed during or after the eutectic freezing; it is possible the graphite might precipitate above the eutectic and primary austenite dendrites, acting as barriers, block movement of the nodules through the liquid.

The large shrinkage associated with spherulitic cast iron (as compared with flake graphite iron) indicates that much of the graphite precipitates (1) above, (2) at the start of, or (3) below the eutectic, or com-

binations thereof. In ordinary gray iron, volume expansion due to graphite precipitation, at the eutectic, tends to counterbalance solidification shrinkage. From the equilibrium diagram one would predict graphite, in hypo-eutectic irons, does not appear until the melt reaches the eutectic temperature. At higher temperatures the melt can only be in equilibrium with primary austenite. This, in turn, leads one to think graphite precipitates below the eutectic freezing via carbide decomposition and/or that the graphite precipitates at an early stage of eutectic freezing. Existence of a mechanism for solid-state nucleation and growth to account for spherulite formation in hypoeutectic irons, would be in sharp contrast to liquidstate nodule formation for hyper-eutectic irons. These sharply differing modes for spherulite formation seem unlikely. This paper will seek to show that only one mode of formation actually exists to account for spherulite formation in hypo- and hyper-eutectic cast

# **Previous Investigations**

Morrogh¹ describes solidification of hyper-eutectic alloys and shows graphite spherulites nucleate as a pro-eutectic constituent. The eutectic phase then freezes according to the metastable diagram with subsequent graphitization of eutectic. Morrogh does not describe the solidification process for hypo-eutectic compositions.

DeSy² reports graphite nodules precipitate as a primary phase in both hypereutectic and hypo-eutectic alloys. In hypo-eutectic alloys he theorizes nodules either precipitate in a supersaturated primary austenite dendrite, or at the austenite-melt interface; DeSy made no definite choice between the modes of formation.

Larson³ quenched portions of magnesium-treated, hypo-eutectic iron by sucking the metal into silica tubes and quickly plunging them into water. Metalographic examination of samples quenched from different temperatures revealed graphite nodules nucleated above the eutectic temperature.

Dunphy and Pellini⁴ recently quenched specimens of *hypo-eutectic* irons from various temperatures along the cooling curve, and also found evidence of

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This paper is a part of work submitted for Degree of Master of Science for J. Keverian.

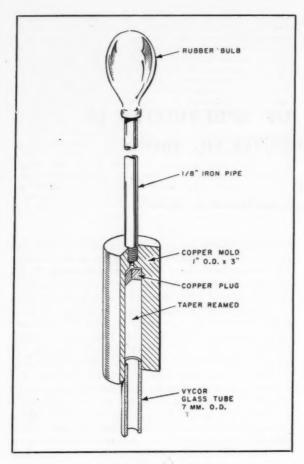


Fig. 1-Copper chill.

graphite nucleated in the melt, immediately preceding the eutectic reaction. They attributed this to existence of localized regions of hypereutectic liquid in the vicinity of growing dendrites of primary austenite. The work reported herein was done simultaneously with and independently of their work, so the substantiation is all the more pertinent. The findings are in general agreement with those of Dunphy and Pellini.

## **Experimental Procedure**

Specimens of molten hypo- and hyper-eutectic irons were quenched from various temperatures to determine the time during solidification when graphite spherulites appeared. It was thought quenching rates used to date by the various investigators might not be sufficiently drastic to retain the high-temperature phases; so some refinements in method were intro-

One method used was to suck small samples of molten metal into a copper chill mold (Fig. 1) of the type used by Carney and Chipman⁵ to sample steel for hydrogen analyses. This method proved quite satisfactory, since very high quenching rates were achieved and several specimens could be extracted from the same heat of metal, thus eliminating any composition variable of the base metal. The obvious

disadvantage was that specimens could not be quenched during late stages of solidification when small amounts of liquid remained. It is recognized this method emphasizes the liquid phase over the solid. Although austenite dendrites were captured, many more were surely left behind in the casting. This meant the relative position of the solid phases at the time of quenching is not rigorously maintained, since there doubtless was some relative motion between these phases during suction of the melt into the copper mold.

To allow equal emphasis of liquid and solid phases, the apparatus shown in Fig. 2 was built; small heats of synthetic cast iron (about 50 grams) could be quenched whether liquid or solid. The metal in the alundum crucible (A) was melted by induction, using an 8 kva Ecco Spark Gap Convertor. A positive pressure of argon was maintained within the glass enclosure to prevent oxidation. Temperatures were measured by optical pyrometer. When the metal reached 1260 C (2300 F) the inoculator was lowered into position and the inoculant (usually cerium) was introduced. A post-inoculant was added (in the form of 99.8 per cent silicon) to raise total silicon by 0.5 per cent.

Cooling curves were obtained by replacing the inoculator with a Pt, 10% Pt-Rh thermocouple. The thermocouple wire was threaded through a three millimeter, two-holed, alundum insulator tube.

A-ALUNDUM CRUCIBLE CONTAINING MELT

C-ALUNDUM TUBE

E-CRUSHED K-25 INSULATING BRICK F-INSULATING BRICK

H-TRANSITE COVER

I - SAUEREISEN SEAL

K-RUBBER SEAL L-GLASS BATTERY JAR 12" LONG - 9 1/2" 0.0.

M-ASBESTOS PAPER N-RUBBER JOINT

O-TWO-HOLED STEEL CONNECTOR

Q-RUBBER CONTAINER FOR INOCULANT

R-GLASS WINDOW FOR OPTICAL PYROMETER

STEEL SHELL FOR QUENCH BATH

V-ASBESTOS PAPER W-STEEL MELTING POT X-BINDING POSTS

Z-WOOD'S METAL

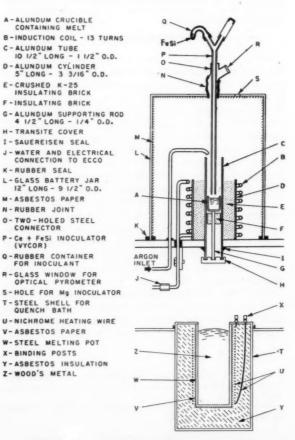


Fig. 2— ace and quenching arrangement.

thermocouple bead was protected by a short, thinwalled, alundum tube closed at one end. This protection tube fitted loosely over the bead so it could fall readily from the thermocouple when the melt was released for quenching. The protection tube was long enough only to be almost wholly immersed in the melt to minimize loss of heat by conduction away from the bead.

At any desired moment, the trap door (H) could be opened and the crucible would plunge into the quenchant, leaving the thermocouple behind. Wood's metal was chosen for the quenching medium because its heat conductivity is high, surface effects would be small, and its melting point is low (about 70 C). All these factors contribute to highly efficient quenching action.

# Specific Tests on Hypo-Eutectic Alloys

The copper chill mold described above was used to obtain quenched samples at various stages of solidification from a magnesium-treated hypo-eutectic iron. A post-inoculant of ferrosilicon was used, which increased total silicon by 0.5 per cent. Specimens were extracted at 1290, 1270, 1220, 1180, 1170 C (2350, 2318, 2228, 2156, 2138 F).

Metallographic examination allows one to distinguish between the solid and liquid portions of the specimen at the time of quench. This is possible since the liquid phase undergoes a high cooling rate as compared to the pre-existent solid phase which was formed under conditions of slow cooling. When the liquid freezes quickly, a fine structure of cementite and austenite is formed. This highly dispersed structure is, therefore, easily distinguished from the larger austenite dendrites that existed in the melt before the quench.

Well-formed spherulites were found in the specimens quenched from 1180 and 1170 C, approximately 25 and 15 C above the eutectic temperature. A graphite spherulite is shown in Fig. 3, surrounded by a matrix of ledeburite, indicating the spherulite was suspended in liquid at the time of quench. Figure 4

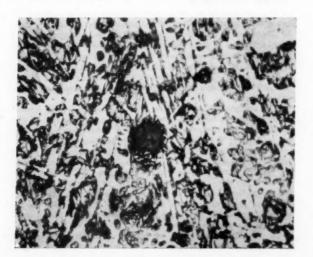


Fig. 3—Hypo-eutectic iron quenched from 1180 C (2155 F). 500x.



Fig. 4—Hypo-eutectic iron quenched from 1170 C (2140 F). 500x.

shows a spherulite completely surrounded by primary austenite.

The fact that many spherulites were not surrounded by primary austenite, indicates the role of austenite is not essential for their formation. This evidence indicates the spherulite is born in the liquid, perhaps in close proximity to the primary austenite dendrite and subsequently engulfed by the advancing austenite front. Unless the small additions of magnesium causes a drastic change in the equilibrium diagram, which seems unlikely, one may hypothesize the formation of primary spherulites in hypoeutectic irons is a non-equilibrium process, which would depend upon the existence of localized hypereutectic pools in the overall hypo-eutectic melt. These localized hypereutectic pools could be generated as a result of the ferrosilicon post-inoculation, or by the microsegregation of silicon and carbon already present in the melt, or by a combination of both mechanisms. It is thought that silicon carbide is precipitated as the ferrosilicon dissolves in the melt. Once precipitated, silicon carbide has a low rate of re-solution, thereby maintaining high silicon levels on a micro-scale for rather long periods of time. These high silicon volumes could easily be responsible for the rejection of carbon from solution even at temperatures above the eutectic. The presence of silicon carbide as a stable phase in the Fe-C-Si constitutional diagram has recently been established by Fulton.6 Hurum7 has also reported the existence of silicon carbide as a stable phase, and has set forth a mechanism for formation of localized hypereutectic pools as a result of ferrosilicon inoculation. This is supported by the fact that silicon carbide itself is known to be a good graphitizer. The evidence is that microsegregation alone can be sufficient mechanism to generate hypereutectic pools, since hypo-eutectic iron can be made with spheroidal graphite without the aid of post-inoculation. Post-inoculation increases the uniformity of structure and reduces the amount of gross carbide, possibly by increasing the number of hypereutectic pools.

It is recognized that the quenching itself could be responsible for some graphite precipitation, as reported by Morrogh. These quench-formed spherulites would be expected to form from the melt or from the decomposition of cementite. It is difficult to imagine how quenching could cause the precipitation of nodules within a slowly cooled austenite dendrite. This strengthens the observation that primary graphite is forming independently of the quench.

# Specific Tests on Hypereutectic Alloys

Quenching experiments were made using the apparatus of Fig. 2. Hypereutectic alloys were quenched from various temperatures above the eutectic. These experiments confirmed the commonly accepted belief that graphite nodules start to precipitate in the melt above the eutectic temperature. No evidence was found to support the theory graphite spherulites result via decomposition of cementite. A typical microstructure is shown in Fig. 5.

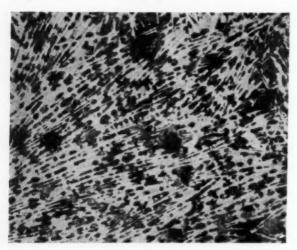


Fig. 5—Hypereutectic iron quenched from 1164 C (2095 F). 750x.

#### Summary

Evidence has been presented that indicates spherulitic graphite precipitates in the melt above the eutectic temperature for both hypo- and hyper-eutectic alloys. Since the same mode of formation persists for both these irons, a separate theory for the mechanism of formation need not be devised, and only one theory need be sought.

Since graphite is not in equilibrium with the melt and the austenite, at temperatures above the eutectic for hypo-eutectic irons, one must assume, in this case, nucleation and growth of graphite is not a problem of equilibrium.

It is suggested that microsegregation of silicon and carbon during solidification and the ferrosilicon postinoculation could generate localized pools of hypereutectic liquid causing the rejection of graphite.

# **Bibliography**

1. H. Morrogh and W. J. Williams, "Graphite Formation in Cast Iron and Nickel-Carbon and Cobalt-Carbon Alloys," *Journal*, Iron and Steel Institute, vol. 155, 1947.

 A. DeSy, "Spherulitic Formation in Nodular Cast Iron," Metal Progress, vol. 59, no. 6, June 1951, pp. 798-808.

3. W. L. Larson, "The Formation of Spherulitic Graphite in Cast Irons," Doctor's Thesis, Mass. Inst. of Tech., Dept. of

Metallurgy, May 1950.

 R. P. Dunphy and W. S. Pellini, "Nodule Genesis and Growth in Magnesium-Treated Hypo-eutectic Irons," Advanced Copy of Report from Metal Processing Branch of Naval Research Laboratory, July 1951.
 D. J. Carney, "The Analysis and Sampling of Liquid Steel

5. D. J. Carney, "The Analysis and Sampling of Liquid Steel for Hydrogen," Doctor's Thesis, Mass. Inst. of Tech., Dept. of

Metallurgy, 1949.

 Private Communication from J. C. Fulton, Master's Thesis, Mass. Inst. of Tech., Dept. of Metallurgy, 1951.

7. F. Hurum, "A Study of the Formation of Nodular Graphite," A.F.S. Transactions, vol. 60, pp. 834-848 (1952).

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#### DISCUSSION

Chairman: J. E. Rehder, Department of Mines & Technical Surveys, Ottawa, Ont., Canada.

Co-Chairman: V. A. CROSBY, Climax Molybdenum Co., Detroit.

Recorder: C. F. Walton, Case Institute of Technology, Cleveland.

E. A. LORIA (Written Discussion): ¹ The authors are to be commended for an excellent paper. Their evidence is so conclusive that it is difficult to question any phase of their work. Pellini's recent study indicates that graphite nodules may form during the short undercooling period of eutectic solidification, whereas Larson found that the nodules are nucleated above the eutectic temperature. Since both cerium and silicon were used as inoculants, did the authors consider or investigate the possibility that these elements (also magnesium) may have raised the eutectic arrest?

In the case of gray irons, the type of graphite flake structure produced in a casting is primarily the result of the length of time the casting spends in going through the eutectic arrest. The longer the time taken to go through the eutectic arrest, the coarser will be the graphite structure. Notwithstanding, type D, or eutectiform type, graphite is not necessarily a characteristic of a low eutectic arrest. Type D graphite can be found at any degree of undercooling provided the solidification rate of the melt is fast enough.

The mechanism of inoculation in gray iron is believed by many to be physical rather than chemical. The microsegregation of silicon and carbon during solidification and the generation of localized pools of hypereutectic liquid causing the rejection of graphite is an interesting possibility. The writer would certainly appreciate the inclusion of some of Fulton's data on this subject.

MR. TAYLOR (Authors' Closure): The authors would like to thank the discussors for their stimulating comments regarding

this paper.

It is recognized that cast iron is in its simplest form a ternary alioy of iron, carbon and silicon. In the more general case, cast iron is a multi-component system. For these reasons, it would be misleading to speak of the eutectic temperature as referred to in the ordinary binary alloy system. In the case of the binary alloy, the eutectic temperature is an invariant temperature in an isobaric system during which three-phase equilibrium exists. However, in the case of a ternary alloy, we have one additional degree of freedom and three-phase equilibrium can exist over a range of temperatures. In our case, we have referred to the eutectic temperature as being that temperature at which a third equilibrium phase appears.

In the plain iron-carbon system, the eutectic reaction takes place at a constant temperature. Unfortunately, there is still some doubt as to the exact equilibrium binary eutectic tempera-

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ture. As silicon is added, the eutectic reaction occurs over a range of temperatures and also starts at a slightly higher temperature. Since the absolute values of these equilibrium temperatures are in doubt, perhaps the temperatures reported in the paper have more significance when reported relative to the temperature of the start of the eutectic mold.

When this is done, it is observed that graphite spherulites are found in specimens quenched 25 and 15 degrees Centigrade

above the start of the eutectic hold.

The small percentages of cerium and magnesium which were added did not seem to affect the temperature of the eutectic hold. Non-equilibrium considerations were therefore sought to

explain the observations.

Although it would be desirable to include some of J. C. Fulton's data in this paper, this is not possible at this time. However, most of his data have been presented in the form of a thesis entitled, "The Solubility of Graphite in Molten Iron-Silicon Alloys," which can be obtained through the M.I.T. library.

Mr. Pellini has suggested that the graphite spherulites found in those specimens quenched from above the eutectic temperature, were quench-formed. The fact that many spherulites were found surrounded by primary austenite when quenched from the same temperatures indicates that graphite was not necessarily quench-formed, since one would expect the graphite to be surrounded by a ledeburite matrix.

The foregoing then indicates that graphite forms in the liquid before the alloy reaches the eutectic temperature. This graphite may then be engulfed by the advancing austenitic

front.

H. Morrogh suggests that it is necessary to consider the binary complex which was done as explained at the start of the authors' closure. As was pointed out, not even this approach is satisfactory and non-equilibrium reactions must be considered.

It is encouraging to note that H. Morrogh has also observed pro-eutectic spherulites in hyper-eutectic irons. However, we cannot directly confirm his observations of the precipitation of austenite around the pro-eutectic spherulite. Such an observation is difficult due to the difficulty in distinguishing between that austenite that might be present at the time of quench, and that austenite that is formed during the quench as one of the phases making up the ledeburite constituent.

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